The cover illustration shows a CCT diagram with the different microstructures occurring during quenching. On the left are surface cooling curves and, on the right, centre cooling curves with a typical component for heat treatment superimposed.
Steel and its Heat Treatment

Bofors Handbook

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Preface

The need for a handbook, which provides practical information together with references to theoretical work, continues to increase in steel-consuming industries. *Steel and its Heat Treatment: Bofors Handbook* has been written to meet this need and especially to bridge the widening gap between practical and theoretical knowledge.

The original Swedish edition (1967), which was received with great interest, was written on the initiative of Mr Tord Krej, Director of Research of A B Bofors and was sponsored by the Swedish Metallographers Society. The contents of that edition were checked by Mr Sven Arwidson, Chief Metallographist of A B Bofors, together with Mr Göran Larsson and Mr Kurt Walter.

In the years since the publication of the Swedish edition the author has been in contact with several readers and their opinions have influenced the contents of the present edition, which has also been carefully revised and supplemented to meet both the demand for recent information and the requirements of students.

This English edition has been checked by Professor Bertil Aronsson, Head of the Swedish Institute for Metal Research, Dr J. A. Charles, University of Cambridge, and Dr Tom Bell, University of Liverpool. The translation was made under the supervision of the Metals Society, formerly The Iron and Steel Institute, mainly by Mr Cecil M. Black.

The author wishes to thank all persons and institutions mentioned above and all others who have assisted or contributed to the publication of this English edition of the *Bofors Handbook*.

Bofors 1974

Karl-Erik Thelning

International Designations and Symbols

INTERNATIONAL SYSTEM OF UNITS

On the 1st of January 1971 the metric system was legally in force in the U.K. For the majority of technicians this involved an adjustment from inches to millimetres. For several years, work has been in progress to devise a common standard international system of units. Such a system, SI (Système International d'Unités), was adopted in 1960.

The following units should be used in ISO Recommendations prepared under the jurisdiction of ISO/TC 17:

1. Stress — N/mm²
2. Hardness — There should be no change in the present practice in that the load applied to the indenter should still be expressed in terms of the unit kgf.
3. Impact — In general it was agreed that the joule should be used, but in some cases the Nm might be a preferable unit.

During an interim period several European countries have used the symbol kp/mm² (kilopond) or kgf/mm² (kilogramme-force) to signify the unit of stress.

The mechanical strength of steel was previously designated ton/in² (TSI) in the U.K and kg/mm² on the Continent.

The present book has been written and published during the period of transition covering the introduction of SI units both in the U.K and other countries that have adopted it. In several diagrams in this book stress is designated kp/mm² and in some tables, kgf/mm² in accordance with some editions of ISO's recommendations. (B.S. 970:1970 gives kgf/mm² as the designation for stress.)

In order to simplify the transition to SI units most diagrams have been drawn with double scales, e.g. inches—millimetres, kp/mm²—TSI—N/mm²
**INTERNATIONAL DESIGNATIONS AND SYMBOLS**

and even Celsius (°C)—Fahrenheit (°F). The symbol joule is not always adopted here since, the final decision will be taken at the end of 1974.

In some tables two systems of units are used. For the conversion of inches to millimetres the factor 25.4 has often been used since the small error introduced thereby is of no practical consequence. For other, more precise applications, such as for Jominy diagrams, the exact conversion factor has been used.

In other ways, too, an international outlook is favoured, viz. the symbols for hardness units, e.g. HB and HRC. In line with this principle the symbol HV is used instead of DPH or VPN. Conversion tables and nomograms are found in Chapter 8.

In connection with passing over to SI units, according to the ISO standard a number of designations for mechanical testing has been changed. What is characteristic of this transition is that certain designations in Greek letters have been replaced by Latin ones. The designations generally used for steel are indicated below, partly old ones, partly according to the new standard.

An example for the use of SI units is given at the same time.

### Old standard

<table>
<thead>
<tr>
<th>$\sigma_{0.2}$</th>
<th>$\sigma_{B}$</th>
<th>$\sigma_{S}$</th>
<th>$\psi$</th>
<th>HB</th>
<th>KV</th>
<th>KCU</th>
</tr>
</thead>
<tbody>
<tr>
<td>kPa/mm²</td>
<td>kPa/mm²</td>
<td>%</td>
<td>%</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>54</td>
<td>81</td>
<td>19</td>
<td>61</td>
<td>249</td>
<td>7.0</td>
<td>9.2</td>
</tr>
</tbody>
</table>

### New standard—SI

<table>
<thead>
<tr>
<th>$R_{p0.2}$</th>
<th>$R_{m}$</th>
<th>A $t$</th>
<th>Z</th>
<th>HB</th>
<th>KV</th>
<th>KU</th>
</tr>
</thead>
<tbody>
<tr>
<td>N/mm²</td>
<td>N/mm²</td>
<td>%</td>
<td>%</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>530</td>
<td>790</td>
<td>19</td>
<td>61</td>
<td>249</td>
<td>69</td>
<td>45</td>
</tr>
</tbody>
</table>

In old ISO documents cited in this book the proof stress is designated by $R_{p}$. The designation $R_{p0.2}$ was adopted in 1973.

**INTERNATIONAL STEEL DESIGNATIONS**

Under the auspices of ISO extensive work is in progress on the standardization of steel grades, in particular with respect to composition and mechanical properties. ISO recommendations covering a large number of steel grades have already been published. Among them may be mentioned the group 'heat-treated steels, alloy steels and free-cutting steels'. The work of standardization for tool steels should be completed by 1976.

The tables covering 'Surveys of various types of steel' contain the standards as published by AISI, BS, DIN and SIS along with such ISO standards as have been issued.

In the text, tool steels are designated only by the type letter and numeral as used in the USA and the UK for standardized tool steels, e.g. H13, O1. These designations are so well known by steel consumers all over the world that no qualifying institutional designations are necessary. Steels for which there are no AISI or BS specifications are designated according to DIN or SIS standards.
1 Fundamental metallographic concepts

Metallography reveals the structure of metals and leads to a better understanding of the relationship between the structure and properties of steel. With the aid of modern developments such as the electron microscope and the scanning electron microscope it is now possible to obtain a much deeper insight into the structure of steel than was possible only some ten years ago.

In order to understand the processes occurring during the heat treatment of steel, it is necessary to have some knowledge of the phase equilibria and phase transformations which occur in steel as well as of its microstructure. Therefore, a brief summary of these topics is given in this chapter which forms the groundwork for subsequent discussion.

1.1 THE TRANSFORMATIONS AND CRYSTAL STRUCTURES OF IRON

On heating a piece of pure iron from room temperature to its melting point it undergoes a number of crystalline transformations and exhibits two different allotropic modifications. When iron changes from one modification to another heat is involved. This is called the latent heat of transformation. If the sample is heated at a steady rate the rise in temperature will be interrupted when the transformation starts and the temperature will remain constant until the transformation is completed. On cooling molten iron to room temperature the transformations take place in the reverse order and at approximately the same temperatures as on heating. During these transformations heat is liberated which results in an arrest in the rate of cooling, the arrest lasting as long as the transformation is taking place.

The two allotropic modifications are termed ferrite and austenite and their ranges of stability and transformation temperatures on heating and cooling are shown in Figure 1.1. The letter A is from the French arrêt, meaning to delay, c from chauffer, meaning to heat and r from refroidir, meaning to cool. Ferrite is stable below 911 °C as well as between 1392 °C and its melting point, under the names α-iron and δ-iron respectively.

![Figure 1.1. Heating and cooling curve for pure iron](image)

Ferrite, besides having an atom at each corner of the unit cell, has another atom at the intersection of the cube body diagonals, i.e. a body-centred cubic lattice (BCC). The length of the unit cube edge or lattice parameter is 2.87 Å at 20 °C (Å = Ångström = 10^-10 cm). Austenite has a face-centred cubic lattice (FCC), the parameter of which is 3.57 Å (extrapolated to 20 °C). The structure of the unit cells of α-iron and γ-iron respectively may be envisaged as shown in Figure 1.2. The γ-iron unit cell has a larger lattice parameter than the α-iron cell but the former contains more atoms and has a greater density, being 8.14 g/cm³ for γ-iron at 20 °C and 7.87 g/cm³ for α-iron.
4 FUNDAMENTAL METALLOGRAPHIC CONCEPTS

1.2 THE IRON–CARBON EQUILIBRIUM DIAGRAM

The most important alloying element in steel is carbon. Its presence is largely responsible for the wide range of properties that can be obtained and which make this metal such a highly useful commodity of everyday life. At room temperature the solubility of carbon in α-iron is very low and therefore the carbon atoms are to be found only very infrequently in between the individual iron atoms. Instead the carbon is combined with iron carbide, also called cementite, Fe₃C. The iron carbide may be present as lamellae alternating with lamellae of ferrite, which together form a constituent called pearlite, the mean carbon content of which is 0.80%. The proportion of pearlite in the
structure increases with the carbon content of the steel up to 0.80%\%. Carbon in excess of this amount separates as grain-boundary carbides. A steel containing 0.80%\% carbon is said to be eutectoid (see Figures 1.3a–d).

When iron is alloyed with carbon the transformation will take place within a temperature range which is dependent on the carbon content as shown in the iron–carbon equilibrium phase diagram. Figure 1.4 illustrates various microstructures appropriate to that part of the iron–carbon diagram which applies to steel heat treatment. For the sake of completeness the phase diagram is reproduced as far as 6\% carbon in Figure 1.5 where it can be observed that the solubility of carbon is much greater in austenite than in ferrite.

### 1.2.1 Heating

It was mentioned earlier that \(\alpha\)-iron transforms to \(\gamma\)-iron on being heated to 911°C. This can be seen by looking at the vertical axis of the left-hand part of the diagram in Figure 1.4. In a steel containing 0.80\% carbon, i.e., a eutectoid steel, the transformation to austenite takes place at about 723°C. The temperature at which \(\alpha\)-iron, \(\gamma\)-iron and cementite are at equilibrium is designated \(A_1\) (line PK in diagram). Steels with less carbon, which are called hypo-eutectoid steels, begin to transform from pearlite to austenite at the same temperature, viz. 723°C. In the equilibrium region between PS and GS there is austenite, formed from pearlite, and unchanged ferrite. The transformation is not complete until a temperature \(A_3\) given by the line GS is reached. Above this line there is only one stable phase, viz. austenite. If the carbon content is more than 0.80\%, the steel is said to be hyper-eutectoid. In these steels, too, the pearlite transforms into austenite at 723°C but the cementite (iron carbide) does not go into solution completely until the temperature rises above the equilibrium line SE, designated \(A_{cm}\).

Let us for a moment return to the course of events taking place during heating. At 723°C we find that the transformation to austenite begins to take place in steels having more than 0.025\% carbon. This means that the atomic configuration changes from ferrite to austenite in which the carbon atoms are more soluble. The positions of carbon atoms in austenite are illustrated in Figure 1.6. At temperatures above G-S-E there is austenite only, all the carbon having been dissolved and evenly distributed throughout the austenite.

### 1.2.2 Cooling

As the temperature of a fully austenitized eutectoid plain carbon steel is slowly lowered below 723°C the transformation from \(\gamma\)-iron to \(\alpha\)-iron begins to take place and as a consequence the carbon is forced out of the lattice and forms cementite. On complete cooling to room temperature the steel has once again its pearlitic structure.

### 1.3 TIME–TEMPERATURE–TRANSFORMATION

The iron–carbon equilibrium diagram is unquestionably of fundamental importance to heat-treated processing. However it only describes the situation when equilibrium has been established between the components carbon and iron. In the great majority of heat treatments the time parameter
is one of the determinative factors, the influence of which is shown by so-called time-temperature-transformation diagrams. From these diagrams it is possible to follow the effect of both time and temperature on the progress of transformation.

1.3.1 Heating

The influence of time is best explained by means of the diagrammatic illustrations in Figures 1.7a-f. Figure 1.7a shows the familiar iron-carbon diagram in which it is seen that a 0.80% carbon steel, on being heated, transforms to austenite at 723°C. However, the diagram tells us nothing about how long this transformation will take. From Figure 1.7b, which applies to a 0.80% carbon steel only, it can be predicted that when the temperature is maintained at 730°C the transformation will be initiated in about 30 s (logarithmic time scale). If instead the steel is rapidly heated to 750°C the transformation will begin in 10 s and if heated to 810°C, in slightly over 1 s, i.e., practically at once on reaching this temperature. The transformation of pearlite to austenite and cementite is completed in about 6 s at 810°C. If the steel is to be fully austenitic it must be held at this temperature for about 5 h. In actual practice, i.e., when heating for hardening, the cementite is seldom brought into complete solution.

Figures 1.7c and d which apply to a 0.45% plain carbon steel show in the same way that at 810°C, for instance, the transformation from pearlite to austenite starts almost immediately. In about 5 s the pearlite has been transformed and the structure consists of ferrite, austenite and cementite. About 1 min later the carbon has diffused to the ferrite which has thereby been transformed to austenite. Residual particles of cementite remain, however, and it takes about 10 h to dissolve them completely.

On heating a hyper-eutectoid steel containing 1.2% carbon to 810°C a structure consisting of austenite and cementite is obtained in about 5 s (Figures 1.7e and f). It is not possible for the cementite to be completely dissolved at this temperature, this is apparent from a perusal of the equilibrium diagram. In order to effect complete solution of the cementite the temperature must be increased to 860°C at least.

In order to study the rate of solution the steel is heated to a predetermined temperature, and after holding it there for a certain time it is quenched in water thereby "freezing" the pre-existing structure. By this treatment, however, the austenite is transformed to martensite (for a discussion of this see p.15).

The result obtained from such an experiment, applied to a 0.45% carbon steel, is shown in Figure 1.8. In all cases the holding time at temperature was 5 min. Figure 1.8a shows the original structure which consists of about 50% of ferrite and 50% of pearlite. On being heated to 725°C for 5 min some of the cementite lamellae are converted to spheroids. This is called spheroidization. A decrease in hardness is a result of this process. There is no transformation (see also Figure 1.7). When it is heated to 735°C (Figure 1.8c) the main part of the pearlite is transformed to austenite which, on being quenched, forms martensite. Some ferrite and pearlite remain untransformed. A holding time

---

Figure 1.7. Structural transformations in heating steels containing: (a) 0.80% C; (b) 0.80% C; (c) 0.45% C; (d) 0.45% C; (e) 1.2% C; (f) 1.2% C. Schematic representation (after Rose and Streefland)

A: austenite, B: bainite, C: cementite, F: ferrite, P: pearlite
Figure 1.8. Microstructure of a 0.45% C steel after various heat treatments:
(a) Untreated. Hardness 220 HV; (b) Heated to 725°C in salt bath. Holding time 5 min. Quenched in water. Hardness 215 HV; (c) Heated to 735°C in salt bath. Holding time 5 min. Quenched in water. Hardness 376 HV.
of 5 min at 750 °C (Figure 1.8d) is sufficient to transform all the pearlite but about 3% of the ferrite remains unchanged. However, 5 min at 775 °C is not sufficient to transform all the ferrite, as is apparent in Figure 1.8e which shows traces of this constituent. At 825 °C, on the other hand, all the ferrite has been transformed in 5 min. Cementite, however, is still in evidence, which is in agreement with Figure 1.7.

Varying the rate of heating to the hardening temperature will have an effect on the rate of transformation and dissolution of the constituents. Figure 1.9 shows a continuous-heating diagram for a steel of almost eutectoid composition and consisting originally of ferrite and pearlite. The third curve from the right represents a heating rate of about 3° per minute. The temperature of transformation increases as the rate of heating increases. The temperature for complete solution of the residual carbide was determined dilatometrically (i.e., by change of dimension accompanying phase changes) and differs therefore somewhat from Figure 1.7, the data for which were determined metallographically.

A complete set of continuous-heating curves for various grades of steel would be of great help in the field of practical heat treatment. Since such diagrams are generally not available we must be content with simple

Figure 1.9. Transformation diagram for continuous heating. Dissolution of ferrite and lamellar pearlite in a 0.70% C steel (after Ross and Straussberg).
estimations of heating and holding times. However, a commendable achievement in this field has been accomplished by the issuing of Atlas zur Wärmebehandlung der Stähle, Volume 3 in 1973.

1.3.2 Cooling

The general appearance of the structure created during cooling is dependent on the temperature of transformation and on the time taken for the transformation to start. As in the case of heating, the iron-carbon equilibrium diagram can tell us nothing about this. In a manner similar to that described above, the transformation of steel at a certain temperature may be investigated by cooling it from the austenitic state to the temperature concerned, letting the transformation take place and then quenching to room temperature. The structure thus obtained is then studied under the microscope.

1.3.3 Formation of pearlite

When a eutectoid steel is cooled from an austenitizing temperature of, say, 850°C to 750°C then, according to the iron-carbon equilibrium diagram no transformation will take place. If the temperature is lowered to 650°C pearlite will start to form after 1 s and the transformation will be completed in 10 s (see curve II in Figure 1.10a). As the temperature of pearlite formation is lowered the pearlite lamellae become increasingly finer and the whole structure becomes harder. If we allow the transformation of the hypoeutectoid steel in Figure 1.10b to take place at 750°C, only ferrite separates and a state of equilibrium is established between ferrite and austenite (curve I). If the transformation takes place at 650°C ferrite separates first followed by a short interval by pearlite. Similarly, in the case of the hypereutectoid steel, Figure 1.10c, cementite separates first followed by pearlite.

Pearlite formation is initiated at the austenite grain boundaries or at some other disarray in the austenite grains. The process has been studied in detail by Hillert who found that pearlite formation can be initiated on either ferrite or cementite and that pearlite growth proceeds by branching. Platelets of cementite and ferrite grow in juxtaposition since carbon transport from the austenite to the edges of the cementite platelets results in a simultaneous carbon impoverishment of the edges of the ferrite platelets. Figure 1.11 shows how pearlite grows according to this model.

1.3.4 Formation of bainite

At temperatures below about 550°C another constituent, bainite, starts to separate along with the pearlite. Its formation is assumed to be initiated on ferrite nuclei which grow as platelets from the grain boundaries. The carbon content of the surrounding austenite increases continuously and when it has reached a limiting value platelets of cementite form in juxtaposition with platelets of ferrite.

As the temperature falls bainite begins to form inside the grains as well; at the same time the mode of formation changes. In the metallurgical microscope it may be difficult to differentiate bainite from other constituents since bainite alters its appearance according to its temperature of formation and the composition of the steel. Figure 1.12 shows bainite in a chromemanganese steel.

Depending on the temperature of formation of bainite it is classified as

![Figure 1.11. Schematic representation of pearlite growth (after Hillert) ![Figure 1.12. Bainite 500 x](image.png)
upper or lower bainite. The mode of formation of the various types of bainite as well as their properties have been described by Pickering. For the present purpose it is sufficient to indicate that upper bainite is relatively brittle and lower bainite is tough.

1.3.5 Formation of martensite

Referring to Figures 1.10a–c, if cooling takes place as represented by curve V, i.e. very rapidly, austenite will start to transform to ferrite on reaching line $M_f$. As the cooling continues below $M_f$, there is very little carbon migration while the austenite is transforming. Thus, the carbon atoms remain in solid solution in the $\gamma$-iron. Since the space available for the carbon atoms is less in $\gamma$-iron than in $\gamma$-iron the carbon atoms will expand the lattice. The resulting state of stress increases the hardness of the steel. We say that the steel has been hardened. The new constituent, called martensite, is a super-saturated solution of carbon in $\gamma$-iron.

The mechanism of martensite formation is the subject of considerable controversy and several theories have been advanced. Figure 1.13 illustrates a simple model of how we can picture the transformation of $\gamma$-iron to $\alpha$-iron during martensite formation. The carbon atoms situated on the edges of the martensite unit cell cause the unit cell to increase in one direction, which results in a tetragonal lattice. Even in the case of high-carbon steels only a very small fraction of the number of possible lattice sites are occupied by carbon atoms. The volume of the martensite increases with increasing carbon content.

From the transformation diagrams it will be seen that the formation of pearlite and bainite progresses with time whereas that of martensite does not. Each temperature below $M_f$ corresponds to a definite proportion of martensite but the amount actually formed depends on the grade of steel, the conditions of the austenitizing treatment and the rate of cooling on quenching for hardening. As the temperature falls it is possible to follow visually the martensitic transformation in a so-called hot-stage microscope. Figures 1.14a–f show the progress of martensite formation. It starts at 220°C and at 175°C the main part of the austenite has been transformed to martensite.

Figure 1.13. Simple model for the transformation of austenite ($\gamma$) to martensite ($\alpha$)

Figure 1.14. Transformations of austenite to martensite at successively lower temperatures: (a) 220; (b) 220; (c) 200; (d) 195; (e) 180 and (f) 175°C
The steel used is a high-alloy steel, designated MAR 2, in which the martensite is composed of nickel and iron.

Marder and Krauss have studied the structure of martensite in iron-carbon alloys and have found two different types called massive martensite and acicular martensite. The range of composition in which only massive martensite exists extends up to 0.6% carbon. This type of martensite consists of packets of parallel platelets (Figure 1.15a), which can only be resolved in

![Figure 1.15. (a) Massive martensite, (b) Acicular martensite](image)

the electron microscope. The more familiar acicular martensite is found only in steels containing 1% carbon and more. The structure of quenched high carbon steels consists of irrationaly arranged plates in a matrix of austenite

1.3.6 Retained austenite

Most of the austenite in a eutectoid steel will transform to martensite during the quenching to room temperature. The untransformed part is called retained austenite. Figure 1.17 shows how the amount of retained austenite in unalloyed steel varies with the carbon content. If the temperature is lowered below room temperature the transformation to martensite continues (see Figure 1.16). This method of increasing the amount of martensite is called sub-zero treatment (see Chapter 5 for details of this treatment).

In the process called martempering (see Chapter 5) the cooling is interrupted just above Ms, or in some cases just below it, and then the steel is allowed to cool to room temperature. This interruption in the cooling stabilizes the austenite somewhat, which causes the martensite formation to start at a lower temperature, thereby resulting in a higher proportion of retained austenite at room temperature. Hence a martempered or air-hardened steel has in

![Figure 1.16. Influence of the austenitic carbon content on the Ms and Mf temperatures in unalloyed steel (after Marder and Krauss)](image)
20 FUNDAMENTAL METALLOGRAPHIC CONCEPTS

general a larger amount of retained austenite than an oil-hardened one. The reason for the stabilizing effect is assumed to be due to the dissolution at the arrest temperature of the martensite nuclei formed during cooling from the austenitizing temperature. The dependence of the amount of retained austenite on the alloy content and hardening temperature will be discussed in several of the following chapters.

1.3.7 TTT diagrams

During the past twenty years there have been extensive investigations into the formation of the above-mentioned structural constituents and at present practically all grades of steel have been charted by means of TTT diagrams (Time–Temperature–Transformation). They are also called isothermal transformation diagrams and are constructed from data obtained from tests on steels transformed at series of constant temperatures, as described in the text above. Another type of diagram (CCT diagram) describes the transformation approximating more closely to that taking place during actual hardening processes. Figures 1.18a and b show that the start of the transformation takes longer for continuous cooling than for isothermal treatment. TTT diagrams play an immensely important part in the heat treatment of steel. This is the first encounter with them in this book.

1.4 DECOMPOSITION OF MARTENSITE AND RETAINED AUSTENITE ON TEMPERING

The martensite formed during hardening is generally too brittle for the steel to be put to practical use without first tempering it. Tempering usually results in an increase in toughness and a simultaneous reduction in hardness. For a better understanding of the mechanism of the tempering process the three stages that a hardened carbon steel passes through when subjected to a continuous rise in temperature are now briefly discussed.

1. 80–160°C. Precipitation of a carbon-rich phase called e-carbide. As a consequence the carbon in the martensite is reduced to approximately 0.3%.
2. 230–280°C. Decomposition of retained austenite to a bainite-like product.
3a. 160–400°C. Formation and growth of cementite (Fe₃C) at the expense of the e-carbide.
3b. 400–700°C. Continued growth and spheroidization of the cementite.

Somewhat different temperature ranges have also been reported and these differences are thought to be mainly to different rates of heating. Figure 1.19 shows a tempering curve for a 1% carbon steel with the different tempering stages marked on it.

For high-alloy chromium steels, hot-work steels and high-speed steels the range of decomposition of retained austenite is displaced towards higher
22. FUNDAMENTAL METALLOGRAPHIC CONCEPTS

Temperatures. The product of decomposition, i.e. bainite or martensite, depends on the tempering temperature and time (Figure 1.20). Bainite formation occurs isothermally, i.e. at constant temperature during the tempering process, whereas martensite forms as the steel is cooling from the tempering temperature.

In high-alloy steels a precipitation of finely dispersed complex carbides occurs at about 500°C. This is called the fourth stage of tempering. The martensite formed from the retained austenite and the, possibly coherent (see p. 34), carbide precipitate together create a hardness peak which is a characteristic feature of high-speed and other high-alloy steels. Figure 1.21

![Figure 1.21: Tempering curves for some current tool steels](image)

shows the variations in hardness of some current tool steels that have been tempered for 2 h at different temperatures.

The interconnection between hardness, temperature and time of tempering may be mathematically illustrated by means of the following expression, formulated by Hollomon and Jaffe:

\[ P = T(k + \log t) \]

where \( P \) = a parameter linked to the tempering process
\( T \) = temperature in °K (absolute temperature °K = °C + 273)
\( k \) = a constant
\( t \) = time in hours

Each value of \( P \) corresponds to a hardness value that can be read off from a so-called master parameter curve which must first be drawn for each grade.
1.5 DIFFUSION

1.5.1 The nature of diffusion

Diffusion is the process of migration of individual atoms within materials and takes place in gases, liquids and solids, and consequently occurs in steels and other metallic systems. Diffusion is of immense importance to most heat treatment processes and therefore some appreciation of the mechanisms and laws of diffusion will facilitate the understanding of many heat treatments including carburizing and decarburizing, nitriding and annealing.

The foreign or alloy atoms that occupy sites of the same type as the host atoms in the iron lattice in a substitutional solid solution move about with the aid of empty sites called vacancies. Since steel contains a large number of vacant lattice sites there is a continuous migration of atoms via these vacancies. Carbon and nitrogen atoms, are small compared with iron atoms, and can therefore be situated at sites between the iron atoms so forming an interstitial solid solution. These small atoms can diffuse without the aid of vacancies and their rate of diffusion is much higher than that of substitutional atoms.

Mathematical studies of diffusion by Einstein have shown that the average distance of diffusion \( x \) cm is equal to \( \sqrt{2Dt} \) where \( D \) is the coefficient of diffusion in \( \text{cm}^2/\text{s} \) and \( t \) the time in seconds. This equation may be applied in practice to obtain an estimate of how far diffusion can proceed in a certain time when the atoms migrate in one direction. For such purposes the equation may be written:

\[
x = \sqrt{2Dt}
\]

If we put \( \sqrt{(2D)} = k \) we obtain

\[
x = kt
\]

This equation may be regarded as fundamental to most diffusion processes. If we take an assumed case and let \( k = 0.1 \) we obtain the graph shown in Figure 1.23. Notice the difference between the diffusion distance for example of one hour's diffusion at the beginning of the process and for the same period of time after 10 h.

1.5.2 Factors that influence the rate of diffusion

Diffusion takes place due to a concentration differential (actually an activity differential) which produces the driving force required for the diffusion process. The basic expression for the dependence of diffusion on the concentration gradient is given by Fick's First Law:

\[
J = -D \frac{\partial c}{\partial x}
\]

where \( J \) = the amount of the substance under investigation that passes in unit time through unit area in a plane normal to the \( x \)-axis (g/cm\(^2\) s)

\( D \) = diffusion coefficient (cm\(^2\)/s)

\( c \) = concentration of the diffusing substance (g/cm\(^3\))

\( x \) = a coordinate (cm)

The diffusion coefficient is very strongly influenced by temperature. On a rough estimate \( D \) is doubled for every temperature increase of twenty degrees. The value of \( D \) for different temperatures is given by:
where \( D_0 \) = frequency factor (cm\(^2\)/s)  
\( Q \) = energy of activation (cal/mol/K)  
\( T \) = temperature degree Kelvin (K)  
\( R \) = gas constant (1.987 cal/mol)

The Table 1.1 gives approximate values of \( D_0 \) and \( Q \) for some substances. Different investigators give somewhat different values.\(^a\),\(^b\)

<table>
<thead>
<tr>
<th>Diffusing element</th>
<th>Diffusing through</th>
<th>( D_0 ) cm(^2)/s</th>
<th>( Q ) cal/mol</th>
</tr>
</thead>
<tbody>
<tr>
<td>Carbon</td>
<td>α-iron</td>
<td>0.0079</td>
<td>18.100</td>
</tr>
<tr>
<td>Carbon</td>
<td>γ-iron</td>
<td>0.21</td>
<td>33.800</td>
</tr>
<tr>
<td>Nickel</td>
<td>γ-iron</td>
<td>0.5</td>
<td>66.000</td>
</tr>
<tr>
<td>Manganese</td>
<td>γ-iron</td>
<td>0.55</td>
<td>67.000</td>
</tr>
<tr>
<td>Chromium</td>
<td>α-iron</td>
<td>30.000</td>
<td>82.000</td>
</tr>
<tr>
<td>Chromium</td>
<td>γ-iron</td>
<td>18.000</td>
<td>97.000</td>
</tr>
</tbody>
</table>

1.5.3 Calculation of diffusion distance

From Fick’s Second Law

\[
\frac{\partial c}{\partial t} = D \frac{\partial^2 c}{\partial x^2}
\]

we can calculate the diffusion distance. The methods of solving various types of differential equations are obtainable in works of reference.\(^c\)

There follows some examples that should be of interest to those involved in heat treatment. They may also be re-read alongside the various processes as they are discussed later in this book.

CARBURIZATION

For the sake of simplicity it is assumed that the surface of the steel acquires a carbon concentration corresponding to the carbon potential of the carburizing medium and that the carbon concentration in the outermost layer remains constant during the whole process. The carbon content of the outer layer is usually about 0.8–1.0%. To calculate the carbon concentration figure at a certain depth below the surface after a certain carburizing time the following formula may be used:

\[
C - C_0 = \left( C_1 - C_0 \right) \left[ 1 - \text{erf} \left( \frac{x}{2 \sqrt{t (D t)}} \right) \right]
\]

where \( C \) = required carbon concentration at a depth \( x \) below the surface (see Figure 1.24).

\[
C_0 = \text{basic carbon content of the steel}
\]
\[
C_1 = \text{carbon content at surface of steel}
\]
\[
x = \text{depth below surface (cm)}
\]
\[
D = \text{diffusion coefficient (cm/s)}
\]
\[
t = \text{time (s)}
\]
\[
\text{erf} = \text{a so-called error function (see Table 1.2)}
\]

Example 1

How long will it take to carburize a 0.20% C steel at 900°C so that the carbon content will be 0.40% at a depth of 1 mm? (0.40% C is roughly the carbon content of a steel that on being hardened will have a hardness of 550 H V. This hardness figure is generally taken to define the limit of the depth of case hardening.) The carbon concentration at the surface is assumed to be 1.0%.
FUNDAMENTAL METALLOGRAPHIC CONCEPTS

Inserting the given carbon values in Equation (1.6) we obtain

\[ 0.4 - 0.2 = (10 - 0.2) \left[ 1 - \mathrm{erf} \left( \frac{x}{2 \sqrt{(D/t)}} \right) \right] \]

\[ \mathrm{erf} \left( \frac{x}{2 \sqrt{(D/t)}} \right) = 0.75 \]

\[ \frac{x}{2 \sqrt{(D/t)}} = 0.814 \quad \text{(from Table 1.2)} \]

\[ x = 1.63 \sqrt{(D/t)} \]  \hspace{1cm} (1.7)

\( D \) is calculated from Equation (1.4) and taking the appropriate values from Table 1.1.

\[ D = 0.21 \exp \left( \frac{33800}{1987 \times 1173} \right) \]

\[ D = 1.1 \times 10^{-7} \]

Since the depth \( x \) is given (0.1 cm) we can calculate \( t \) from Equation (1.7)

\[ 0.1 = 1.63 \sqrt{(1.1 \times 10^{-7} \times t)} \]

\[ t \approx 34200 \]

Converting to hours we find that the carburizing time is 9.5 h. If we apply Einstein's equation and take the 2 outside the root sign we obtain

\[ x = 1.414 \sqrt{(D/t)} \quad \text{c.f. Equation (1.7)} \]

In this case the carburizing time is 12.6 h. We see therefore that Einstein's equation could have been used to give a rough estimate of the carburizing time.

DECARBURIZATION

Provided that the graph representing decarburization has the appearance shown in Figure 1.23 it is possible, by using Equation (1.8) below, to calculate the depth at which there will be a given carbon concentration.

\[ C = C_0 \cdot \mathrm{erf} \left( \frac{x}{2 \sqrt{(D/t)}} \right) \]  \hspace{1cm} (1.8)

EXAMPLE 2

A plain carbon steel, 0.85\%, C, has been heated in air for 1 h at 900 °C and as a result the carbon content of the outermost layer has been reduced practically to zero. What depth of stock must be removed from the surface if the permissible carbon concentration in the outer layer is 0.80%? The value of \( D \) is taken from the foregoing example.

28 FUNDAMENTAL METALLOGRAPHIC CONCEPTS

From Equation (1.7) we obtain

\[ 0.80 = 0.65 \times \mathrm{erf} \left( \frac{x}{2 \sqrt{(1.1 \times 10^{-7} \times 3600)}} \right) \]

\[ 0.941 = \mathrm{erf} \left( \frac{x}{2 \sqrt{(3.96 \times 10^{-4})}} \right) \]

\[ 1.35 = \frac{x}{2 \sqrt{(3.96 \times 10^{-4})}} \]

\[ x = 0.054 \text{ cm} \]

HOMOGENIZING ANNEALING

Segregations always occur in steel to a greater or lesser extent. By means of a homogenizing annealing, which is carried out at a relatively high tem-

\[ \text{Amount of alloying element} \]

Before annealing

\[ C_0 \times \]

After annealing

Figure 1.26. Reducing concentration differential by means of homogenizing annealing
FUNDAMENTAL METALLOGRAPHIC CONCEPTS

Differences in concentration of the segregating elements can be reduced. This is illustrated schematically in Figure 1.26. Provided that the variation in the concentration of the element in question is sinusoidal it is possible to calculate data for the homogenizing annealing according to the equation:

$$C = C_0 \exp \left( -\frac{D \cdot t \cdot \pi^2}{l^2} \right)$$  (1.9)

where $C$ = concentration differential after annealing
$C_0$ = concentration differential before annealing

EXAMPLE 3

The concentration differential for manganese is found to be 0.6%. This differential is to be reduced to 0.4% by a homogenizing annealing at 1100°C. What is the holding time required if the diffusion path $l = 0.01$ cm? $D_{Mn}$ at 1100°C = $2 \times 10^{-11}$ cm²/s.

According to Equation (1.9) we obtain

$$0.4 = 0.6 \exp \left( \frac{2 \times 10^{-11} \times t \times \pi^2}{0.01^2} \right)$$
$$t = 2.06 \times 10^3 \text{ s} = 5.72 \text{ h}.$$

1.6 DISLOCATIONS

In a perfect iron crystal the atoms are arranged in a regular pattern, which we have previously discussed. In order to produce shear in a perfect crystal all the atoms in a slip plane must move simultaneously. Figure 1.27 shows a section through the slip plane, illustrating how this slip can take place. A nearly perfect crystal of iron (whisker) has a strength more than 100 times the strength of ordinary mild steel. In general, however, the lattice is far from perfect. This is due to the fact that a crystal of an ordinary steel contains imperfections of various kinds. We have mentioned earlier the presence of vacancies, i.e., empty positions in the lattice by means of which atoms can move about. The crystals also contain linear faults termed dislocations. The symbol L in Figure 1.28 marks the end of what is called an edge dislocation which extends further into the lattice, i.e., at right angles to the plane of the paper.

The concept of dislocations helps us to understand and explain many characteristic features of steel, for instance why it is not so strong as it would be if it consisted of a perfect crystal. The following example illustrates this point.

If shear forces are applied as shown in Figure 1.26a the distance between atoms $E3$ and $E4$ will increase and the cohesive force between them will decrease. As the shear force approaches a critical value atom $E4$ will move into the sphere of influence of atom $D3$ as seen in Figure 1.26b where $E4$ has now jumped into position $D4$. By still applying the shear force we find that the atoms will jump across, one after the other, into new positions and the lattice will assume the configuration shown in Figure 1.26c. As a consequence of this intermittent movement of the atoms and compared with the above-described case of the perfect crystal a much smaller shear force is now required. Hence it is apparent that dislocations facilitate the plastic deformation of the material. But they have other functions.

During plastic deformation new dislocations are created and as a result it becomes gradually more difficult for the gliding dislocations to move through the crystal grains, i.e., the steel work hardens. In a material with precipitate particles, which are too hard to be sheared, work hardening is particularly marked and depends on the rings around particles left by the gliding dislocations, which have passed through a row of particles as indicated in Figure 1.29. The actual existence of dislocations has been verified with the aid of the electron microscope. Even though it is not possible to see the dislocations themselves the structural anomalies they give rise to are very obvious (Figure 1.30).

Dislocations markedly affect the properties of metals. The above presenta-
Figure 1.30. Dislocations piling up at a grain boundary in an 18/8 steel.

Dislocation formation is merely an introduction and the interested reader will find a wealth of pertinent literature 10, 11, 12.

1.7 GRAIN SIZE

When a tool or machine component has failed it is customary to examine the fractured surface in order to see whether it is fine or coarse grained. If the latter, faulty heat treatment may have been the cause (generally too high a hardening temperature) or the steel may have been unsuited to the treatment in question. A great number of the properties of steel are influenced by grain size, such as machinability, impact strength, hardenability, etc.

1.7.1 Grain boundaries

Steel contains many grains separated by grain boundaries, of which there are three distinguishable types.

LOW-ANGLE GRAIN BOUNDARIES

Such grain boundaries may be regarded as a series of dislocations lying in the same plane in an otherwise perfect crystal. The crystal lattices of the two adjacent grains give rise to an angle, the size of which is proportional to the dislocation density at the surface boundary (Figure 1.31).
34 FUNDAMENTAL METALLOGRAPHIC CONCEPTS

Another term used is phase boundary, i.e. the boundary between two grains of different phases or crystal structures. Such boundaries are comparable to high-angle grain boundaries. It can sometimes happen that two crystal structures have specific similarities along certain planes, with a regular and mutual alignment as a result. Such crystal lattices are said to be coherent. Since the grain boundaries contain considerably more vacancies and dislocations than the grains themselves, diffusion proceeds much more easily along the boundaries than through the grains. This is particularly the case at low temperatures. In addition preferential concentration of certain foreign atoms takes place at grain boundaries. Likewise, precipitation of particles (e.g. carbides in steel) from a super saturated solution often occurs at grain boundaries, thus making these easy to observe in a metallurgical microscope.

1.7.2 Methods of determining grain size

When determining the grain size of hardened steel the austenitic grain size is implied, i.e. the size of the austenitic grains before the steel is cooled and before the austenite has transformed to other structural constituents. The original austenite grains can generally be revealed after cooling the steel to room temperature and their size may be determined by different methods, four of the most commonly accepted ones being described below. It is important that the designation of the method used is stated in connection with the grain size specifications.

<table>
<thead>
<tr>
<th>ASTM Standard Classified Grain Size</th>
</tr>
</thead>
<tbody>
<tr>
<td>The grain size is determined in a microscope at a magnification of 100 ×. The ASTM Grain Size Number corresponds to a certain number of grains/\text{in}^2 according to Table 1.3.</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th></th>
<th>0</th>
<th>1</th>
<th>2</th>
<th>3</th>
<th>4</th>
<th>5</th>
<th>6</th>
<th>7</th>
<th>8</th>
<th>9</th>
<th>10</th>
</tr>
</thead>
<tbody>
<tr>
<td>ASTM Grains/\text{in}^2</td>
<td>0.5</td>
<td>1</td>
<td>2</td>
<td>4</td>
<td>8</td>
<td>16</td>
<td>32</td>
<td>64</td>
<td>128</td>
<td>256</td>
<td>512</td>
</tr>
</tbody>
</table>

The relationship between the Grain Size Number and the number of grains/\text{in}^2 is given by the expression:

\[ n = 2^n - 1 \]  \hspace{1cm} (1.10)

where \( N \) = ASTM Grain Size Number, \( n \) = number of grains/\text{in}^2 at a magnification of 100 ×.

The evaluation of the ASTM Grain Size Number is carried out with the aid of standard grain size charts or with special eyepieces (see Figure 1.33).
2 ISO-INDEX G

The number of grains, viewed at a certain magnification, is counted inside an area of 5000 mm$^2$, corresponding to a circle of 79.8 mm diameter, on the ground-glass screen of the microscope or on a micrograph. The magnification is chosen so that there are at least 50 grains within this area. The number of grains $n$ considered to be contained in this area is deduced from

$$n = n_1 + n_2$$

where $n_1$ is the number of grains wholly inside the circle and $n_2$ is half the number of grains intercepted by the circle.

36 FUNDAMENTAL METALLOGRAPHIC CONCEPTS

From the value of $n$ we calculate $m$, the number of grains per mm$^2$ from

$$m = 2n(g/100)^2$$

where $g$ is the linear magnification.

The ISO-index $G$ is obtained from the expression

$$G = \frac{\log m}{\log 2} - 3 \quad (1.11)$$

According to ASTM E 112–63 the correlation between $G$ and ASTM is

$$G = ASTM - 0.046 \quad (1.12)$$

Since only whole or half units are recorded we can, for all practical purposes, put $G$ equal to ASTM. Hence ISO-index $G$ can be evaluated from the charts or eyepieces that apply to the ASTM Grain Size Number.

3 INTERCEPT OR SNYDER–GRAFF METHOD

In this method, which is applied mainly to high-speed steels, a count is taken at a magnification of 1000× of the number of grains intercepted by a line 0.005 in length. This number is a direct indication of the Intercept Value $I$. The estimation can be carried out either on a ground-glass screen or by means of a graduated eyepiece. Ten counts are normally made, the average being taken as the Intercept Value. If we express the mean grain size on the Intercept Scale $I$ in the same way as the ASTM Grain Size Number, i.e. as the number of grains $n$/in$^2$ on the micrograph at 100× magnification, we obtain:

$$n = (2.1)^2 \quad (1.13)$$

Solving for $N$ in Equation (1.10) we obtain:

$$N = 2 \frac{\log I}{\log 2} + 3 \quad (1.14)$$

This correlation is illustrated graphically in Figure 1.34. Note that 9 5 ASTM is equal to 9.5 I.

4 FRACTURE TEST

For this test we use as a scale a series of hardened and fractured test-pieces, numbered from 1 to 10 (No. 1 being the coarsest and No. 10 the finest fracture). There are two such series, viz. Shepherd's series and Jernkontorets series (Figure 1.35). For most practical purposes these two scales can be regarded as identical. It has been shown that when a hardened steel is fractured the fracture surface usually follows the austenitic grain boundaries that existed prior to hardening. Hence a study of this surface should give an indication of the austenitic grain size. The correlation between the ASTM
Grain Size Number and Jernkontoret's fracture number (JK) or Shepherd's fracture number is shown in Figure 1.36.

**CARBIDE PARTICLE SIZE**

By means of linear analysis in the metallurgical microscope it is possible to estimate the carbide volume fraction. By counting the number of carbide grains in a section it is then possible to compute the mean carbide particle size (in $\mu m^3$).

### 1.7.3 Examples of grain size determinations

**CASE-HARDENING STEELS**

The grain size is usually determined by the McQuaid–Ehn method. The steel is carburized at 925°C for 8 h followed by cooling while it is still packed in the carburizing medium. This test gives an indication of the susceptibility of the steel to grain growth during carburization. As the steel slowly cools after the relatively long carburizing period, carbides precipitate at the grain boundaries whereby the estimation of the grain size is facilitated (see Figure 1.57).

When using the McQuaid–Ehn method nitrogen absorption from the carburizing medium may give a more favourable estimated value for the grain size of steel where the deoxidation has been incomplete.
HIGH-SPEED STEEL

The determination of grain size in hardened high-speed steel by means of microscopy is best carried out on the quenched and untempered sample since tempering tends to obscure the grain boundaries. However, at temperatures above 600°C the grain boundaries become sharp once more (see Figure 1.38). In this instance the sample has been etched in 5% nitric acid. More distinct contours would result if a grain-boundary etchant containing picric acid had been used.

CONSTRUCTIONAL STEEL AND TOOL STEEL

The grain boundaries of quenched and tempered Cr-Ni or Cr-Ni-Mo grades of steel may be difficult to discern. By tempering at 525°C for 3 h followed by slow cooling it is possible to bring out the grain boundaries more distinctly. It is then necessary to use a grain-boundary etch (see Figure 1.39). A saturated solution of picric acid in distilled water with an addition of 1% of a wetting agent is usual.

When assessing the quality level of a steel or its heat treatment the grain-size rating is very useful. Fine-grain case-hardening steels usually have a stipulated grain-size rating of 5–8 ASTM according to the McQuaid–Ehn test. The Bofors fine-grain case-hardening steel has a standard grain size of 8 ASTM. Quenched and tempered tool steel should have a grain size not coarser than 7 ASTM which is about the same as the J K fracture number 7. A high-speed steel quenched from its correct hardening temperature shows in normal cases a grain size of 9–10 ASTM or J K. Modern molybdenum-alloyed high-speed steels may show ratings higher than 10 J K. Since the
2
Materials testing

The properties conferred on steel as a result of heat treatment need to be tested and verified. There are many testing methods available but only those that are standardized and widely used are considered here. (Before reading this chapter further consideration of the section 'International Designations and Symbols' on p. vii is recommended.)

2.1 THE HARDNESS TEST

After heat treatment a steel component is usually hardness tested, and the value obtained is a good indication of the effectiveness of the treatment. The hardness test is carried out by pressing a ball or point with a predetermined force into the surface of the specimen. The hardness figure is a function of the size of the indentation for the Brinell (HB) and Vickers (HV) tests and of the depth of the penetration for the Rockwell (HRC) test. These three methods are the most commonly used tests and each has its special range of application and between them they cover almost the whole of the hardness field that is of interest to the steel producer and user.

2.1.1 The Brinell test

In the Brinell test a ball of hardened steel or sintered carbide is pressed into the surface of the specimen to be tested (Figure 2.1). Depending on the material to be tested and the ball diameter, a load up to 3000 kp may be used. For steel the following discrete ball diameters and loads have been standardized:

<table>
<thead>
<tr>
<th>Diameter of ball</th>
<th>mm</th>
<th>2.5</th>
<th>5</th>
<th>10</th>
</tr>
</thead>
<tbody>
<tr>
<td>Load</td>
<td>kp</td>
<td>187.5</td>
<td>750</td>
<td>3000</td>
</tr>
</tbody>
</table>

The diameter of the impression is measured and the Brinell hardness, which is the quotient of the load divided by the spherical area of the impression, is read off from a table. The unit, kgf/mm$^2$ or kp/mm$^2$, is not recorded after the hardness number.
The piece to be tested should have a thickness of at least 8 times the depth of the impression. The minimum distance between the centres of two adjacent impressions or from any edge should be 4 and $2\frac{1}{2}$ times the impression diameter, respectively. The finer the surface finish the more accurate are the results but a reasonable rough-ground surface is usually acceptable. The upper and lower surfaces do not have to be absolutely parallel since the anvil of the testing machine is usually adjustable to compensate for this. The Brinell hardness test is used for materials of a hardness up to about 550 HB, i.e. unhardened or fully hardened and tempered steel. For a hardness greater than 450 HB it is advisable to use a sintered-carbide ball which may be used up to about 650 HB. For plain and low-alloy carbon steels the Brinell hardness number is approximately proportional to the ultimate tensile strength, UTS (UTS in kgf/mm² or kP/mm² = HB/3). The Brinell hardness test is often used to supplement other mechanical tests since it is much quicker and cheaper than the tensile test and may be considered as a non-destructive test.

2.1.2 The Vickers test

For the Vickers test a pyramid-shaped diamond indenter is used, the apex angle being 136°. Normally the indentation load varies between 1 kp and 30 kp, but both higher and lower loads may be used. The average length of the diagonals of the impression is measured and the Vickers Pyramid Number is read off from a table. Like the HB the HV is measured as the quotient of the load divided by the pyramidal area of the impression (kP/mm²).

For steel, the thickness of the specimen or the layer to be measured must be at least 1.2 times the impression diagonal. The minimum distance between the centres of adjacent impressions or between the centre and any edge must be $2\frac{1}{2}$ times the impression diagonal. The specimen is firmly clamped horizontally, usually in a vice placed on the anvil of the machine. The Vickers hardness test is used for both hard and soft materials. Due to the small depth of penetration this test is particularly well suited to testing the superficial hardness of a material. When measuring the hardening depth of case-hardened, nitrided, induction of flame hardened steel, the ISO stipulates that the applied load shall be 1 kP, on a section cut at right angles to the surface of the steel.

The magnitude of the load used (in kP) is recorded after the hardness number, e.g. 850 HV 10. By using different loads it is possible to determine whether a steel surface has been unintentionally carburized or decarburized during the course of the heat treatment. If a low-load hardness is substantially lower than a high-load hardness—in the latter case the indenter penetrates more deeply into the material—decarburization might be suspected. The Vickers test is also suitable for hardness testing on a taper-ground section—which is used when hardness variations across a superficial zone or layer need to be measured. Vickers microhardness testing machines capable of applying loads of only a few grams are also available.

The Vickers test can be used for both very low and very high hardenes.

For example, the hardness of pure iron is about 60 HV and the hardness of certain carbides occurring in steel and sintered carbides is about 2000 HV.

2.1.3 The Knoop test

When the hardness of very thin layers is of primary interest the Knoop hardness test is particularly suitable. Figure 2.2 shows a comparison between the Vickers and Knoop methods. Under the same load the Vickers indenter penetrates more deeply into the material of the specimen than the Knoop indenter. The same depth of penetration is obtained with, say, HV 03 and HK 071B. However, the hardness values may differ slightly from the one method to the other.

2.1.4 The Rockwell test

There are several Rockwell tests, the most usual being Rockwell C (HRC), Rockwell N Superficial (usually HR 30N) and Rockwell B (HRB). Of all hardness tests based on the measurement of the penetration depth or size, the Rockwell test is the quickest since a direct reading is obtained on a dial indicator. The accuracy is relatively high, viz. ± 1 unit.

At the start of the test a pre-load or minor load ($F_1$) is applied to the indenter. The dial indicator is set at zero and the major load is applied, the time at full load being 5–10 s. After removing the major load only—the minor
load still acting—the hardness is read on the dial direct. The hardness is a function of the increase in penetration depth caused by the application and subsequent removal of the major load (see Figure 2.5).

The accuracy of the testing machine should be verified with test-pieces of known hardness before each testing operation. For this purpose standard hardness blocks are recommended. The hardness of these blocks should lie as near as possible to that of the workpiece being tested. If the testing machine has not been in continuous use, consistent readings are obtained only after the third and subsequent indentations.

Since Rockwell C (HR C) is the most frequently used Rockwell test it will be treated in more detail. The description covering this test is also valid, where applicable, for the other Rockwell tests.

For the Rockwell C (HR C) test the Brinell indenter is used, i.e. a diamond cone with an apex angle of 120°, the apex being somewhat rounded (r = 0.2 mm). This indenter is pressed into the surface to be tested with a minor load of 10 kp, and the dial indicator is set at zero. After application of the major load of 150 kp and its subsequent removal the hardness is read off direct on the dial, the minor load still being applied.

A difference in hardness of 1 HR C corresponds to a difference of 0.002 mm in penetration depth. This calls for a high degree of precision when carrying out the test and therefore test surfaces should be ground to a 280 mesh finish, the opposite surface should be smooth and parallel. The anvil and specimen should be wiped clean before testing. At least three indentations should be made, the first one being disregarded since the possible presence of dirt between the specimen and the anvil can influence the hardness reading. After the first indentation these dirt particles are compressed and will have little influence on subsequent readings. Rockwell C (HR C) impressions should not be closer than 3 mm and at least 3 mm from an edge.

The Rockwell impression is relatively deep. If carried out on thin parts incorrect readings will result; therefore the thickness of the sample should be not less than 10 times the penetration depth. With a hardness of, say, 20 HR C the specimen must be at least 1.6 mm thick, with 70 HR C at least 0.6 mm.

Round, small-diameter parts give low values if hardness tested on the cylindrical surface. This error may be corrected with the aid of Table 2.1. Rockwell testing should be avoided on slanting surfaces, partly because misleading results are obtained and partly because the machine may be damaged by the resulting non-axial or off-centre loading. It is advisable to compensate for the slant by inserting cuneiform gauge blocks between the specimen and the anvil.

When case-hardened, nitrided or thin layers generally are being hardness tested, the softer underlying core will be deformed and the values obtained
will not represent the true hardness of the superficial layer. The hard case must be at least 0.5 mm thick if a true Rockwell C hardness figure is to be obtained. For thinner layers it is necessary to employ a test method that uses a smaller load, e.g. HR 30N or HV with 1–30 kp. The Rockwell C test may, however, be applied to case-hardened parts having a case less than 0.5 mm thick. The hardness measured will be lower than the true hardness but the test may be used on mass-produced parts once the relevant hardness limits have been determined, since the values obtained will enable parts with too shallow or too large case depths to be picked out.

The Rockwell C test is normally used on steels in the hardness range 30 to 70 HR C since softer materials give unreliable values.

The Rockwell N test employs the same diamond cone indenter as Rockwell C. The minor load is 3 kp and the total major load is 15, 30 or 45 kp, HR 15N, HR 30N or HR 45N respectively. The test is used for thin sections or for case-hardened or nitrided steel where the superficial layer is less than 0.5 mm in thickness. HR 30N is used within the hardness range 40 to 90 HR 30N.

The Rockwell B test is carried out in principle in the same way as the other Rockwell tests. The indenter is a 3/8-in diameter steel ball. The total load is 100 kp. Rockwell B is used within the hardness range 30 to 110 HR B for unhardened and hardened and tempered steel.

Table 2.1 CORRECTION VALUES FOR CYLINDRICAL SURFACES

<table>
<thead>
<tr>
<th>HRC</th>
<th>Bending radius of tested surface, mm</th>
</tr>
</thead>
<tbody>
<tr>
<td>20</td>
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<td>2  5  65  8  95  11  125  16  19</td>
</tr>
</tbody>
</table>

2.1.5 The Scleroscope test

This is carried out by allowing a diamond-tipped hardened steel cylinder to fall freely through a tube on to the surface to be tested. On the rebound up the tube the cylinder is caught at its highest point and held there. The height of the rebound is a measure of the Shore C or Shore D hardness. This method does not give very reliable results since the height of the rebound is dependent on the mass of the tested object. It is used for grading the hardness of rolls for cold-rolling. When used in this connexion it has been found that the hardness measurements made on rolls less than 150 mm in diameter give very erratic results whereas larger-diameter rolls give relatively consistent results. Schmitz and Schüller have demonstrated the difficulties in using the scleroscope to obtain reproducible results. As seen in Figure 2.4 a scatter of some 20 Shore units is to be expected with each individual apparatus and a similar variation exists between the testers themselves (see Figure 2.5).

At the Bofors Company the figures in Table 2.3 are used to obtain an approximate correlation between HV and Shore C or Shore D. Here it should be stressed that the Vickers test is the only reliable one for measuring the hardness of rolls for cold-rolling. The Shore test is used only as a guide during the manufacture of these rolls.
2.1.6 Conversion tables for various scales of hardness

All hardness tests are not applicable in all circumstances and conversion tables are very necessary. Such tables have been prepared by institutions such as Jernkontoret, ASTM and EURONORM. There are some relatively minor differences between these tables but until the publication of a new comparative set of hardness scales Boles will continue to use the conversion Tables 2.2 and 2.3.

![Figure 2.3. Compilation of calibration curves and test results for separate Shore hardness testing apparatus and their comparison with Rockwell C test after Schmidt and Stekelenburg](image)

Table 2.2 CONVERSION TABLE FOR HARDNESS—ULTIMATE TENSILE STRENGTH OF STEEL

<table>
<thead>
<tr>
<th>Hardness (HV)</th>
<th>Hardness (HRB)</th>
<th>Hardness (HB)</th>
<th>Tensile strength (kp/mm²)</th>
<th>Tensile strength (N/mm²)</th>
<th>Hardness (HV)</th>
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Table 2.3 CONVERSION OF DIAMOND PYRAMID HARDNESS (HV) TO ROCKWELL C (HR C) AND DIFFERENT SHORE VALUES

<table>
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<th>Rockwell C (HR C)</th>
<th>Diamond Pyramid Hardness (HV)</th>
<th>Roll calibration</th>
<th>Shore C</th>
<th>Standard calibration</th>
<th>Shore D</th>
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</table>
2.2 THE TENSILE TEST

The maximum design stresses that a material may be subjected to on static loading are determined by means of the tensile test. The following properties of the material are usually determined by this test:

<table>
<thead>
<tr>
<th>SI designations</th>
<th>Old designations</th>
</tr>
</thead>
<tbody>
<tr>
<td>Yield point</td>
<td>$R_{p0.2}$ N/mm²</td>
</tr>
<tr>
<td>$0.2%$ offset</td>
<td>proof stress</td>
</tr>
<tr>
<td>Lower yield point</td>
<td>$R_y$ N/mm²</td>
</tr>
<tr>
<td>Upper yield point</td>
<td>$R_u$ N/mm²</td>
</tr>
<tr>
<td>Tensile strength</td>
<td>$R_m$ N/mm²</td>
</tr>
<tr>
<td>Elongation</td>
<td>$A_s$ %</td>
</tr>
<tr>
<td>Reduction of area</td>
<td>$Z$ %</td>
</tr>
</tbody>
</table>

In addition, Young's Modulus of Elasticity $E$, N/mm² (kN/m²) may be determined.

Figure 2.6 shows a test bar for the tensile testing of steel. The bar is gripped in a tensile testing machine and subjected to a successively increasing load until fracture occurs, the extension of the test bar can be traced on the diagrams in Figures 2.7a and b.

The Yield Stress is the stress at which a relatively large extension of the test bar takes place while it is being subjected to a constant load (Figure 2.7a). For steels having a clearly defined yield stress range, such as unalloyed steels, both yield higher stress $R_{p0.2}$ and lower yield stress $R_y$ are determined. In the case of alloy or hardened steels this point is not clearly defined or it may be completely absent (Figure 2.7b). For such steels the yield stress is taken to mean the Proof Stress, i.e. the stress that, after the load has been removed, produces a permanent elongation of, say $0.2\%$, designated $R_{p0.2}$ ($σ_{p0.2}$). If a structural part has been stressed beyond its yield point the material suffers plastic deformation. Hence the value of the yield stress is taken as the basis for more accurate calculations of mechanical strength.

The Ultimate Tensile Strength or just Tensile Strength $R_m$ ($σ_m$) is the maximal numerical value of the stress which, computed from the original area of the test bar, causes fracture of the specimen.

* In old ISO documents cited in this book the proof stress is designated by $R_p$. The designation $R_{p0.2}$ was adopted in 1973.

+ The designations within brackets were used before the SI was introduced.
The Elongation \( A_\text{d} \) is the percentage increase in length of the test bar after completion of the tensile test. It is measured on a gauge length that is generally 5 times the diameter of the test bar and is then designated \( A_{5d} \). On a gauge length of 10 times the diameter, the elongation is designated \( A_{10d} \), etc. The amount of elongation is a measure of the ductility of the material and, taken along with the figure for the reduction of area, it gives an indication of the amount of plastic deformation to which the material can be subjected. The elongation is calculated from:

\[
A_{\text{d}} = \frac{L_1 - L_0}{L_0} \times 100 \quad \text{(see Figure 2.6)}
\]

The Reduction of Area \( Z \) is the percentage area reduction at the point of fracture. It is calculated from:

\[
Z = \frac{A_0 - A_1}{A_0} \times 100
\]

The Modulus of Elasticity (Young’s Modulus), \( E \), is obtained by dividing the stress (at any point below and up to the elastic limit) by the corresponding strain. The E-modulus is roughly the same for all steels, viz. about 206,000 N/mm² (21,000 kp/mm²) and is a measure of the slope of the straight-line part of the diagram. The E-modulus may also be computed from data obtained by measuring the resonance frequency of a test bar.

The E-modulus is an indication of the stiffness of the material within the elastic range. There is a widespread belief that hard steels are stiffer than soft ones, even if the load is insufficient to cause a permanent set in the material. However, this is a misconception, since the E-modulus of different steels subjected to different heat treatments usually agree to within 2 - 3%.

If a structural member or machine part is to be stiff and strong it must be adequately dimensioned to ensure sufficient rigidity. Thus the possibility of influencing the rigidity by a judicious choice of steel is very small. On the other hand it is quite obvious that a hard material can be subjected to heavier loads before a permanent set occurs.

### 2.2.1 Comparison between mechanical properties obtained according to different specifications

Tensile and yield strengths obtained according to different specifications are often directly comparable. Only a recalculation of the units is required. Values of the elongation as obtained according to various standard specifications or measured on different gauge lengths are not directly commensurable after recalculation (see Table 2.4).

### 2.3 THE IMPACT TEST

The impact test gives an indication of the amount of energy absorbed by the material at fracture. The test consists in allowing a tup on a rigid pendulum arm to swing from a fixed height and strike a standardized specimen set at the lowest point of the swing (Figure 2.8). The angle of displacement of the pendulum after fracturing the specimen is measured and from this is calculated the energy required to fracture the specimen.

The prevalent European specimens for impact testing incorporate a U or V notch and are designated Charpy U or Charpy V respectively (see Figure 2.9). The Charpy U value is recorded as kpdm² or joule while
the Charpy V value is kpm or joule. In Germany the DVM specimen is the current one, in France it is the Mennager specimen and in English-speaking countries the Izod specimen prevails. The dimensions of the various specimens are shown in Table 2.5. In the Izod test the V-notched specimen has one end tightly clamped in the vice of the testing machine and the tup strikes the free end.

Many grades of steel, particularly general structural steels, show high notch impact values at room temperature but very low values at lower temperatures. The transition from high to low notch impact values takes place within a rather narrow temperature range. The lower transition temperature is the one usually quoted. The figure 270 (2.8 kpm) is a frequently occurring minimum value of the Charpy V notch impact toughness. Under some circumstances the upper transition temperature is also quoted.

The V-notched specimen gives a sharply defined transition range whereas the U-notched one usually gives successively falling impact values with decreasing temperatures (see Figure 2.10). Whereas the values of ultimate tensile strength and yield point may be applied direct to expressions dealing with strength calculations, impact values can only be used for judging, from
### Table 2.6 IMPACT TESTING

<table>
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<tr>
<th>Charpy U</th>
<th>Charpy V</th>
<th>DVM</th>
<th>V notched</th>
<th>Charpy</th>
<th>J</th>
<th>Load</th>
<th>ft-lb</th>
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</thead>
<tbody>
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<td>(KU) J</td>
<td>(kpm/cm²)</td>
<td>(kpm)</td>
<td>(kpm)</td>
<td></td>
<td>(ft-lb)</td>
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</tbody>
</table>

Experience, the suitability (or otherwise) of a material for a specific purpose. There are no exact conversion factors for the impact values obtained by the various methods. Published conversion tables only give an indication of the mutual relationship between the various values. This means that the impact values can be guaranteed valid only for the test method actually used (cf. Table 2.6).

### 2.4 THE TORSION IMPACT TEST

By means of the Carpenter Torsion Impact machine the torsion impact may be determined on specimens illustrated in Figure 2.11. To conduct the test the specimen is clamped to a flywheel which rotates at a certain speed. The other end of the specimen is suddenly gripped and its rotation instantaneously stopped, and the specimen fractures as a result of the torsional impact blow. The energy required to break it is recorded in ft-lb. The diagram in Figure 2.12 has been constructed from the results obtained from hardened and tempered grade O1 steel. This test is very useful in selecting a suitable steel and its heat treatment for components subjected to simultaneous impact and torsion. The test can also be used to give an indication of the tempering temperature range for the maximum impact resistance of the steel. However, the method is not recommended when a choice is to be made among steels differing greatly in composition and characteristics.

### 2.5 THE FATIGUE TEST

#### 2.5.1 Fatigue in general

Machine component failures are in 90% of all cases due to fatigue. By fatigue is meant crack formation caused by fluctuating loads, the maximum stress always being less than the ultimate tensile strength of the material. The fatigue crack usually starts at some stress raiser in the material such as a slag inclusion or an excessively sharp fillet radius. At first the crack propagates slowly, but its speed of propagation increases successively.

![Figure 2.11. Test specimen for the Carpenter torsion test impact](image-url)

![Figure 2.12. Torsion impact and hardness as a function of the tempering temperature. Steel O1 (Refines RT 1733)](image-url)
In many service failures the fracture surface exhibits markings resembling the annual growth rings in a tree trunk. These are usually referred to in the literature as ‘clam shell’ markings or arrest lines. They have been attributed to periods of crack extension followed by crack arrest as a consequence of variation in the loading pattern of the component. These arrest lines indicate the initiation area and growth direction of the fatigue crack. After some time the crack has grown so much that the stress in the remaining part of the section exceeds the ultimate tensile strength of the material and the final fracture occurs. This usually takes place in a brittle manner and with little or no deformation. Figure 2.13 shows a typical fatigue fracture.

A study of a fatigue fracture on a microscopic level, i.e. in an electron microscope reveals the stepwise propagation of the fatigue crack with one step per loading cycle. This stepwise propagation gives rise to a fine line pattern in the fracture surface known as ‘fatigue striations’ (see Figure 2.14).

2.5.2 Test procedure

In a fatigue test the fatigue stress may be regarded as resolved into a constant static and a fluctuating stress which varies with time. Hence the applied stress varies between an upper and a lower limit, $\sigma_{\text{max}}$ and $\sigma_{\text{min}}$ respectively. The constant stress is called the mean fatigue stress ($\sigma_m$) and is defined as:

$$\sigma_m = \frac{\sigma_{\text{max}} + \sigma_{\text{min}}}{2}$$

The fluctuating stress may be defined by its amplitude $\sigma_a$, Figure 2.15, as follows:

$$\sigma_a = \frac{\sigma_{\text{max}} - \sigma_{\text{min}}}{2}$$

The applied fatigue stress may be axial (tensile or compressive), flexural, shear or torsion stress. If $\sigma_{\text{max}}$ and $\sigma_{\text{min}}$ have opposite signs during each load cycle they are referred to as alternating stresses; if the same sign, pulsating stresses (see Figure 2.16). In addition there is a special type of fatigue test, the rotating-cantilever test. The specimen is loaded so that it is deflected as it rotates, Figure 2.17. The bending load rotates round the test bar, as it were, giving rise to alternating bending stresses in it. The results of fatigue tests

![Figure 2.13. Illustration of fatigue fracture, with arrest lines and final structure](image)

![Figure 2.14. Fatigue fracture showing striations; 8700 x](image)

![Figure 2.15. Mean fatigue stress $\sigma_m$ and stress amplitude $\sigma_a$](image)
with alternating bending stresses and reversed flexural stresses respectively, are numerically equivalent. The results of fatigue tests are usually plotted in Wöhler diagrams. Figure 2.18. In these the amplitude of the stress $\sigma$ is plotted as a function of the number $N$ of load cycles to fracture; the latter scale usually being logarithmic. For amplitudes less than the endurance limit, $\sigma_0$, the specimen will not fracture.

### 2.5.3 Different types of fatigue fractures

By studying the appearance of the fractured surface of a fatigue failure, it is possible to determine the extent of overloading as well as the relative degree of stress concentration, see Figure 2.19. If there are no stress concentrations, the fatigue zones are convex-shaped, viewed from the point of fracture.

<table>
<thead>
<tr>
<th>Stress category</th>
<th>No stress concentration</th>
<th>Low stress concentration</th>
<th>High stress concentration</th>
</tr>
</thead>
<tbody>
<tr>
<td>Light overloading</td>
<td>Heavy overloading</td>
<td>Light overloading</td>
<td>Heavy overloading</td>
</tr>
<tr>
<td>a</td>
<td>b</td>
<td>c</td>
<td>d</td>
</tr>
<tr>
<td>e</td>
<td>f</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Figure 2.19. Illustrations of different fatigue fractures caused by various types of flexural fatigue
initiation. As the stress concentrations increase there is a gradual and continuous change to concave-shaped zones. The size of the final fracture area increases with the load. In contrast to pulsating flexural stresses, alternating and rotating flexural stresses can initiate fatigue fracture from two sides as well as from one, depending on the size of the stress raisers.

![Figure 2.22. Goodman diagram](image)

A case illustrating fracture type 2e (Figure 2.19) is shown in Figure 2.20, where the fracture occurred in a loading hook, the rather poor surface finish of which gave rise to high stress concentrations. Figure 2.21 illustrates a fracture of type 3e, i.e., caused by rotating flexural stresses with high stress concentrations and a small amount of overloading.

### 2.5.4 Goodman diagram

The endurance limit $\sigma_n$ for different mean fatigue stresses $\sigma_m$ can be shown in a fatigue diagram or Goodman diagram, Figure 2.22. The abscissa is graduated in $\sigma_m$ units and the ordinate in $(\sigma_n - \sigma_m)$ units. The diagram is constructed by plotting values of $\sigma_m$ against the corresponding values of $\sigma_n$ on either side of the 45° reference line. In this way two curves are obtained as functions of $\sigma_n$: one for $(\sigma_n - \sigma_m)$ and the other for $(\sigma_n - \sigma_m)$. These two curves enclose an area. A structural member subjected to stresses, the values of which are confined inside this area, is exposed to only a slight risk of failure by fatigue. The endurance limit $\sigma_n$ decreases as the mean fatigue stress $\sigma_m$ increases.

Usually it is not expected that there should be any deformation of the material during the fatigue test run and therefore the maximum allowable
stress \( \sigma_{\text{max}} \) has its upper limit defined by the yield stress in tension \( \sigma_y \), and its lower limit by the yield stress in compression \( \sigma_c \). Hence the diagram is truncated at both ends as shown in Figure 2.22.

The difference between the negative and positive mean fatigue stresses is mostly so small that only the positive part of the diagram need be constructed. The limiting curves may often be approximated by straight lines. This implies that the diagram can be constructed simply from two separate fatigue tests.

### 2.5.5 Endurance limit—ultimate tensile strength

For the majority of hardened and tempered steels the relationship between the endurance limit \( \sigma_D \) and the ultimate tensile strength is \( \sigma_D = 0.5 \times \sigma_b \) for values of \( \sigma_b \) up to about 130 kg/mm² (1280 N/mm²), which can be seen to hold for the SAE steel 4140 (see Figure 2.23). According to this figure the critical limit increases with increasing carbon content. A steel that has high values of both endurance limit and tensile strength is grade H 13 which will be discussed further on in this book. The values obtained in a fatigue test depend on the type of test used. The approximate relationships given below between the endurance limit obtained from various fatigue tests are valid in most cases.

- Endurance limit for rotating flexural stresses \( \sigma_D = 0.5 \times \sigma_b \) (in tension–compression)
- Endurance limit for reversed flexural stresses \( \sigma_D = \sigma_b \)
- Endurance limit for alternating stresses \( \sigma_D = 0.7 \times \sigma_b \)

These relationships are illustrated in the modified Goodman diagram, Figure 2.24.

### 2.5.6 Surface finish

The surface finish of a structural part subjected to fatigue stresses is of paramount importance. In general a highly polished surface gives the highest endurance limit. If the finish is not up to this standard the figure for the endurance limit obtained from a polished specimen must be reduced accordingly. For each individual case the reduction is determined by means of a surface factor \( K \). This factor is a numerical expression of the influence of the surface finish on the endurance limit.

The reduced value of the endurance limit \( \sigma_D \) is given by:

\[
\sigma_{D\text{red}} = K \times \sigma_D
\]
where $\sigma_{O,red}$ is the reduced endurance limit for the surface finish concerned. Figure 2.25 shows the surface factor $K$ as a function of the ultimate tensile strength of the material. This diagram is valid, by and large, for the majority of constructional materials with the notable exception of grey cast iron, the surface finish of which has little influence on its endurance limit.

![Figure 2.25](image)

**Figure 2.25. Influence of the surface finish on the endurance limit:** (a) Brightly polished surface; (b) ground surface; (c) rough-machined surface; (d) sharp angular fillet (= V-groove); (e) surface with skin of mill scale; (f) surface exposed to corrosion in fresh water; (g) surface exposed to corrosion in sea water.

The following example is worth studying.

A steel shaft, $\sigma_n = 70$ kp/mm$^2$ (680 N/mm$^2$), fails due to fatigue. There is a skin of mill scale on the surface of the shaft. The endurance limit for a highly polished finish is

$$\sigma_{O,red} = 0.5 \sigma_n$$

i.e.

$$\sigma_{O,red} = 0.5 \times 70 = 35 \text{ kp/mm}^2 \ (340 \text{ N/mm}^2)$$

The value of $K$ is 55%.

$$\sigma_{O,red} = 0.55 \times 35 = 19 \text{ kp/mm}^2 \ (180 \text{ N/mm}^2)$$

68 MATERIALS TESTING

In order to avoid failure by fatigue, a steel having an UTS of $\sigma_n = 110$ kp/mm$^2$ (1080 N/mm$^2$) is chosen. For this steel

$$\sigma_{O,red} = 0.5 \times 110 \times 0.04 = 22 \text{ kp/mm}^2 \ (220 \text{ N/mm}^2)$$

In this case the increase in the endurance limit is insignificant. By choosing a rough-machined surface it is seen that the first steel gives a better result:

$$\sigma_{O,red} = 0.5 \times 70 \times 0.83 = 29 \text{ kp/mm}^2 \ (280 \text{ N/mm}^2)$$

Another example of the influence of notches or stress raisers on the endurance limit is shown in Figure 2.26 which refers to grade H13 with $\sigma_n = 180$ kp/mm$^2$ (1770 N/mm$^2$).

![Figure 2.26](image)

**Figure 2.26. The influence of the stress concentration factor ($K_k$) on the endurance limit in tension-compression for steel H13 (Bolts B2P 19), heat treated to an ultimate tensile strength of 190 kp/mm$^2$ (1770 N/mm$^2$).**

This steel has been tested with different stress concentration factors ($K_k$) which are computed from special formulae and diagrams. $K_k = 10$ represents a smooth test bar. The larger the value of $K_k$, the greater is the stress concentration effect.
2.5.7 Influence of change of section

All deviations from the straight and simple test-bar shape cause local concentrations. The maximum stress may be calculated by making use of a notch sensitivity factor $\alpha$, the numerical value of which depends on the type of stressing and on the shape of the specimen. Figure 2.27 shows how $\alpha$ is estimated from given values of $D$, $d$ and $e$. $\sigma_{\text{max}}$ is then calculated from

$$\sigma_{\text{max}} = \alpha \times \sigma_{\text{nom}}$$

The above simplified passage on the influence of change of section is included solely for the purpose of drawing the reader's attention to the great importance of this subject. Detailed instructions on how to cope with calculations involving the influence of change of section are to be found in reference books for design engineers.

2.5.8 Ways of increasing the endurance limit

There are many ways of increasing the endurance limit of steel. Most of them involve the introduction of compressive stresses into the surface of the steel by means of special heat treatments. These points will be discussed later in the book under the appropriate heat treatments.

2.6 THE CREEP TEST

If a steel is stressed at elevated temperature it is subject to a continuous and progressive plastic deformation with time. This deformation is called creep.

70 MATERIALS TESTING

If creep is permitted to continue for a sufficiently long time, fracture will occur ultimately. The fracture itself starts in a characteristic manner by microscopic intercrystalline cracking, the final fracture being typically of a brittle nature.

It may be uneconomical to let a structure carry such a small load that creep can be ignored. Since the amount of creep depends on time, and provided the creep strength is known, a structure may be so designed that it will last for a specific time. The resistance to creep of a material is specified as the creep rupture strength and/or the creep strength.

THE CREEP RUPTURE STRENGTH

This is a measure of the tensile stress that will cause the steel to fracture after a specified time at a specified temperature. The time is usually specified in periods of 100, 1000, 10000 or 100000 h and the rupture stress is reported as, say, $\sigma_{10000}$ 10000/650 where $c$ signifies creep, B stands for fracture, 10000 the number of hours and 650 the temperature in °C.

THE CREEP STRENGTH

This is a measure of the tensile stress that causes the steel to show a specified plastic elongation in a specified time at a specified temperature. The plastic elongation, or strain, is usually specified in terms of strains of 0.2%, 0.5%, or 1% after 100, 1000, 10000 or 100000 h. The creep strength is reported as, say, $\sigma_{0.2}$ 1/10000/650 which indicates that the creep stress reported will produce 0.2% permanent strain in 10000 h at 650 °C.
Figure 2.29. Idealized creep curves:
1st = first creep stage (decreasing rate of creep)
2nd = second creep stage (steady-rate creep)
3rd = third creep stage (accelerating creep)

THE CREEP TEST

When carrying out the creep test, the specimen is held between two pull rods and heated to the desired temperature, whereupon the load is applied (Figure 2.28). This load is kept constant during the whole test. The extension of the specimen is plotted as a function of the time and a creep or time extension curve is obtained, the idealized shape of which is shown in

Figure 2.30. The resistance to creep of steel 12CrMoV 1/4 at 550 and 600 °C. (Creep rupture strength and creep strength for 0.5% plastic strain)

72 MATERIALS TESTING

Figure 2.29. The creep curve is characterized by the following stages:
- an initial extension
- 1st stage. Decreasing rate of creep
- 2nd stage. Steady-rate creep
- 3rd stage. Accelerating creep

The results obtained from the creep tests are usually brought together and summarized in a creep diagram. As an example, the following information can be obtained from Figure 2.30. An applied stress of 290 N/mm² (30 kp/mm²) at 550 °C produces in 5000 h a permanent strain of 0.5%.

2.7 BRITTLE AND DUCTILE FRACTURES

When a material breaks the fracture can be related to either of two fundamental mechanisms, viz. cleavage or shear (or sometimes a combination of both). On the scale of atomic dimensions it is easy to differentiate between the two main fracture types. In the case of cleavage, fracture occurs at right angles to the direction of the tensile stress. This results in a disruption of the atomic forces in the tensile direction. Fracture by shear originates in slip and proceeds under the influence of shear stresses, Figure 2.31.

Figure 2.31. Schematic sketch illustrating fractures at atomic levels

Fracture by cleavage occurs as a result of the application of a well-defined stress and in most cases without appreciable deformation. Shear fracture, on the other hand, develops gradually and is completed at considerably greater stress and strain than those operating at the instant the fracture was initiated. A cleavage fracture follows certain definite crystallographic planes in the individual grains and hence it has a crystalline appearance, which is
described as brittle. This type of fracture is called transcristalline which means that the crack is propagated right through the body of the grains.

A shear fracture, which is always preceded by a large plastic deformation, has a fibrous or ropy appearance and is termed ductile. On examination of the fractured surface in the electron microscope it is found that a cleavage fracture shows the characteristic river pattern, Figure 2.32, in which the flow lines follow the direction of crack propagation. A shear fracture shows little hollows, termed dimples, Figure 2.33.

During the course of tensile testing with a notched test specimen there arises in the immediate vicinity of the notch tip a triaxial system of stresses which creates what is called constrained plastic deformation. An increase in tensile stress, applied parallel to the axis of the test-piece, is now required in order that the shear force shall overcome the strength of the material and allow the plastic deformation to proceed. A notched tensile specimen may thus show higher values of yield strength and ultimate tensile strength than a smooth one. As a result of strain-hardening, stresses up to $3 \times \sigma_t$ may
develop at sharp notches. According to Figure 2.34 this would imply that the transition from ductile to brittle fracture takes place at a higher temperature when the specimen is notched.

If the steel is very brittle, a notched specimen will not give higher values of yield strength and ultimate tensile strength. Hence, the toughness of a steel may be judged by the results obtained from a notched specimen. If a smooth and a notched test piece gives the same results, the steel may be classified as ductile. If the value obtained from the notched specimen is about 1.5 times that of the value from the smooth specimen, the steel can be regarded as tough.

The geometry of the notch plays a decisive part in the type of fracture that the material will develop whether in impact or in tensile testing. This is utilized in the Charpy V test-piece, as has already been touched upon (see Figure 2.10). In tests conducted at temperatures below the transition

![Figure 2.35: Impact test curve for a steel with 0.18\% C and 1.5\% Mn (Before N 54) after different treatments. Approximately 60 kJ/mm² (90 K/mm²) UTS for all conditions.](image)

76 MATERIALS TESTING

temperature the steel breaks with a brittle fracture in the Charpy V test, but is ductile in the Charpy U test. The steel-making process used for the steel as well as the heat treatment has an influence on the impact value, the transition temperature and the type of fracture, which is illustrated in Figure 2.35.

The above, much simplified account of fracture phenomena is included in order to explain, if only superficially, why the same steel, used in the same construction, may sometimes fail by brittle and sometimes by ductile fracture. A brittle fracture may have as its cause an unsuitable grade of steel, a faulty design or poor finish, e.g., inferior machining or may result from unsatisfactory heat treatment. It also emphasizes the necessity of using the appropriate heat treatment.

2.8 FRACTURE TOUGHNESS

During the last decade the mechanism of fracture has been studied very closely and the newly developed methods of testing are beginning to be applied. The term fracture toughness is coming into use more and more frequently in the literature. Since the heat treatment very often has a decisive influence on the fracture toughness, mention of this material property will be found in subsequent chapters. A short note on the concept of fracture toughness is therefore called for.

The fracture toughness of a material is a measure of its capacity to withstand crack propagation from stress raisers when it is subjected to tensile stresses. In practice such stress raisers may be surface defects from a machining operation, incipient fatigue cracks, or a faulty structure derived from improper heat treatment. In calculations dealing with fracture mechanics a differentiation is made between the concepts of plane stress and plane strain.

The material property $K_c$ comes under consideration only in conditions of plane strain, which is the case in heavy-section constructions. The value of $K_c$ is independent of the form of the specimen and the geometry of the stress raiser, but does depend on the temperature. The material property $K_c$ is reckoned with in plane stress conditions which is the case in light-section constructions. $K_c$ is dependent on the dimensions of the specimen, the geometry of the stress raiser and the testing temperature.

In order to evaluate the fracture toughness, standards covering test specimens, methods of measurement and assessment of results have been set up by a committee within ASTM. In the following only the CT (Compact Tension) test bar (see Figure 2.36), will be touched upon.

A fatigue-testing machine is first used to initiate a fatigue crack in a previously machined notch and this crack is allowed to grow to a certain depth in the test bar. The final testing is carried out at the desired temperature, using slowly applied loading—tensile or bending as the case may be—until the test bar fractures (see Figure 2.37). The typical appearance of a fractured specimen is also shown in Figure 2.36. The left-hand part of the test-piece shows the machined notch which terminates in a sharp tip. Proceeding from this there is the starter crack which ends in the final ductile fracture. If a fracture test result is to be recognized and accepted it must satisfy certain
specific requirements. During the course of the test an X–Y recorder is used to register load and COD (Crack Opening Displacement). In this way a curve is obtained which resembles, as a rule, one of the three representative curves shown in Figure 2.38.

Evaluation is carried out by using the load–COD curve obtained during the test. The critical load $P_0$ in Figure 2.38 represents the load acting just at the instant when the material in front of the crack tip gives way under the applied force. By applying the formula appropriate to the type of specimen used the fracture toughness of the material can be calculated from this value. The unit of Critical Stress Intensity, $K_{IC}$, is kN/mm$^{3/2}$, Nmm$^{3/2}$ or MNm$^{-3/2}$. For conversions see Table 8.30. In American literature the unit is given as PSI x $\sqrt{\text{in}}$. The fracture toughness increases with decreasing tensile strength. There is, however, no unequivocal conversion factor that can be applied to convert ultimate tensile strength or yield strength to fracture toughness since different grades of steel at the same strength level

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**Figure 2.37. CT specimen under test with COD gauge attached**

**Figure 2.38. Representative type of load–COD diagrams.**

Type I: The load increases more or less steadily to fracture, which has a ductile appearance.

Type II: The curve shows a pop-in which corresponds approximately to the yield point.

Type III: The curve resembles that of the tensile stress–strain curve for a hardened and tempered steel. Regarded on a macroscale, the fracture has a brittle appearance.
80 MATERIALS TESTING

present different fracture toughness values. This point is exemplified in Figure 2.39.

Fracture toughness data for hard steels and white cast iron are rather scarce. This is mainly due to the difficulty of precracking brittle materials in a controlled way. It is generally accepted that during fatigue of brittle materials, unstable crack propagation will occur immediately on crack initiation. Since a fracture toughness test is the only ductility test that gives relevant data for engineering design, the Jernkontoret Research Organization considered it as interesting to make a fracture mechanics approach to hard steels and cast irons.

A fatigue method permitting stable crack propagation from a machined notch was developed and another method of precracking brittle materials requiring a considerably shorter time than fatigue precracking (10 min as compared with 8 h) was presented. The test results for some well-known tool steels and high-speed steels are shown in Figure 2.40. It is interesting to note that the fracture toughness for D 2 is higher than for A 2 and O 1. This is, however, in accordance with certain practical results obtained.

2.8.1 The Impliedation of Fracture Toughness

A possession of some knowledge of the fracture toughness of a material enables the designer to dimension the members of the construction so that catastrophic crack growth and other design faults are avoided. It is also possible to calculate the maximum acceptable size of possible defects already inadvertently present in existing constructions. For this calculation it is necessary to know not only the value of the fracture toughness but also the design loading, the yield point of the material and the size and shape of the defects in the material, such as cracks and similar inhomogeneities.

For further information on fracture toughness the reader is referred to the specialized literature, some of which is quoted in references 4–10.

REFERENCES

3. SCOTT, R., Trans. ASM, 48, 715 (1956)
3
Alloying elements in steel

It is a long-standing tradition to discuss the various alloying elements in terms of the properties they confer on steel. For example, the rule was that Chromium (Cr) makes steel hard whereas Nickel (Ni) and Manganese (Mn) make it tough. In saying this one had certain types of steel in mind and transferred the properties of a particular steel to the alloying element that was thought to have the greatest influence on the steel under consideration. This method of reasoning can give false impressions and the following examples will illustrate this point.

When we say that Cr makes steel hard and wear-resisting we probably associate this with the 2% C, 12% Cr tool steel grade, which on hardening does in fact become very hard and hard-wearing. But if, on the other hand, we choose a steel containing 0.10% C and 12% Cr, the hardness obtained on hardening is very modest. It is quite true that Mn makes for toughness in steel if we have in mind the 13% manganese steel, the so-called Hadfield steel. In concentrations between 1% and 5%, however, Mn can produce a variable effect on the properties of the steel it is alloyed with. The toughness may either increase or decrease. The effects produced by the various alloying elements on different types of steel is described in detail in books of reference1, 2, 3, 4. The present chapter will deal with only some of the basic aspects of the effects of alloying elements on steel.

3.1 SOLIDS

A property of great importance is the ability of alloying elements to promote the formation of a certain phase or to stabilize it. These elements are grouped as austenite-forming, ferrite-forming, carbide-forming and nitride-forming elements.

3.1.1 Austenite-forming elements

The elements C, Ni and Mn are the most important ones in this group.
Sufficiently large amounts of Ni or Mn render a steel austenitic even at room temperature. An example of this is the so-called Hadfield steel which contains 13% Mn, 1.2% Cr and 1% C. In this steel both the Mn and C take part in stabilizing the austenite. Another example is austenitic stainless steel containing 18% Cr and 8% Ni.

The equilibrium diagram for iron–nickel, Figure 3.1, shows how the range of stability of austenite increases with increasing Ni-content. An alloy containing 10% Ni becomes wholly austenitic if heated to 700 °C. On cooling, transformation from γ to α takes place in the temperature range 700–300 °C.

3.1.2 Ferrite-forming elements

The most important elements in this group are Cr, Si, Mo, W and Al. The range of stability of ferrite in iron–chromium alloys is shown in Figure 3.2. Fe–Cr alloys in the solid state containing more than 13% Cr are ferritic at all temperatures up to incipient melting. Another instance of a ferritic steel is one that is used as transformer sheet material. This is a low-carbon steel containing about 3% Si.

Figure 3.1. Fe–Ni equilibrium diagram (reproduced by permission from Metals Handbook, vol. 8, American Society for Metals, 1973)

Figure 3.2. Cr–Fe equilibrium diagram (reproduced by permission from Metals Handbook, vol. 8, American Society for Metals, 1973)

3.1.3 Multi-alloyed steels

The great majority of steels contain at least three components. The constitution of such steels can be deduced from what are called ternary phase diagrams. A typical example is shown in Figure 3.3. This figure is a modified S austenite diagram (from Metal Progress, Nov., 1949).
diagrams (3 components). The interpretation of these diagrams is relatively difficult and they are of limited value to people dealing with practical heat treatment since they represent equilibrium conditions only. Furthermore, since most alloys contain more than three components it is necessary to look for other ways of assessing the effect produced by the alloying elements on the structural transformations occurring during heat treatment.

One approach which is quite good, is the use of Schaeffler diagrams (see Figure 3.3). Here the austenite formers are set out along the ordinate and the ferrite formers along the abscissa. The original diagram contained only Ni and Cr but the modified diagram includes other elements and gives them coefficients that reduce them to the equivalents of Ni or Cr respectively. The diagram holds good for the rates of cooling which result from welding.

**SOME EXAMPLES TAKEN FROM THE DIAGRAM**

A 12% Cr steel containing 0.3% C is martensitic; the 0.3% C gives the steel a nickel equivalent of 9.

An 18/8 steel (18% Cr, 8% Ni) is austenitic if it contains 0.05% C and 2% Mn. The Ni content of such steels is usually kept between 9% and 10%.

Hadfield steel with 13% Mn (mentioned above) is austenitic due to its high carbon content. Should this be reduced to about 0.20% the steel becomes martensitic.

### 3.1.4 Carbide-forming elements

Several ferrite formers also function as carbide formers. The majority of carbide formers are also ferrite formers with respect to Fe. The affinity of the elements in the line below for carbon increases from left to right.

Cr, W, Mo, V, Ti, Nb, Ta, Zr.

Some carbides may be referred to as *special carbides*, i.e. non-iron-containing carbides, such as Cr₃C₂, W₂C, VC, Mo₂C. *Double or complex carbides* contain both Fe and a carbide-forming element, for example FeₓWₓCₙ.

High-speed and hot-work tool steels normally contain three types of carbides which are usually designated M₇C₃, M₈₂₅₃ and MC. The letter M represents collectively all the metal atoms. Thus M₇C₃ represents Fe₇W₇C₃ or Fe₇Mo₇C₃; M₈₂₅₃ represents Cr₂₅C₅ and MC represents VC or V₇C₅.

### 3.1.5 Carbide stabilizers

The stability of the carbides is dependent on the presence of other elements in the steel. How stable the carbides is depends on how the element is partitioned between the cementite and the matrix. The ratio of the percentage by weight, of the element contained in each of the two phases is called the partition coefficient K. The following values are given for K:

<table>
<thead>
<tr>
<th>Element</th>
<th>K</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al</td>
<td>0.2</td>
</tr>
<tr>
<td>Cu</td>
<td>0.2</td>
</tr>
<tr>
<td>P</td>
<td>0.2</td>
</tr>
<tr>
<td>Si</td>
<td>0.2</td>
</tr>
<tr>
<td>Co</td>
<td>0.2</td>
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<tr>
<td>Ni</td>
<td>0.2</td>
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<tr>
<td>W</td>
<td>0.2</td>
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<tr>
<td>Mo</td>
<td>0.2</td>
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<tr>
<td>Mn</td>
<td>0.2</td>
</tr>
<tr>
<td>Cr</td>
<td>0.2</td>
</tr>
<tr>
<td>V</td>
<td>0.2</td>
</tr>
<tr>
<td>Ti</td>
<td>0.2</td>
</tr>
<tr>
<td>Nb</td>
<td>0.2</td>
</tr>
<tr>
<td>Ta</td>
<td>0.2</td>
</tr>
</tbody>
</table>

Increasing Note that Mn, which by itself is a very weak carbide former, is a relatively potent carbide stabilizer. In practice, Cr is the alloying element most commonly used as a carbide stabilizer. Malleable cast iron (i.e. white cast iron that is rendered soft by a graphitizing heat treatment called malleabilizing) must not contain any Cr. Steel containing only Si or Ni is susceptible to graphitization, but this is most simply prevented by alloying with Cr.

### 3.1.6 Nitride-forming elements

All carbide formers are also nitride formers. Nitrogen may be introduced into the surface of the steel by nitriding. By measuring the hardness of various
ALLOYING ELEMENTS IN STEEL

3.1.7 Effect on ferrite hardness

All alloying elements that form solid solutions in ferrite affect its hardness. The amount by which they alter it is shown in Figure 3.5. Si and Mn, the most frequently occurring alloying elements, have a relatively potent effect on the hardness of ferrite, while Cr gives the smallest hardness increase. For this reason Cr is a most convenient alloying element in a steel that is to be cold worked during the course of its processing.

Figure 3.5. Effect of alloying element additions on ferrite hardness (after Bain).

3.1.8 Effect on grain growth

An alloying element that plays an important part in limiting grain growth is V, which in small amounts, around 0.10%, will prevent grain growth in steel on its being heated for hardening. The reason for this is that the vanadium is present as widely dispersed carbides and nitrides. To dissolve them, a high temperature is required and therefore at ordinary hardening temperatures the dispersed V phase acts as an inhibitor to grain growth. If the temperature is increased so much beyond normal range that the dispersed phase is dissolved, a large increase in the grain size of the steel results. Ti and Nb are also two other alloying elements whose effect is similar to that of vanadium. In high-speed steel and other high-alloy tool steels the double carbides of W and Mo also act as inhibitors to grain growth.

In the production of the so-called fine-grained steel grades, which are mostly used as case-hardening steels, the desired effect is obtained by the addition of appropriate amounts of Al to the molten steel. The normal practice is that the oxygen concentration is first reduced to a suitable level and then Al is added in amounts corresponding to the nitrogen content of the steel. As the steel cools a dispersion of Al-N particles is produced and as a result the steel is rendered resistant to grain growth at temperatures normally employed for heat treatment. The diagram in Figure 3.6 shows that a steel containing, say, 0.007% N and 0.03% Al should remain fine-grained up to about 1000°C. The critical or coarsening temperature, at which grain growth sets in, is dependent on time as well as on temperature. The diagrams in Figures 3.7 and 3.8, which apply to the case-hardening steel En 352, illustrate this point.

Figure 3.6. Curves showing the solubility of Al-N in austenite at temperatures below 1260°C (after Bain).
3.1.9 Effect on the eutectoid point

$A_t$ is lowered by the austenite-formers and raised by the ferrite-formers. A chrome steel of eutectoid composition, e.g. one containing 12% Cr and 0.4% C, will require a higher austenitizing temperature than a eutectoid carbon steel, whereas a 3% Ni steel will already begin to austenitize below 700°C, see Figure 3.9. This state of affairs is clearly of great practical importance when these steels are being used at temperatures around $A_t$.

![Figure 3.8. En 352 (Boferi D.R 44). Grain size as a function of heat-treatment time. Heating at 900 and 1000°C](image)

The eutectoid point in the iron–carbon equilibrium diagram occurs at 0.80% C and 723°C ($A_t$). All alloying elements reduce the carbon concentration of this point. For example, a steel containing 5% Cr has its eutectoid point at 0.55% C. The influence of Cr and Mn is illustrated in Figures 3.10 and 3.11 respectively.

3.1.10 Effect on the temperature of martensite formation

All alloying elements with the possible exception of C, lower $M_f$, the temperature of the start of the martensite formation, as well as $M_s$, the finish of the martensite formation, i.e. at 100% martensite. For the majority of steels containing more than 0.50% C, $M_f$ lies below room temperature.
82 ALLOYING ELEMENTS IN STEEL

This implies that after hardening these steels practically always contain some residual austenite. $M_s$ may be calculated from the equation given below on inserting the percentage concentration of each alloying element in the appropriate term. The equation is valid only if all the alloying elements are completely dissolved in the austenite.

$$M_s = 561 - 474 \text{C} - 33 \text{Mn} - 17 \text{Ni} - 17 \text{Cr} - 21 \text{Mo}$$

For high-alloy and medium-alloy steels Stuhmann has suggested the following equation:

$$M_s(\text{C}) = 550 - 350 \text{C} - 40 \text{Mn} - 20 \text{Cr} - 10 \text{Mo} - 17 \text{Ni} - 8 \text{W} - 35 \text{V} - 10 \text{Cu} + 15 \text{Co} + 30 \text{Al}$$

Of all the alloying elements it is seen that carbon has the strongest influence.

Figure 3.10. Effect of Cr and C on the austenite field (after Bain).
The dashed line traces the austenite field of the base alloy. The displacement of the martensite point, for instance, indicates that an alloy containing only iron and carbon has not been used.

Figure 3.11. Effect of Mn and C on the austenite field (after Bain).

Figure 3.12. Effect of Mn on the $M_s$ temperature (after Rusel and McGuire, Payson and Savage, Zysin, Grange and Stewart).

Figure 3.13. Effect of Ni on the $M_s$-temperature (after Chiusik and Greninger, Payson and Savage, Zysin, Grange and Stewart).
on the $M_s$ temperature (see Figure 1.16). In Figures 3.12 and 3.13 an attempt is made to show in a diagram the experimental results of the effect of Mn and Ni on the $M_s$ temperature of various types of steel.

3.1.11 Effect on the formation of pearlite and bainite during the isothermal transformation

All alloying elements except Co delay the formation of ferrite and cementite, i.e. the transformation curves in the TTT diagram are displaced to the right. It is only that part of the element in solution that affects the transformations. It is very difficult to formulate any general rules regarding the influence exerted by the various alloying elements. However, it has definitely been found that some elements affect the bainite transformation more than the pearlite transformation, while other elements act in the opposite manner. Certain elements will, paradoxically, accelerate the transformations if their concentration increases beyond a certain limiting value, this limit been affected by other alloying elements present. For case-hardening and tool steels the time taken to initiate the pearlite–bainite transformation is reduced as the carbon content exceeds about 1%. For tool steels and constructional steels Si-concentrations of 1.5% and above have been found to promote pearlite formation. This point is further discussed in Chapter 4.

As a general principle it may be stated that by increasing the concentration of one alloying element by some few per cent, the basic carbon content being kept about 0.5%, only a relatively small retardation of the transformation rates is noticed. For plain carbon steels a successive increase in C from 0.30% to 1% produces but a negligible effect. It is only in conjunction with several alloying elements that a more noticeable effect is produced.

PLAIN CARBON STEELS

The diagram in Figure 3.14, applicable to steel W 1 (1% C) will serve as a basis for this discussion. The shortest transformation time for this steel is less than 1/8th second. Note that the time scale is logarithmic; hence there is no zero time. As has been mentioned previously, both pearlite and bainite form simultaneously in this steel at about 550 °C. Since the curves overlap it is customary to draw only one curve. With increasing contents of certain alloying elements, however, the noses of the pearlite and bainite curves will separate.

The structures shown in Figure 3.14 are obtained by austenitizing samples of steel W 1 at 780 °C for 10 min and quenching in a salt bath at various temperatures. After holding them for predetermined times at various temperatures they are finally quenched in water. Before the salt-bath quenching the steel contains undissolved carbides but in view of the composition of the austenite the steel may be regarded as a eutectoid one. The diagram should be studied with the aid of the explanatory text below.

Figure 3.14 TTT diagram for isothermal transformation of steel W1 (1% C steel): A = Austenite, B = Bainite, $M_s$ = start of martensite transformation, $M_p$ = 80% $M_t$ = pearlite
1. Quenching in a liquid bath at 700 °C; holding time 4 min. During this interval the C has separated out, partly as pearlite lamellae and partly as spheroidized cementite. Hardness 225 HV.

2. Quenching to 575 °C; holding time 4 s. A very fine, closely spaced pearlite as well as some bainite has formed. Note that the amount of spheroidized cementite is much less than in the preceding case. Hardness 380 HV.

3. Quenching to 450 °C; holding time 60 s. The structure consists mainly of bainite. Hardness 410 HV.

4. Quenching to 20 °C (room temperature). The matrix consists of, roughly, 93% martensite and 7% retained austenite. There is some 5% cementite as well which has not been included in the matrix figure. Hardness 850 HV.

**Alloy Steels**

In Figures 3.15a-d it is seen that Cr produces approximately the same effect on the pearlite nose as on the bainite nose. At Cr concentrations above around 3%, the noses separate completely. Figures 3.16a-d show that Ni exerts only a modest influence on the transformation and that this element displaces the whole diagram to the right, practically unchanged. The influence exerted by Mo is shown in Figure 3.17, from which it is seen that the pearlite nose is displaced farther than the bainite nose. Si, Mn and W in concentrations up to some 2% have only a very slight effect on prolonging the transformation times.

The transformations are affected to a much greater extent by the simultaneous presence of several alloying elements, even if the sum of their concentrations often is less than that of two of the elements discussed above. In fact, a multiplicative effect is obtained. In such instances the influence of C up to 0.5% is considerable and this is illustrated by the diagrams in Figures 3.18a-d and 3.19a-d. With regard to the Cr–Mo steel, an increase in the C content tends to produce a marked retardation of the pearlite transformation, whereas for the Cr–Ni steel it is the bainite transformation that is retarded with increasing C content. The so-called ternary-alloy steels, besides containing C, Si and Mn, have three additional alloying elements which exert an even more potent influence on the transformation rates. This may be seen in Figures 3.20 and 3.21. Examples of TTT diagrams for high-alloy steels are shown in Figures 3.22, 3.23 and 3.24. It is very interesting to observe the effect produced by boron on the ferrite and pearlite transformations in the first instance. Concentrations as low as 0.001–0.003% B are sufficient to affect very appreciably the rates of these transformations (see Figure 3.25). A small change in the bainite transformation may also occur. Boron produces such a powerful effect that when it is added to steel, the contents of the other alloying elements may be reduced by about a half without causing any change in the positions of the noses of the ferrite and pearlite curves. The enhancing influence of B is most conspicuous in low C steels but dwindles to insignificance as the C increases to eutectoid composition.
Figure 3.16. Ni-alloyed steels TTT diagrams for isothermal transformation (USS): (a) Nominal analysis 0-1\% C, 10\% Ni. Solution temperature 800\(^\circ\)C; (b) Nominal analysis 0-6\% C, 2-9\% Ni. Solution temperature 800\(^\circ\)C; (c) Nominal analysis 0-6\% C, 4-6\% Ni. Solution temperature 800\(^\circ\)C; (d) Nominal analysis 0-6\% C, 5\% Ni. Solution temperature 925\(^\circ\)C.

Figure 3.17. Mo-alloyed steels TTT diagrams for isothermal transformation (USS): (e) Nominal analysis 0.4\% C, 0.2\% Mo. Solution temperature 870\(^\circ\)C; (f) Nominal analysis 0.1\% C, 0.5\% Mo. Solution temperature 870\(^\circ\)C; (g) Nominal analysis 0.4\% C, 0.8\% Mo. Solution temperature 870\(^\circ\)C; (h) Nominal analysis 0.1\% C, 2\% Mo. Solution temperature 1040\(^\circ\)C.
Figure 2.18 Cr-Mo-alloyed steels with different carbon contents. Nominal analysis: 0.8% Mn, 1.1% Cr, 0.2% Mo. Solution temperature 850 °C. TTT diagram for isothermal transformation: (a) Carbon content 0.25% (b) Carbon content 0.35% (c) Carbon content 0.40% (d) Carbon content 0.50% (The above heat has its content of alloying elements close to the upper limit of the composition range)

Figure 2.19 Cr-Ni-alloyed case-hardening steel with different carbon contents. Base analysis: (a) 0.18% C, 0.8% Mn, 0.9% Cr, 1% Ni. STS 351 = S 347 M 17 (En 316); (b) 0.47% C; (c) 0.77% C; (d) 0.97% C
3.1.12 Effect on resistance to tempering

Apart from some exceptions, all alloying elements enhance the ability of the steel to withstand loss of hardness on tempering after hardening. There is no simple relationship between resistance to tempering and alloy content since there are so many factors involved. Jaffe and Gordon\(^4\) have formulated a method of calculation which is applicable to low-alloy steels. According to this method the tempering temperature required to give a desired hardness in the steel may be calculated from its chemical composition. The method assumes that, after the hardening operation, the steel has a mainly martensitic structure. The original expression had the following form:

$$T = 30 (H_e - H_d)$$

where $T$ = temperature °F
$H_e = $ Rockwell hardness calculated from the analysis
$H_d = $ Rockwell hardness desired after tempering

Converting to degrees Celsius the expression becomes:

$$T = 16.67 (H_e - H_d) - 17.8$$
Figure 3.25: TTT diagram for continuous cooling transformation (CCT) for steel 20MnCr5 (0.20% C, 0.30% Cr, 0.50% Mo, with and without boron). Solution temperature 920°C (after Eligil)

Figure 3.26: The hardness increment of H₄ derived from the carbon in the steel: also the hardness as a function of the carbon content after hardening, quenching treatment, and tempering at 100°C (after Jaffe and Gordon)

Figure 3.27: Increments of H₄ derived from various alloying elements (after Jaffe and Gordon)
The various increments of $H_v$ are obtained from the diagrams in Figures 3.26 and 3.27. The expression holds good for a tempering time of 4 h. The grain size of the steel also plays a part in the sense that a fine grain increases the resistance of the steel to softening according to the following scale:

<table>
<thead>
<tr>
<th>Grain size</th>
<th>ASTM</th>
<th>4</th>
<th>6</th>
<th>8</th>
<th>10</th>
</tr>
</thead>
<tbody>
<tr>
<td>Increment of $H_v$</td>
<td>0.6</td>
<td>0.9</td>
<td>1.2</td>
<td>1.5</td>
<td></td>
</tr>
</tbody>
</table>

**ILLUSTRATIVE EXAMPLE**

Which tempering temperature should be used when a hardness of 40 HRC is required for steel S15 25417? Two heats with the following analyses are concerned:

<table>
<thead>
<tr>
<th>Heat No.</th>
<th>% C</th>
<th>% Si</th>
<th>% Mn</th>
<th>% Cr</th>
<th>% Ni</th>
<th>% Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>0.33</td>
<td>0.25</td>
<td>0.60</td>
<td>1.30</td>
<td>1.20</td>
<td>0.15</td>
</tr>
<tr>
<td>B</td>
<td>0.37</td>
<td>0.35</td>
<td>0.80</td>
<td>1.50</td>
<td>1.50</td>
<td>0.25</td>
</tr>
</tbody>
</table>

The grain size is ASTM 6 in both cases. According to the diagrams the following values of $H_v$ are obtained:

<table>
<thead>
<tr>
<th>Heat No. A</th>
<th>$H_v$</th>
<th>Heat No. B</th>
<th>$H_v$</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.33 C</td>
<td>62</td>
<td>0.37 C</td>
<td>63</td>
</tr>
<tr>
<td>0.25 Si</td>
<td>0.15</td>
<td>0.35 Si</td>
<td>0.30</td>
</tr>
<tr>
<td>0.60 Mn</td>
<td>0.95</td>
<td>0.80 Mn</td>
<td>1.20</td>
</tr>
<tr>
<td>1.50 Cr</td>
<td>3.30</td>
<td>1.50 Cr</td>
<td>3.70</td>
</tr>
<tr>
<td>1.20 Ni</td>
<td>0.20</td>
<td>1.50 Ni</td>
<td>0.30</td>
</tr>
<tr>
<td>0.15 Mo</td>
<td>0.20</td>
<td>0.25 Mo</td>
<td>0.60</td>
</tr>
<tr>
<td>6 ASTM</td>
<td>0.90</td>
<td>6 ASTM</td>
<td>0.90</td>
</tr>
</tbody>
</table>

On inserting these values in the expression we obtain:

A  
\[ T = 16.67 (67 - 40) - 17.8 \quad T = 444 \, ^\circ C \]

B  
\[ T = 16.67 (70 - 40) - 17.8 \quad T = 482 \, ^\circ C \]

These results are only approximate. For a more accurate calculation it is necessary to make a few adjustments. Those who require more details are referred to the original sources.

This method is not suitable for calculating the tempered hardness of steels containing appreciable amounts of such alloying elements as give secondary hardening. As a result of systematic research\(^ {15}\) it has been possible to determine the influence exerted by such elements when present in fairly large amounts. In the case of Cr, for instance, it has been found that certain steels containing 2.5% of this element are rendered more resistant to tempering than those containing 5%. In other steels their resistance to tempering increases as Cr increases from 2.5 to 5%. Figure 3.28 shows the influence of Mo, W and V on the resistance to tempering in steels containing 0.3% C and 2.5% Cr. The resistance to tempering is here represented as that temperature that, when applied for 2 h, tempers the steel to a UTS of 150 kp/mm\(^2\) (after Bangard, Mollers and Lemay\(^ {16}\)).

It may be seen from Figure 3.27 that Si in steel has a very potent influence on its resistance to tempering. This effect diminishes as temperatures begin to exceed 600 \(^\circ C\)\(^ {15}\).

### 3.2 GASES

#### 3.2.1 Hydrogen

**RATE OF DIFFUSION AND SOLUBILITY**

Of all the elements hydrogen has the highest rate of diffusion in steel. The diffusion coefficients of hydrogen in \(\alpha\)-iron and \(\gamma\)-iron, respectively, are given as:

\[ \text{Coefficient in } \alpha\text{-iron} = \text{Coefficient in } \gamma\text{-iron} \]

\[ \text{Coefficient in } \alpha\text{-iron} = \text{Coefficient in } \gamma\text{-iron} \]
\[ D_a = 2.2 \times 10^{-3} \exp \left( \frac{-2900}{RT} \right) \text{cm}^2/\text{s} \]

\[ D_p = 1.1 \times 10^{-2} \exp \left( \frac{-9950}{RT} \right) \text{cm}^2/\text{s} \]

The solubility of this element in the different modifications of iron, including molten iron, is shown in Figure 3.29. This difference between the solubilities in γ-iron and α-iron is a feature that is in a large measure responsible for certain specific defects in steel.

The hydrogen content immediately before tapping, of plain-carbon or low-alloy steels made according to normal practice, is approximately as follows:

- Acid open hearth: 3–5 ppm
- Electric-arc, double slag: 4–8 ppm
- Basic open hearth: 5–9 ppm

As the steel cools the solubility of hydrogen decreases and the gas may separate in the molecular state, giving rise to pressure of several 1000 atmospheres which causes cracks in the steel. These defects are commonly known as flakes. Figure 3.30 shows a fractured surface containing typical flakes. Flakes and pores can generally be discovered in the finished steel by means of ultrasonic inspection. The occurrence of flakes and the risk of failure increases with the dimensions of the steel.

Figure 3.30. Fractured surface showing flakes. Natural size

Austenitic stainless steels of the 18/8 type are not as susceptible to the detrimental influence of hydrogen as are the heat-treatable steels. Contents of up to 12 ppm can be allowed in the former steels but plain-carbon and low-alloy steels are liable to develop flakes when the hydrogen concentration is as low as 3 ppm. For small sections up to 5 ppm may be tolerated. The figures cited apply to molten steel; in the finished steel the concentrations are considerably lower.

**HYDROGEN REMOVAL**

On account of the favourable rate of diffusion of hydrogen it is possible, by means of an annealing treatment, to remove that proportion of hydrogen that gives rise to the defects. Figure 3.31 shows the amount of hydrogen evolved for various times up to 100 min and at different temperatures during the vacuum extraction of the steel bars. The amount of hydrogen evolution is a maximum at an annealing temperature of 600°C. In actual practice the hydrogen anneal is carried out between 600 and 650°C.

The amount of hydrogen given off during annealing may be calculated with the aid of the diagram in Figure 3.32 in which \( U \) is set off as a function of \( T \). The plotted values, obtained from a rotor forging during the process of annealing, show satisfactory agreement with the theoretically calculated curve.

\[ U = \frac{\text{Remaining hydrogen content}}{\text{Original hydrogen content}} \]

\[ T = \frac{D \cdot t}{r^2} \]

where \( D = \text{diffusion coefficient} \) (cm²/s)

\( t = \text{time of treatment} \) (seconds)

\( r = \text{diameter of sample} \) (cm)
110 ALLOYING ELEMENTS IN STEEL

The harmful effect of hydrogen may be diminished by allowing the steel to cool slowly after hot working or by subjecting it to an isothermal anneal.

3.2.2 Nitrogen

SOLUBILITY AND RATE OF DIFFUSION

The solubility of nitrogen in iron at various temperatures is shown in Figure 3.33. The diffusion coefficient of nitrogen in α-iron is given as:

\[ D_n = 6.6 \times 10^{-5} \exp \left( -\frac{18,600}{RT} \right) \text{cm}^2/\text{s} \]

It is seen from this expression that the rate of diffusion of N is somewhat higher than that of C at temperatures below 400°C. For γ-iron the diffusion coefficient at 950°C is given as 6.5 × 10^{-8} cm²/s. The more intimate the contact between the molten steel and the atmosphere, the greater is the opportunity for N to be absorbed by the steel. Hence it is not surprising that steel grades produced by the Bessemer or Thomas processes are the ones that contain the highest amounts of nitrogen. The simplest way of reducing the nitrogen content is to prevent contact between the steel and the atmosphere.
ALLOYING ELEMENTS IN STEEL

For the sake of completeness some typical values of nitrogen in steels, produced by different processes, are given in Table 3.1.

Table 3.1

<table>
<thead>
<tr>
<th>Steelmaking process</th>
<th>% N</th>
</tr>
</thead>
<tbody>
<tr>
<td>Bessemer</td>
<td>0.014</td>
</tr>
<tr>
<td>Themos</td>
<td>0.012</td>
</tr>
<tr>
<td>Electric arc</td>
<td>0.006</td>
</tr>
<tr>
<td>Basic open hearth</td>
<td>0.003</td>
</tr>
<tr>
<td>Acid open hearth</td>
<td>0.004</td>
</tr>
<tr>
<td>L.D</td>
<td>0.004</td>
</tr>
<tr>
<td>Kalido</td>
<td>0.003</td>
</tr>
</tbody>
</table>

INFLUENCE OF NITROGEN ON THE PROPERTIES OF STEEL

Nitrogen has both a beneficial and a harmful influence on the properties of steel. As an example of a beneficial influence may be mentioned the properties obtained on nitrizing which usually takes place in ammonia gas at 510°C. The nitrogen combines with the iron and certain alloying elements to form hard, wear-resisting nitrides. The extreme surface hardness is, however, attributable rather to the state of fine dispersion of the nitride particles than to the inherent hardness of the nitrides themselves. Another example of an advantage accruing from the presence of nitrogen is the capacity of the latter for raising the yield point of 18/8 steel (see Figure 3.34)\(^9\). The practical details of the nitriding process are described in Chapter 6.

The harmful influence of nitrogen begins to make itself felt when the concentration exceeds 0.005%\(^n\), provided that the steel has not been killed with Al. Higher contents promote so-called strain-ageing which may lead to harmful effects. One result of strain-ageing is that the yield point of the material is raised if, prior to the test, it had been deformed beyond its original yield point and then aged at some temperature between room temperature and 300°C. Strain-ageing is said to be caused by an accumulation of N atoms at the dislocations. During the course of the first deformation the dislocations move away from the N atoms, but on subjecting the steel to ageing these atoms find their way again to the lattice defects and once in their new positions they are a contributory cause of impeding the movement of the dislocations on renewed deformation. Carbon produces an effect similar to that of N but since the rate of N diffusion at low temperatures is higher than that of C—at 100°C it is more than double—it is most likely that N is responsible for the main part of the strain-ageing effect. Strain-ageing makes the steel brittle
at room temperature. The transition-temperature range, as manifested in impact testing, is displaced to higher temperatures. Figure 3.35. An impact test carried out in the low-impact region shows a brittle fracture. It is this type of fracture that was found in connexion with the catastrophic failures occurring in ships and bridges. As recently as some thirty years ago these failures were regarded as complete mysteries.

As mentioned previously the simplest way to reduce the dangerous influence of N is to kill the steel with Al which combines with the N to form nitrides. At the same time the steel is rendered less susceptible to grain growth, which in turn leads to a displacement of the transition-temperature range to lower temperatures. It appears that the N content is not particularly critical when the steel has been Al killed. Standard specifications for steel which allow 0.009% N maximum have a rider to the effect that up to 0.015% N is permissible for a fine-grained steel, i.e. a steel that after deoxidation has received an effectual addition of Al.

12% chrome steels containing more than about 0.010% N are subject to reduced impact strength, but also to a rise in the tempering hardness level.

3.2.3 Oxygen

Rate of Diffusion

Oxygen has a very small diffusion coefficient, in fact it is the smallest one of the gases at present under consideration. For γ-iron the following value is given:

\[ D_o = 1.2 \exp \left( - \frac{50000}{RT} \right) \text{cm}^2/\text{s} \]

In unalloyed and low-alloy good-quality constructional steel the oxygen concentration is about 60 ppm. By means of refining processes (deoxidation) or vacuum degassing the oxygen can be reduced to about 30 ppm and by vacuum remelting to below 10 ppm.

Influence of Oxygen on the Finished Steel

Investigations into the influence of oxygen on the mechanical properties of pure iron have shown that in amounts up to saturation level it has no noticeable effect on the ultimate tensile strength, yield point or hardness at temperatures down to −73°C. However, values of elongation and reduction of area diminish rapidly as the oxygen increases and the temperature falls. Further, impact strength is reduced and the transition-temperature range is raised as oxygen increases.

Some harmful effects, e.g. poor surface finish, may be produced by otherwise acceptable oxygen contents since some alloying elements form oxides that may remain in the steel as non-metallic inclusions. The amount of micro-inclusions, which may be estimated on ground and polished samples in the metallurgical microscope, is an approximate measure of the oxygen content. On the other hand, macro-inclusions derive generally from some local oxidation in connexion with tapping or teeming and are not directly related to the mean oxygen content of the steel. The amount of macro-inclusions may be estimated with reference to various scales, one of which is shown in Figure 3.36. The samples, in the shape of discs, are hardened, tempered and then fractured and blue-tempered which reveals the inclusions very clearly. It is also possible to estimate the amount of macro-inclusions in test bars which, after machining and grinding, are subject to visual inspection. The amount and types of micro-inclusions may be assessed from Jerkonslof’s Slag Inclusion Chart, the ‘JK’ chart (Figure 3.37), which has gained international recognition. Compare also ASTM Designation E 45: Determining Inclusion Content of Steel.
3.3 NEW STEELMAKING PROCESSES

For the production of high-quality steel the basic electric-arc furnace is the predominant unit. The furnace time is divided into a melt-down period, an oxidizing and dephosphorizing period and finally a deoxidizing and desulphurizing period. Since the introduction of oxygen lancing the oxidizing and dephosphorizing periods have been shortened. On the other hand it has not been found possible to obtain a more economic or efficient oxygen and sulphur removal in the electric-arc furnace than the ordinary deoxidizing and desulphurizing practice. As a consequence there has been intensive research during the last decade into steelmaking practices in order to develop steel production methods that will yield steel of high cleanliness. The primary aim is to reduce the oxygen and sulphur contents, since oxides and sulphides have a deleterious influence on the properties of steel, including surface finish, impact strength and fatigue strength. During the 1960s several such methods were developed. These newer steelmaking practices have had a significant influence on the properties of steel and therefore a brief account of these methods is given in the following:

3.3.1 Vacuum remelting

Vacuum-melted steel is produced in a specially designed high-frequency induction furnace under a reduced pressure of about $10^{-4}$ torr (1 torr = 1 mm Hg) or by melting conventional air-melted steel in a special electric-arc furnace at a reduced pressure of about $10^{-3}$ torr. Figure 3.38 illustrates schematically a vacuum electric-arc remelting furnace (also called consumable-electrode furnace). A heavy-section air-melted steel electrode is progressively melted down in an electric arc, while the molten steel is allowed to solidify in a water-cooled copper mould. The reduced pressure promotes certain metallurgical reactions which, for the remelted steel, result in a large reduction in the contents of hydrogen and oxygen, and also nitrogen to some extent. Vacuum-remelting is applied to steels and alloys intended for elevated-temperature service as well as to constructional and tool steels for which a high degree of cleanliness is stipulated. The quality of the steel is enhanced in several respects, in particular the mechanical properties in the transverse direction, i.e. at right angles to the direction of rolling or forging. The increased cleanliness of this steel is conducive to its being stipulated for parts requiring surfaces free from non-metallic inclusions, for example parts to be highly polished or electroplated. Its high fatigue properties make it attractive as material for ball-bearings, springs, rolls for cold rolling, etc. The remelting and re-solidification of steel in the vacuum electric-arc furnace result in a reduced amount of segregation. Hence, subsequent operations involving rolling or forging are facilitated which is of especial advantage for
steels and alloys intended for elevated-temperature service. For such vacuum-
remelted steels there are additional advantages, e.g. enhanced creep properties
and reduced notch sensitivity. Figures 3.38a-f show examples of improvements
obtained as a result of changing from air-melted to vacuum arc remelted
steal.

3.3.2 Electroslag refining

This process is called the ESR process (Electroslag Refining or Remelting)
in English-speaking countries and also in Sweden. The German designation
is ESU (Elektroschmelz-Umschmelzung). The method was developed
during the 1960s and towards the end of this period some 400000 tons per
annum were produced in the USSR where the process had reached its
greatest development. In the USA the corresponding figure was only some
400000 tons per annum.

The principle of ESR is illustrated in Figure 3.40. A consumable steel
electrode dips into a hot slag bath and melts progressively. An ingot is thereby
built up in a water-cooled mould. The method has certain similarities to
vacuum arc remelting (VAR) but a distinctive difference is that in the ESR
process the metallurgical reaction involved in purifying the melting electrode
takes place in the slag bath, first between this and the thin liquid-steel film

Figure 3.38. Schematic sketch of the vacuum electric-arc remelting furnace at the Bofors company

![Figure 3.38](image)

Figure 3.39. Comparison between the properties of air-melted steel and vacuum electric-arc
remelted (VAR) steel: (a) Oxygen content of high-temperature 17Cr steel; (b) Inclusion-
content; (c) Impact strength of Cr-Ni-Mo steel 30 Cr24 NbM (En 24); (d) Dimension 30 mm ×
UTS 95 kp/mm²; (e) Air-melted; (f) Time to creep rupture. High-temperature steel A 286;
UTS 150 kp/mm²; (g) Creep rupture elongation of high-temperature steel A 286;
on the electrode tip and then as the steel droplets drip through the slag. The purifying reaction in the VAR process takes place as a result of the vacuum. In the ESR process the heat necessary to melt the electrode is created in the liquid slag by its acting as an electric resistance to the current passing through it. In the case of the VAR process free electric arcs are struck in the vacuum between the electrode and the ingot. Direct current is necessary for this process whereas the ESR can use either dc or ac, the latter being preferable.

![Figure 3.40](image1.png)  
**Figure 3.40**: Sketch illustrating the principle of the ESR process for steel production

The slag is usually made up of the three ingredients: quicklime (CaO), fluorspar (CaF₂) and alumina (Al₂O₃). The lime is present mainly to remove sulphur and oxygen from the steel; the fluorspar lowers the melting point of the slag and increases its fluidity; the alumina increases the electric resistance of the slag which in turn causes its temperature to rise and thereby increases the melting rate of the steel electrode. The relative proportions of the constituent ingredients are generally a compromise between cleaning effect, melting rate and ingot surface. A standard composition is 60% CaF₂, 20% CaO and 20% Al₂O₃ but the optimum composition varies widely for different grades of steel.

By optimizing conditions so that an acceptable melting rate, ingot structure and surface are obtained it is possible to effect, for most grades of steel, a removal of 50–70% of the sulphur and about 50% of the oxygen. Hereby the amounts of sulphide and oxide inclusions respectively are reduced in direct proportion in the remelted steel. However, the hydrogen and nitrogen contents are not changed by the ESR process. The evenly fine-grained structure of ESR ingots ensures that a smaller forging reduction can be

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120 ALLOYING ELEMENTS IN STEEL

accepted than would be the case for conventionally teemed ingots. Thanks to the absence of pipe a considerably greater yield is obtained.

Tests carried out with ESR steel have shown that, compared with air-melted steel, noteworthy improvements in physical properties are obtained, in particular fatigue properties, impact strength and ductility. The improvements have been most striking in transverse samples taken from central areas. In comparison with air-melted steel the degree of scatter in the values obtained has in some instances diminished appreciably. See Figure 6.30.

3.3.3 Vacuum degassing

Vacuum degassing can be carried out by several methods. In principle the molten steel is subjected to a reduced pressure of about 1 torr (1 mm Hg), either while it is being tapped from the furnace or after completed tapping. When ladle degassing, for example, the ladle is placed in a vacuum chamber or alternatively, the ladle is fitted with a vacuum-tight cover and the de-

![Figure 3.41](image2.png)  
**Figure 3.41**: Sketches to illustrate the principle of steel production sequence according to the ASEAS method

1. Tapping the molten steel into the ladle-furnace.
2. Degassing.
3. Reheating.
4. Teeming
gassing is started. The degassing unit used at Bofors (ASEA–SKF method) consists of a ladle-furnace which, during the actual degassing operation, is placed inside an induction coil stand in order to set up a stirring action in the melt. This in turn facilitates gas evolution and also gives an even distribution of the alloying elements in the melt. When the chemical composition has been adjusted the steel can be reheated by means of the electric-arc electrodes (see Figure 3.4.1). The degassing process removes about one half of the amount of dissolved gases, which implies that vacuum degassing produces roughly the same cleanness as deoxidized and hydrogen-annealed steel, i.e. steel containing about 30 ppm of oxygen and 2.5 ppm of hydrogen.

3.3.4 The Bofors method of sulphur removal

When processing steel by the ASEA–SKF method, no sulphur is removed in the ladle-furnace. When it is stipulated that the steel shall have a low sulphur content it must be desulphurized before it is transferred to the ladle-furnace. In order to develop a method whereby desulphurization could be carried out in the degassing unit, Bofors instituted a joint development project with ASEA, the Royal Institute of Technology in Stockholm (KTH) and the Metallurgical Research Station at Luleå. Based on the results obtained from this investigation the first trial melts were carried out in the Company’s 50-ton capacity ladle-furnace. The outcome was the Bofors method of sulphur removal.

The desulphurization process was primarily developed to operate in conjunction with the ASEA–SKF process but it is adaptable to other processes which incorporate a strongly basic fluid slag and good stirring facilities. In principle the method consists of obtaining an initial S removal in the carefully deoxidized steel by means of a strongly basic slag, using an ASEA–SKF ladle-furnace. Then, if required, the S content may be reduced still further by the addition of some agent that has a strong affinity for S, e.g. Ca or Ce. It is essential that the oxygen activity in the steel be kept low during the whole course of the desulphurizing process and right up to the point of teeming. By means of a strongly basic slag alone, a 30% S reduction is obtainable, which is quite satisfactory in most cases. By an addition of misch metal (an alloy of the rare-earth metals containing about 53% Ce) the S content of unalloyed steel may be reduced to extremely low values, e.g. below 0.001%. Alloy constructional steels are available with S below 0.003%.

The mechanical properties are enhanced by a reduction of the S content, in particular ductility, impact strength and fatigue strength in the transverse direction of the steel. This diminished anisotropy is due to the almost complete absence of non-metallic inclusions.

3.3.5 Effect of sulphur on the properties of steel

The following section should be discussed under the heading ‘Solids’, but since the coming information is based on the results obtained from the application of the Bofors method of sulphur removal it is thought more appropriate to insert this section immediately after a description of this method. Steel always contains large numbers of non-metallic inclusions which, depending on their shape, size and composition, can influence the properties of steel in various ways. Sulphur forms plastic inclusions (FeMnS). The amount of sulphides is determined by the S content of the steel but their shape is to a large extent determined by the deoxidation technique and also by the composition of the steel.

As a general rule, non-metallic inclusions impair the mechanical properties of steel, especially the fatigue strength and the transverse ductility. Sulphides, however, are often regarded as more or less innocuous, in fact it has been reported that S has a beneficial influence on the fatigue strength of ball-bearing steel. As regards transverse fatigue properties, it is considered that sulphide inclusions—at least the larger ones—have a deleterious influence on these properties. This is substantiated in Figure 3.4.2 which shows the

![Graph showing the relationship between sulphur content and endurance limit](image)

**Figure 3.4.2**: Endurance limit (alternating stress: tension-compression) in transverse direction v. varying sulphur contents of steel S1S 2225 heat treated to 250–290 HB. Nominal analysis: 0.25% C, 1% Cr, 0.25% Mo

alternating-stress transverse endurance limit of steel S1S 2225 in the hardness range 250–290 HB, made according to conventional practice and with S contents from 0.008 to 0.035%. Although the scatter is large the values show that for the low-sulphur steels there is a considerable improvement in fatigue strength.

The tensile strength of the steel is scarcely affected by low S contents but from Table 3.2 it is apparent that as the S content is reduced the elongation and reduction of area are improved, in particular if measured in the transverse...
### Table 3.2 Mechanical Properties of Normalized Steel SIS 1650 (C 50)

Specimen 150 mm square, $L =$ longitudinal, $T =$ transverse

(Mean values of test results obtained from samples taken from top, middle and bottom of ingot)

<table>
<thead>
<tr>
<th>Sulphur content (%)</th>
<th>Position of specimen</th>
<th>Yield point $k_p/mm^2$</th>
<th>Ultimate tensile strength $k_p/mm^2$</th>
<th>Elongation $%$</th>
<th>Reduction of area $%$</th>
</tr>
</thead>
<tbody>
<tr>
<td>$&lt;0.001$</td>
<td>$L$</td>
<td>38</td>
<td>67</td>
<td>24</td>
<td>52</td>
</tr>
<tr>
<td></td>
<td>$T$</td>
<td>38</td>
<td>67</td>
<td>23</td>
<td>47</td>
</tr>
<tr>
<td>$0.01$</td>
<td>$L$</td>
<td>39</td>
<td>68</td>
<td>23</td>
<td>52</td>
</tr>
<tr>
<td></td>
<td>$T$</td>
<td>39</td>
<td>69</td>
<td>20</td>
<td>36</td>
</tr>
<tr>
<td>$0.012$</td>
<td>$L$</td>
<td>37</td>
<td>68</td>
<td>22</td>
<td>44</td>
</tr>
<tr>
<td></td>
<td>$T$</td>
<td>36</td>
<td>64</td>
<td>10</td>
<td>11</td>
</tr>
</tbody>
</table>

**Figure 3.43**: Impact strength of steel SIS 2913, BS 420537 (En 56C) with different sulphur contents. Longitudinal specimen, 20 mm diameter, heat treated to 222 HB. Nominal analysis: 0.28% C, 17% Cr, 0.25% Mo.

**Figure 3.44**: Impact strength of steel SIS 2225 with different sulphur contents. Specimen 100 mm diameter. Heat treated to 222 HB. Longitudinal and transverse specimens. Nominal analysis: 0.28% C, 17% Cr, 0.25% Mo.

A general consequence of a reduced S content is an improvement in the impact properties, both in the longitudinal and transverse directions, but most markedly in the latter. Figure 3.43 shows typical impact-strength curves for a martensitic stainless 12% Cr steel. A reduction in the sulphur content has resulted in a marked improvement in the longitudinal impact strength which would imply a displacement of the transition-temperature range by about 40°C towards lower temperatures. At the same time there is an increase in the absolute values of the impact strength in the upper range. The impact strength of steel SIS 2225, hardened and tempered, is shown in Figure 3.44. A reduction of sulphur from 0.012 to 0.005% has not affected the

longitudinal impact-strength but the transverse impact value has been practically doubled. Figure 3.45 shows how the transverse impact strength of a degassed 0.35% C Cr-Ni-Mo steel, hardened and tempered, increases as the S is reduced.

In pipelines for the transport of gas and oil through areas with low temperatures, a steel is required which has a low transition temperature. This
ALLOYING ELEMENTS IN STEEL

Figure 3.45: Impact strength of transverse specimens, hardened and tempered to 550 H. B. of steel containing 0.35% C, 1.4% Cr, 3% Ni, 0.4 Mo and varying sulphur contents. Testing temperature: -40°C.

can be obtained by using a steel of an extremely low S content. The only steelworks which in 1971 was able to undertake the manufacture of a steel of max. 0.005% S was AB Bofors, which company on behalf of Molycorp made a 50-ton charge of the stipulated steel X-80 according to the ASEA-BOFORS method. Careful examinations carried out by the customer established that the properties desired were obtained by using the steel supplied.

Sulphur is used as an alloying element in many constructional steels in order to improve their machinability. It might therefore be expected that by reducing the S the machinability should be impaired. Results from some machinability tests are given in Table 3.3. The machinability index refers to the standard test method used by Bofors. \( B_{\text{m}} \) and \( B_{\text{mt}} \) indicate the cutting speed for a tool life of 30 min, and \( B_{\text{b}} \) a tool life of 45 min. The results indicate that while the drilling properties of the steel are somewhat impaired, the turning properties are not appreciably affected by a reduction in the S content.

### Table 3.3: DATA OBTAINED FROM MACHINING TESTS OF NORMALIZED STEEL SIS 1650 (C 0.5)

<table>
<thead>
<tr>
<th>Sulphur content %</th>
<th>SIS 1650</th>
<th>Hardness HB</th>
<th>Drilling High-speed steel ( B_{\text{m}} ) m/min</th>
<th>Drilling High-speed steel ( B_{\text{mt}} ) m/min</th>
<th>Turning Cemented carbides ( B_{\text{b}} ) m/min</th>
<th>Turning Cemented carbides ( B_{\text{b}} ) m/min</th>
</tr>
</thead>
<tbody>
<tr>
<td>&lt;0.001</td>
<td>1650</td>
<td>195</td>
<td>40</td>
<td>32</td>
<td>140</td>
<td>140</td>
</tr>
<tr>
<td>0.001–0.01</td>
<td>1650</td>
<td>197</td>
<td>48</td>
<td>32</td>
<td>135</td>
<td>135</td>
</tr>
<tr>
<td>0.005</td>
<td>1650</td>
<td>198</td>
<td>48</td>
<td>33</td>
<td>140</td>
<td>140</td>
</tr>
<tr>
<td>0.012</td>
<td>2225</td>
<td>229</td>
<td>18</td>
<td>33</td>
<td>160</td>
<td>160</td>
</tr>
</tbody>
</table>

REFERENCES
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7. **Ilgem, H.,** "A Recently-developed Direct-hardening Boron- alloyed Case-hardening Steel, Possessing a High Core Strength", New Härter, 16, No 4, 238–243 (1971) (in German)
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14. **Bungardt, K., Mölders, O., and Lennartz, G.,** "Influence of Chromium, Molybdenum, Tungsten and Vanadium on the Properties of Steels containing 0.3% C", Archiv für das Eisenhüttenwesen, 32, No 12, 823–831 (1961) (in German)
4 Hardenability

4.1 General Remarks

Provided the rate of cooling is greater than the critical cooling rate, i.e.
the rate at which the formation of pearlite or bainite is just suppressed, the
hardness obtained on quenching depends principally on the C content of
the steel. If the cooling rate is lower than the critical rate the amount of martensite
is reduced, thus lowering the overall hardness of the steel. Carbon content in
this context means the amount of C dissolved in the austenite. Carbon which
remains combined as carbide after the austenitizing treatment does not take
part in the martensite reaction and has therefore no influence on the hardness
of the martensite. The relationship between the hardness, carbon content
and amount of martensite is shown in Figure 4.1.

![Graph showing relationship between hardness, carbon content, and amount of martensite](image)

Figure 4.1. Relationship between hardness, carbon content, and amount of martensite (after Hodge and Orehovski)

Hardenability is the ability of the steel to harden by the formation of
martensite on quenching. The hardenability determines the depth of hardening
obtained on quenching, which is usually specified as the distance below the
surface where the amount of martensite has been reduced to 50%, or more
precisely to 50%, martensite and bainite. The characteristic property of a steel
possessing high hardenability is that it shows a large depth of hardening or
that it hardens through entirely in heavy sections. Without electron micro-
scopy it may be difficult to distinguish martensite from lower bainite.

Since the depth of hardening is of great importance to tools and con-
structional parts it is customary to indicate this property by means of diagrams
of the type shown in Figure 4.2. When the depth of hardening is measured

![Diagram showing depth of hardening for various grades of steel](image)

Figure 4.2. Depth of hardening for various grades of steel. Bars 100 mm diameter. Steel W1
(Bofors B 20 V) water quenched, the rest oil quenched.

in this way the cooling medium must also be stated. In this instance the plain
carbon steel, W 1, was quenched in water and the other steels in oil. The
heavier the section to be hardened the smaller is the depth of hardening and
the lower is the core hardness, which is illustrated in Figure 4.3. The reason
why a steel is harder at the surface than at the centre is explained by referring
to a continuous-cooling-transformation (CCT) diagram. By studying the
schematic CCT diagram in Figure 4.4 it is obvious that since the surface
cools at a considerably faster rate than the centre, the cooling curve repre-
senting the surface will pass in front of the ferrite and bainite noses and as a
result only martensite is formed. At the centre which cools more slowly some
bainite will be formed, as may be inferred from the figure, and this will result
in a lower hardness in the core.
130 HARDENABILITY

As the dimensions of the steel increase, the rate of cooling decreases and the core hardness will be still further reduced owing to the formation of ferrite and pearlite. The surface hardness will also decrease when the cooling curve is so displaced as to be to the right of the critical cooling curve.

Fundamental hardenability data are of considerable use to the steel consumer and heat treaters and therefore a number of simple methods have been developed whereby hardenability can be determined. Some of the best known methods are described below.

4.2 THE GROSSMANN HARDENABILITY TEST

To determine hardenability according to Grossmann’s method, a number of cylindrical steel bars of different diameters are hardened in a given cooling medium. By means of metallographic examination the bar that has 50% martensite at its centre is singled out and the diameter of this bar is designated as the critical diameter (Dc), the unit generally being inches.

The cooling intensities of the different cooling media have been determined and are called the H-factors. The values of H are given in Table 4.1. Using

<table>
<thead>
<tr>
<th>Agitation</th>
<th>Coefficient of severity of quench</th>
<th>H</th>
<th>Cooling medium</th>
</tr>
</thead>
<tbody>
<tr>
<td>None</td>
<td>0.25-0.30</td>
<td>0.9-1.0</td>
<td>2.0</td>
</tr>
<tr>
<td>Mild</td>
<td>0.30-0.35</td>
<td>1.0-1.1</td>
<td>2.0-2.2</td>
</tr>
<tr>
<td>Moderate</td>
<td>0.35-0.40</td>
<td>1.2-1.3</td>
<td></td>
</tr>
<tr>
<td>Good</td>
<td>0.4-0.5</td>
<td>1.4-1.5</td>
<td></td>
</tr>
<tr>
<td>Strong</td>
<td>0.5-0.8</td>
<td>1.6-2.0</td>
<td></td>
</tr>
<tr>
<td>Violent</td>
<td>0.8-1.1</td>
<td>4.0</td>
<td></td>
</tr>
</tbody>
</table>

The appropriate value of the H-coefficient of the cooling medium under consideration, the Dc-value can be converted to the ideal diameter Dc which is defined as the bar diameter which, when the surface is cooled at an infinitely rapid rate (H = ∞), will yield a structure, at the centre, containing 50% martensite. The diagrams correlating Dc and Dc are shown in Figure 4.5.

The value of Dc obtained is hence a measure of the hardenability of the steel and is independent of the cooling medium. In practice the Dc-values are used to determine the values of Dc for bars quenched in various cooling media, using the diagrams in Figure 4.5.

EXAMPLE

By subjecting a steel whose Dc value is 2.0 in to an oil quench, the H-coefficient of which is 0.4, it would yield a Dc-value of 0.8 in. Familiarity with concept of hardenability enables a good indication of the hardenability of a steel to
be obtained from its $D_1$-value since this is a useful figure for comparison purposes.

4.2.1 Calculation of $D_1$-values from the chemical composition

The hardenability may be calculated from the composition of low-alloy and medium-alloy steels, taking into account only the amount of each element in solution at the austenitizing temperature. The austenite grain size must also be taken into consideration. The smaller the grain size the lower is the hardenability. This is due to the fact that the total surface area of the grain boundaries increases as the grain size decreases, thereby in turn giving rise to an increasing number of nuclei which serve as initiating points for pearlite formation.

The computation starts from the C content and the grain size. By means of Figure 4.5 a 'base' hardenability characteristic for $D_1$ is obtained. For the other alloying elements the curves in Figure 4.7 indicate the multiplying factor that corresponds to each alloy content. The factors given in this diagram have been selected from a number of test results obtained by different research workers and approved by the American Iron and Steel Institute.

Figure 4.6 is applicable to C contents above 0.5%, but only on the assumption that all carbides are in solution at the austenitizing temperature. However, this is generally not the case since an unnecessarily high temperature would then have to be employed. Furthermore, complete dissolution of the carbides might result in deleterious grain growth effects and high retained austenite content in the steel. Consequently if conventional hardening temperatures are used for low-alloy steels lying in the higher C ranges of the
equilibrium diagram, a falling-off in the hardenability must be expected when the C exceeds 0.8%. This is because the excess carbon combines with carbide and hardenability-inducing elements such as Cr and Mo. In spite of this reduction in hardenability, steels are still alloyed with about 1% C, but under these circumstances the carbides are beneficial in increasing the wear resistance of the steel.

**Examples of Hardenability Calculations**

1. Steel SIS 2225, ASTM grain size 7, has the following composition:

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Mo</th>
</tr>
</thead>
</table>
   | 0.25  | 0.3   | 0.7| 1.1| 0.2%

   From Figure 4.6 the base value of \( D_b \) is 0.17 in. On multiplying this value with the appropriate factors we obtain:

   \[ D_i = 0.17 \times 1.2 \times 3.3 \times 3.4 \times 1.6 = 3.7 \text{ in} \]

2. Steel SIS 2541 (BS 816M40), ASTM grain size 6, has the following composition:

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
</tr>
</thead>
</table>
   | 0.35  | 0.3   | 0.7| 1.4| 1.4| 0.2%

   \[ D_i = 0.22 \times 1.2 \times 3.3 \times 4.0 \times 1.5 \times 1.6 = 8.4 \text{ in} \]

The \( D_i \)-values obtained may be converted to \( D_b \) by means of diagrams in Figure 4.3 as described above. For example, by quenching in oil with moderate agitation (\( H = 0.4 \)) the critical diameter of steel SIS 2225 is \( D_b = 2 \text{ in} \) and of steel SIS 2541, \( D_b = 6.4 \text{ in} \). These values are marked in the diagram, Figure 4.5.

In order to calculate \( D_i \) slide rules are obtainable which have scales graduated according to the different multiplying factors. Values of \( D_i \) calculated as above, are only approximate but they are useful as a means of comparing the hardenability of different grades or heats.

The hardenability factors proposed by Grossmann have been re-examined by Moser and Legat and a more precise relationship between grain size and C content has been obtained and this is presented in Figure 4.8. The revised multiplication factors by Moser and Legat are shown in Figure 4.9.

**Figure 4.7. Multiplying factors for different alloying elements for hardenability calculations (after AISI)**

**Figure 4.8. Relationship between base hardenability, carbon content and grain size as obtained in actual practice (after Moser and Legat)*

Using the results obtained by Moser and Legat the hardenability may be calculated from the following expression:

\[ D_i = D_{1C} \times \frac{\% \text{ Mn}}{2.11} \times \frac{\% \text{ Si}}{2.13} \times \frac{\% \text{ Cr}}{3.275} \times \frac{\% \text{ Mo}}{1.47} \]

Recalculated values of \( D_i \) for the steels in the examples are as follows. (Grossmann’s values are given in brackets.)

- SIS 2225 \( D_i = 2.5 \text{ in (3\,7 in)} \)
- SIS 2541 \( D_i = 8.0 \text{ in (8\,4 in)} \)
The agreement is not very good for the first steel but is acceptable for the second.

Kramer, Siegel and Brooks\(^2\) have also examined Grossmann's equations and have published a diagram that is practically identical with the one in Figure 4.8. Jatczak\(^3\) has extended this diagram by adding multiplying factors for other alloying elements. Fairly close agreement with the values of Moser and Legat were obtained in the case of low- and medium-carbon steels. Jatczak has also published multiplying factors for steels containing C between 0.90% and 1.10% and which take account of the hardening temperatures. For these cases calculations can proceed straight from the composition of the steel without the necessity of applying any corrections for such alloying elements as are not in solution in the austenite. A comparison between the diagrams in Figures 4.10 and 4.11 shows the effect of raising the hardening temperature.

As a practical example illustrating the effect produced by the C content in the range mentioned, it was found that the depth of hardening in grinding-mill rods, made from a shallow-hardening steel, increased from 11 to 15 mm as the C content was reduced from 0.90% to 0.85%.

### Figure 4.9
Multiplying factors for calculating the hardenability as influenced by Mo, Mn, Cr, Si and Ni (after Moser and Legat\(^1\)).

Jatczak's factors for Si, however, are not in agreement with the results obtained at Bolors where it has been shown beyond question that the hardenability of steels containing about 1% C first falls as the Si content increases to about 1% and then rises as the Si increases to about 2%. The changing influence of Si in this range of composition is illustrated in Figure 4.11.

### Figure 4.10
This chart is used for determining the multiplying factors for high-carbon steels austenitized at 830°C (1525°F) (base C is 1.1%). (after Jatczak\(^3\)).

### Figure 4.11
This chart is used for determining the multiplying factors for high-carbon steels austenitized at 830°C (1525°F) (base C is 1.1%). (after Jatczak\(^3\)).
HARDENABILITY

4.12 which shows the hardenability, denoted by the Jominy distance to 500 HV, as a function of the Si content. As the Mn content increases, the troughs of the curves are displaced towards higher Si contents.

4.3 THE JOMINY END-QUENCH HARDENABILITY TEST

On account of the high cost of Grossmann's method it is used nowadays only to a very small extent, albeit it must be regarded as the most exact test. The most commonly used method at present has been developed by Jominy. For this test a round bar specimen is used, 25 mm in diameter and 100 mm in length. The specimen is heated to the hardening temperature of the steel, with a holding time of usually 20 min. One end-face of the specimen is quenched by spraying it with a jet of water, as illustrated in Figure 4.13. Hereby the rate of cooling decreases progressively from the quenched end along the length of the bar. When it is cool, two diametrically opposite flats, 0.4 mm deep and parallel to the axis of the bar are ground and the hardness

Figure 4.13. Hardening of Jominy end-quench specimen
is measured along the flats. The hardness values are plotted in a diagram against their distances from the quenched end. Jominy curves for some steel grades are shown in Figures 4.14 to 4.23. The upper curve represents the maximum hardness values corresponding to the upper composition limit of the steel and the lower curve the minimum hardness values corresponding to the lower limit of the composition range. Together the curves form what is called a Jominy or hardenability band.

4.3.1 Calculation of Jominy curves from the chemical composition

Using regression analysis and proceeding from the chemical composition, Just has derived expressions that enable the hardness at different Jominy distances to be calculated direct. It is found that all alloying elements have an increasing influence up to a Jominy distance of about 10 mm, but beyond this their influence remains practically constant. Carbon starts off at a Jominy distance zero with a factor of 50 but all other alloying elements have factor 0 at distance zero. This implies that the hardness at Jominy distance zero is solely governed by the carbon content. Using the derived factors it is possible to prevent a hardenability expression for each Jominy distance. Since the factors vary by only negligible amounts for Jominy distances exceeding 7 mm, a single expression can be formulated for the whole test specimen except for distances shorter than 6 mm.

The equation given below is taken from a report of a paper by Just.

\[ J_{6-80} = 95V - C - 0.0028s^2 + C + 20Cr + 38Mo + 14Mn + 6Ni + 6Si + 39V + 96P - 0.8K - 0.75s + 0.9s - 13HRC \]

where \( J \) = Jominy hardness
\( s \) = Jominy distance
\( K \) = ASTM grain size

The equation is valid for Jominy distances from 6 to 80 mm. The following limits are specified for the alloying elements:
C < 0.6\% Cr < 2\% Mn < 2\% Ni < 4\% Mo < 0.5\% V < 0.2\%

For Jominy distances shorter than 6 mm the influence of alloying elements other than C on the hardness is negligible. Hence the following equation holds for the end-face of the Jominy specimen:

\[ J_0 = 60 \times \sqrt{C} + 20 \quad \text{HRC} \quad (C < 0.6\%) \]

It is found that the C content on the one hand and the rest of the alloying elements on the other interact mutually on the hardness. For instance, Cr acts more powerfully in medium-carbon steels than in low-carbon, i.e. case-hardening, steels.

The article already referred to, contains a number of expressions based on available Jominy curves. These expressions are to a great extent dependent on the origins of these curves. Thus, for steels specified by SAE, USS-Atlas and MPI-Atlas, the expressions have different forms. Medium-carbon steels for hardening and tempering) and low-carbon (case-hardening) steels have been given different expressions for hardenability, as seen below:

\[ J_{sc-10} \quad \text{(case-hardening steel)} \]

\[ = 74 \sqrt{C} + 14 \text{Cr} + 5.4 \text{Ni} + 29 \text{Mo} + 16 \text{Mn} - 16.8 \sqrt{s} + 1.386 s + 7 \text{HRC} \]

\[ J_{sc-10} \quad \text{(steel for hardening and tempering)} \]

\[ = 102 \sqrt{C} + 22 \text{Cr} + 21 \text{Mn} + 7 \text{Ni} + 33 \text{Mo} - 15.47 \sqrt{s} + 1.102 s - 16 \text{HRC} \]

In order to assess the accuracy of the first-mentioned expression a calculation has been carried out for two case-hardening steels and two steels for hardening and tempering, for which Jominy curves have been determined according to SIS. As may be seen from Figures 4.24 to 4.27, in some instances the conformity is reasonably satisfactory.

### 4.3.2 Practical applications of Jominy curves

Jominy curves may be used in the first instance to compare the hardenabilities of different heats of the same grades of steel, hereby serving as a valuable method of quality control. For more than 20 years there have been available in the USA special so-called H-steels for which there are specifications defining their hardenability bands (BS 970: Part 2:1970 and BS 970: Part 3:1971 also contain H-steels.)

The curves may also be used to predict the expected hardness distribution obtained on hardening steel bars of different dimensions by cooling in various quenching media. The rates of cooling prevailing at different distances in the Jominy specimen may be compared with the cooling rates prevailing in bars of different diameters cooled in various quenching media. Such a comparison may be made by referring to the curves drawn in Figure 4.28. From this it may be seen that the rate of cooling at a Jominy distance of, say, 14 mm is the same as that prevalent at a point 2 mm below the surface of a 75 mm diameter
Hardenability

bar, or at 10 mm below the surface of a 50 mm diameter bar, or at the centre of a 38 mm diameter bar; all the bars being quenched in moderately agitated oil. Using the Jominy curves and Figure 4.28 it is therefore possible to construct a diagram showing the hardness distribution, after hardening, in the manner shown in Figure 4.29.

If a tempering curve for the Jominy specimen is available, a diagram showing the transverse hardness of a hardened and tempered bar may be constructed in a similar way. The form such Jominy tempering diagrams take is shown in Figure 4.30.

Figure 4.28 applies to one rate of cooling only. The diagram may be supplemented to include various rates of cooling, the values of which correspond to those given in Table 4.1. Examples of such diagrams are given in Figures 4.31 to 4.36. This type of diagram is constructed in a different way and its mode of application will therefore be explained by means of an example. Suppose we wish to know for a certain grade of steel the transverse hardness in a bar, say 100 mm in diameter, that has been quenched in well-agitated oil ($H = 0.5$). The first step is to consult Figure 4.21. From this we obtain a Jominy distance of 12 mm, which refers to the hardness at the surface of the bar. From the curves in Figure 4.22r we obtain a Jominy distance of 17 mm which refers to a zone situated $0.9 \times R$, i.e. $0.9 \times 50 = 45$ mm from the centre, i.e. 5 mm below the surface. In a similar manner we
consult the other diagrams to obtain the correlated Jominy distances and hardness zones. The actual hardness values are then obtained from the Jominy curve of the steel grade concerned. A diagram may now be drawn showing the hardness as a function of the distance below the surface. More examples of the application of the diagrams will be given in greater detail later in this chapter.

**Figure 4.29.** Transverse hardness in a 75 mm diameter test-piece, steel BS 708A42 (Bofors R.O. 952). Constructed from Jominy curves and Figure 4.28.

**Figure 4.30.** Jominy curves for steel D19 3CrNi. Corresponding to BS 708A42 (Bofors R.O. 952), after tempering at different temperatures.

**Figure 4.31** and 4.32. Curves showing correlation between rates of cooling in the Jominy specimen and rates of bars cooled in various quenching media. The diagram applies to the surface of the bar and to points situated 0.9 R from the centre of the bar (after Jominy).
Figures 4.35 and 4.36. Curves showing correlation between rates of cooling in the Jominy specimen and rates of cooling in various quenching media. The diagrams apply to points situated at distances of 0.7 R and 0.5 R from the centre of the bar respectively (after Lamont).
Figure 4.28 as well as Figures 4.31 to 4.36 are valid for Jominy distances up to 50 mm only. Therefore, the diagrams can be used to assess the core hardness of bars only up to about 150 mm in diameter. Of course, the diagrams may be extended but their general limit is about 80 mm which is usually the maximum Jominy distance. To evaluate bars of larger cross-sections other methods must be used and these will be described in the next section.

The information given in Figures 4.31 to 4.36 applies to round bars only. When bars of square or rectangular cross-section are hardened these sections must be converted by estimation to equivalent round cross-sections for which diagrams are available. Such conversion diagrams are shown in Figures 4.37 and 4.38 in which due consideration has also been taken of the cooling rate. For example, a plate of thickness 125 mm and quenched in still oil is subjected to the same cooling intensity as a round bar, 180 mm in diameter.
In connexion with the compilation of an international steel grades standard a conversion diagram has been constructed, by means of which square and rectangular cross-sections may be converted to equivalent round ones. Figure 4.39. For example, a 38 mm square and a 25 x 100 mm rectangular cross-section are each equivalent to a 40 mm diameter round section; a 60 x 100 mm cross-section is equivalent to an 80 mm diameter round one.

4.4 PRACTICAL APPLICATION OF THE TTT AND THE CCT DIAGRAMS

The familiar TTT diagrams, which show the progress of the transformation at constant temperature (isothermal transformation) are useful when

![Image of a TTT diagram showing transformation temperatures and phases]

**Figure 4.40.** Continuous-cooling transformation (CCT) diagram with some cooling curves for corresponding Jominy distances. The lower graph shows the hardness values along a Jominy test specimen.

isothermal processes are concerned, such as isothermal annealing, austempering or martempering. These processes are described in the chapter on thermal treatments. However, TTT diagrams of this type are suitable only for an approximate assessment of the hardenability of the various steels. Continuous-cooling transformation diagrams (also called the CCT diagram) are so constructed as to enable the people dealing with heat treatment to predict which constituents will be produced as the rate of cooling varies. Since each

![Image of a CCT diagram with various cooling curves and hardness values]

**Figure 4.41.** CCT diagram and Jominy hardenability chart for AISI 1141 (Metal Progress, Oct. 1965, p. 134)

individual point on the Jominy specimen corresponds to a known rate of cooling, it has been possible, by using data from the Jominy test, to construct CCT diagrams on which are superimposed cooling curves corresponding to the rates of cooling prevalent at various Jominy distances. Such a diagram is shown in Figure 4.40. The rate of cooling at each point on the Jominy specimen is equivalent to some definite rate of cooling in a steel bar.
—see Figure 4.28—and hence it is possible to predict from this type of transformation diagram which structural constituents will form in the steel bar, as well as the temperature hardness. However, it is not quite correct to assume that the same cooling curves apply to all grades of steel, since different steels have different thermal conductivities.

During the 1960s a series of CCT diagrams was published in Metal Progress. Two of these diagrams are reproduced in Figures 4.41 and 4.42. Let us follow the constitutional transformations in the steel grade AISI 3140 at a point 19 mm from the quenched end-face of the specimen. In 25 s ferrite begins to form, in 30 s pearlite begins and in 45 s bainite. 50% of the austenite has been transformed in 90 s. In about 140 s from the start of the cooling of the test-piece the temperature has fallen to 315 °C and martensite formation begins.

Figure 4.42. CCT diagram and Jominy hardenability test for AISI 4340 (Metal Progress, Sep., 1964, p. 106)

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154 HARDENABILITY

Based on data obtained from dilatometric investigations, Rose and co-workers have constructed diagrams with continuous-cooling curves. The hardness obtained for each cooling rate is indicated on the appropriate cooling curve. In addition, cylindrical test-pieces have been cooled in various quenching media and the rate of cooling has been measured at points 0.5 mm below the surface and at the centre. The values so obtained were plotted in diagram form on a sheet of transparent paper which is to be laid on top of the CCT diagram. In order to assess the hardness from these diagrams, select the hardness figure marked at the end of that cooling curve (drawn on the CCT diagram) that comes closest, immediately below the M s line, to the cooling curve drawn on the transparent sheet that represents the bar dimension concerned.

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Figure 4.43. CCT diagram for steel D1N 42CrMo4 which corresponds to BS 708A42 (Befors RO 932). Cooling curves for 95 mm diameter bar are superimposed and represent cooling curves on the transparent overlay sheet in Atlas zur Wärmebehandlung der Stähle. Figures attached to the lines in the diagram indicate percentage amount of structural constituent. For example, fourth thin line from the left indicates that 2% ferrite has formed; when bainite starts and that 75% bainite has formed when martensite starts to develop.
EXAMPLE

From Figure 4.43, on which the cooling curves for a bar, 95 mm in diameter, have been superimposed, it can be seen that on quenching this bar in water, the surface will contain 2.5% bainite and have a hardness of 52.5 HRC. At the centre, 5% of the austenite will have transformed to ferrite and 75% to bainite before the start of the martensite formation. The final hardness will be 34 HRC. In the case of oil quenching it is more difficult to assess the surface hardness, which will be somewhere between 34 and 52 HRC. On air cooling the hardness will lie between 220 and 230 HB at both surface and centre.

The German Stahl-Eisen Werkstoffblatt Nos. 200–69 and 250–70 contain CCT diagrams which indicate: (a) numerical values of the time, in seconds × 10^{-3}, taken to cool between 800 and 500°C—e.g. 0.22, 2.0 and 2.5 (the figures are to be multiplied by 100 in order to give the time in seconds) and (b) the cooling rate in degrees C per minute between 800 and 500°C. One of these diagrams is reproduced in Figure 4.44. In the book Phase Transformation Kinetics and Hardenability of Medium-Carbon Alloy Steels, issued by Climax Molybdenum there are a number of CCT-diagrams, four of which are reproduced in Figures 4.45–4.48. The diagrams show the influence of the molybdenum content on the transformations of a steel having a base analysis of 0.38% C, 0.34% Si, 0.84% Mn, 1.46% Ni and 0.74% Cr.

![Figure 4.44. CCT diagram for DIN X40CrMoV51 (H13) / Stahl-Eisen-Werkstoffblatt 250–70).](image1)

![Figure 4.45. Continuous Cooling Transformation diagram for the steel containing 0.38% C, 0.33% Si, 0.85% Mn, 1.46% Ni, 0.74% Cr and 0.0% Mo, quenched at 830°C (1526°F) for 30 min (by courtesy of the Climax Molybdenum Company).](image2)

![Figure 4.46. Continuous Cooling Transformation diagram for the steel containing 0.38% C, 0.33% Si, 0.85% Mn, 1.46% Ni, 0.74% Cr and 0.0% Mo, quenched at 830°C (1526°F) for 20 min (by courtesy of the Climax Molybdenum Company).](image3)
Figure 4.47. Cooling Transformation diagram for the steel containing 0.38% C, 0.34% Si, 0.84% Mn, 1.40% Ni, 0.71% Cr and 0.48% Mo, austenitised at 1120°C (2048°F) for 30 min (by courtesy of the Climax Molybdenum Company)

Figure 4.48. Cooling Transformation diagram for the steel containing 0.38% C, 0.34% Si, 0.84% Mn, 1.40% Ni, 0.71% Cr and 0.48% Mo, austenitised at 1120°C (2048°F) for 30 min (by courtesy of the Climax Molybdenum Company)

Figure 4.49. Cooling time for surface and core for different bar dimensions on cooling in water, oil and air respectively (after Blanket and Hauzer*)
In order to calculate the depth of hardening in large cross-sections Bandel and Haumer have carried out exhaustive theoretical and empirical calculations. Russell's theoretically deduced cooling curves have been compared with the results from a number of experiments with thermocouples placed inside steel bars, the diameters of which ranged from 250 mm to 950 mm. From the data obtained they have drawn amended curves pertaining to cooling in water, oil and air. Figure 4.49. The shape of the curves does not alter with austenitizing temperature provided the appropriate temperature scale is chosen. Obviously the curves must be regarded as representing mean values since it is not possible to take into consideration those aberrations and irregularities that occur during cooling. By their investigations Bandel and Haumer have been mainly concerned in finding, with the aid of cooling curves and CCT diagrams, some connexion between Grossmann's $D_i$ values and the diameter of the bar that, on oil cooling, become ferrite-free at the centre. The calculations apply to heat-treatable steels and have resulted in the curves shown in Figure 4.50 in which the decisive influence of grain size can be seen. A certain heat-treatable steel, the $D_i$ and grain size of which are 200 mm and 7-8 ASTM respectively, becomes ferrite-free at the centre on hardening in oil when its diameter is 350 mm. When the grain size is 3-4 ASTM the corresponding dimension is about 600 mm. Owing to the long heating times used for large forgings it is not unusual for the grains to grow to a size corresponding to 3-4 ASTM.

Figure 4.50. Relationship between ideal critical diameter $D_i$, austenite grain size and bar diameter $D$ (after Bandel and Haumer).

By making use of the diagrams in Figure 4.49 it is a fairly simple matter to estimate the hardness and structural constitution from the CCT diagrams. Below is given an example indicating how this is carried out.

Steel grade X 40 Cr-Mo-V 51 (Figure 4.44). Austenitizing temperature 1020°C.

If we choose the cooling curve marked $8.5 \times 10^{-2}$ we see that it intersects the 350°C horizontal in $2 \times 10^{-1}$ s. From Figure 4.49 we see that the dimensions of a bar giving this time-temperature relationship at the surface and centre on oil quenching are roughly 500 mm in diameter and 300 mm in diameter respectively. If the same structure is required on air cooling the bar diameter

![CCT-diagram according to Atkinson and Andrew. The vertical lines in the upper diagram give the cooling rates for bars with different diameters when quenching in different media. (By courtesy of the British Steel Corporation.)](image)
must be less than 100 mm. The maximum bar diameter that will result in a non-pearlite structure on air cooling is about 125 mm. The diagram in Figure 4.44 has superimposed on it the cooling curves (oil quench) for the centres of bars 50 mm, 100 mm and 200 mm in diameter, respectively.

Within the Swedish Jernkontoret's (the Swedish Ironmasters' Association) Research Organization, a committee has been studying the possibilities of evaluating from CCT diagrams the hardness values and microstructures obtained in practical hardening of steel. It has been found that the shape of the cooling curve is decisive for the structure and hardness obtained in hardening. Most CCT diagrams are worked out mainly with a 'natural cooling' almost corresponding to the cooling process for the centre of steel bars cooled in oil. There are also graphs which have been made with linear cooling. In order to improve the practical use of CCT diagrams it appears to be necessary to make separate graphs for each steel, covering cooling rates corresponding to those valid for different levels below the surface, e.g. 0.8 r, 0.5 r, and the centre, for cooling in water, oil and air. Atkins and Andrew have actively contributed to the development in this field by establishing a type of graph highly valuable for practical heat treatment. Their work should be considered as the best compromise so far presented.

4.5 PRACTICAL APPLICATION OF HARDSKINABILITY

There are two depth-of-hardening limits between which the hardenability should lie in order that the steel, after hardening, may be used to good technical and economic advantage. The lower limit, applicable mainly to tool steels, is set by that hardenability that will give a depth of hardening of about 2 mm, this hardened zone having at least a 90% martensitic structure derived from the heat treatment and the mass effect of the cross-section. (This does not apply to case-hardening steels.) The upper limit is set by the hardenability that will result in an approximately 90% martensitic structure at the centre. A steel having a higher hardenability, i.e. a steel that in a heavier section would give the same hardness result as in a smaller cross-section, would be regarded as uneconomic when used in the smaller dimension, provided that only the price factor is considered and that price is equated with hardenability. For every tool and every constructional part there is a rather narrow range of hardenability that will give optimum results, but it is usually for highly stressed parts only that such results are required. From the point of view of stock-keeping and safety it often happens that steels having a hardenability higher than necessary are used.

Which properties are concomitant with high hardenability and which with low hardenability, respectively?

4.5.1 High hardenability

Where steels for hardening and tempering in large cross-sections are concerned, the benefits of high hardenability are obvious. Such steels can be used in heavy dimensions for parts that are later to be machined to various cross-sectional areas and for which strength requirements are equally high in the smaller and in the heavier sections. Unfortunately the test methods as stipulated by both SIS and ISO are concerned only with a zone extending to about 20 mm below the surface.

Even if the hardness is the same at the surface as at the centre in a hardened and tempered bar it must be accepted generally that the impact strength of the central regions of a bar are usually lower than at the surface. This is because the surface zones, as a rule, consist of tempered martensite or lower bainite whereas at the centre the structure is composed mainly of upper bainite or pearlite. These two latter constituents have considerably lower impact strength and higher transition temperature ranges than tempered martensite or lower bainite. As an example of this, see Figure 4.52 which shows

Figure 4.52. The influence of the microstructure on the impact strength of hardened and tempered steel S1 S 2225 (Bofors RO 653) with 0.29% Cr, 1% Mo and 0.2% V. (a) Structure composed mainly of upper bainite. Grain size 5 ASTM hardness 285 BHN, impact value Charpy V 16 kpm. (b) Structure composed mainly of lower bainite. Grain size 5 ASTM hardness 285 BHN, impact value Charpy V 6-8 kpm

the microstructure and impact values of two different bars, steel S1 S 2225 (Bofors RO 653), the composition and hardness of which were in close agreement. In spite of this, the steel (a) which contained upper bainite, failed on welding. In contrast to this, the steel, the structure of which was mainly lower bainite (b), gave excellent service on being welded. In the first case the rate of cooling from the hardening temperature had not been rapid enough. When service requirements really do call for high impact values, not only should the composition of the steel be considered but also the hardening process.

Another instance when high hardenability is required, concerns hardened and tempered forged blooms for drop-forging dies or pressure diecasting dies. These forgings are engraved subsequent to the heat treatment. Tools subjected to high pressures must also be made from steel that hardens through entirely, otherwise the softer core might be deformed. In such cases the surface
layers will be exposed to excessive stresses which are liable to cause this layer to rupture.

4.5.2 Low hardenability

In many cases a low hardenability is to be preferred, both for machine parts and for tools. A shallow-hardening steel, such as W 1, can give a remarkably good performance. Long experience tells us that such steels can stand up to impact stresses extremely well and it is said that they can do so because of the presence of a hard case and a tough core. This point will be dealt with further on in this chapter.

Tools with varying cross-sectional areas often need to have an even depth of hardness on all working faces. For such uses steel S1S 2092 (Bofors SR 1855) has shown itself to be eminently serviceable under working conditions. This has been subsequently verified by metallographic investigations and practical heat treatments. These investigations, which are of interest since they tend to verify the deductions drawn from data obtained from Jominy diagrams, are described below.

The steels used for cold pilger rolls are S1S 2092 (Bofors SR 1855), AISI O 1 (Bofors RT 1733) and Bofors grade RO 169. This last steel is used also for cold-reduction rolls and large ball-bearings. However, since this steel is not always entirely satisfactory for, say, ring rolls, the two first-mentioned steels are more widely accepted. In order to harden a pilger roll made from O 1, outer diameter 340 mm, cross-sectional area 145 x 82 mm², it was quenched in oil from 830 °C. The hardness at the corners was 55 HRC.

![Figure 4.53. Fractured surfaces of ring roll made in steel O 1 (Bofors RT 1733), oil hardened from 830 °C. The dark areas on the top part of the photograph are round notches to initiate the fracturing of the roll.]

Figure 4.53. Fractured surfaces of ring roll made in steel O 1 (Bofors RT 1733), oil hardened from 830 °C. The dark areas on the top part of the photograph are round notches to initiate the fracturing of the roll.

Figure 4.54. Fractured surfaces of ring roll made in steel O 1 (Bofors RT 1733), oil hardened from 860 °C.

164 HARDENABILITY

but on the cylindrical surfaces and in the groove the hardness was only 45 HRC. The appearance of the fractured surfaces is shown in Figure 4.53. Since it was suspected that the hardening temperature used for this rather large roll was too low, a temperature of 860 °C was tried next. A hardness of 62 HRC on most surfaces was the result. As may be seen in Figure 4.54 the roll had hardened throughout the section, shown in the photograph on the right, whereas in the other section (see photograph on the left) the result of the hardening was unsatisfactory both in the central area of the roll and at one surface.

Yet another ring roll was hardened, a temperature of 890 °C being used this time. Full hardening of 62 HRC was realized all over the roll but it cracked in service after a short time. Metallographic examination revealed that the roll had a very coarse fracture, viz. fracture No. 5 according to the J K scale, and it had hardened throughout the whole section. It was apparent that to find the ideal hardening temperature for this tool was quite a problem.

Examination of a similar roll made from steel S1S 2092 and hardened from 860 °C revealed the fracture as shown in Figure 4.55. The hardened zone has approximately the same depth all the way round. Rolls made in this steel give satisfactory service.

In order to obtain a proper insight into the behaviour of these various steels a systematic investigation was initiated which involved the hardening of Jominy test-pieces from different hardening temperatures. The results are shown in Figures 4.56 to 4.58. From these curves we see that steel S1S 2092 (Bofors SR 1855) has the lowest hardenability and that the depth of
Figure 4.55. Fractured surfaces of ring roll made in steel S15 2092 (Befors SR 1855), oil hardened from 866°C.

Figure 4.56. Steel S15 2092 (Befors SR 1855). Jominy curves and ASTM grain size obtained with different hardening temperatures.

Figure 4.57. Steel O1 (Befors RT 1733). Jominy curves and ASTM grain size obtained with different hardening temperatures.

Figure 4.58. Befors RO 169 (steel for rolls for cold rolling). Jominy curves and ASTM grain size obtained with different hardening temperatures.
HARDESTABILITY 167

hardening and grain size are least influenced by the hardening temperature.

By referring back to the curves in Figures 4.31 to 4.36 it is possible to con-
struct depth-of-hardening diagrams for different rates of cooling. The different
severities of quench were as follows:

- $H = 0.5$ — oil quench, good agitation,
- $H = 0.7$ — oil quench, strong agitation,
- $H = 1.0$ — oil quench, violent agitation.

The calculations were carried out for a bar diameter of 120 mm, which is
roughly equivalent to the cross-sectional area of the ring rolls. Figures 4.59
to 4.61 show the results. Since the depths of hardening down to 60 HR C
are of particular interest these values were read off from the curves in the
above-mentioned figures (see Table 4.2). From this table it is clear that in
the dimensions concerned, steel S15 2092 obtains the most uniform hardness
penetration, 6–9 mm, from a hardening temperature of 890 °C. We see that
on quenching from this temperature the steel is independent, by and large,
of cooling rates lying between the limits $H = 0.5$ and $H = 1.0$. By increasing
the cooling rates to values lying in the range, say, $H = 0.7$ to $H = 1.0$ a

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<tr>
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hardening temperature between 850 and 890 °C may be used. Should 890 °C
not be high enough the temperature may be increased by some 20° without
causing undue grain growth in the steel, as may be seen in Figures 4.56
and 4.62.

As regards the other two steels, the intermediate hardening temperatures
of 850 °C or 860 °C can be used but any change in the depth of hardening
is largely dependent on the rate of cooling. It is not advisable to try the higher
temperatures if the steels are to remain fine-grained. For sections exceeding
120 mm in diameter and for which increasing depths of hardening are
required the hardenability of steel SIS 2092 is not high enough to render this steel suitable for oil quenching.

With regard to the three steels under investigation only steel SIS 2092 is capable of being what is known as contour hardened. This concept will be dealt with further in Chapter 6.

4.6 THE INFLUENCE OF THE DEPTH OF HARDENING ON THE STRESS PATTERN

In the preceding section we have discussed the advantages attendant upon a favourable depth of hardening. In addition, an instance has been quoted illustrating the benefits accruing to a steel possessing a limited hardenability. However, this is only one aspect of the matter. The residual stresses that are created as a consequence of the hardening operation are of such importance that they must be taken into consideration when a steel and its heat treatment are being selected.

As a result of the hardening operation, thermal and transformation stresses are set up in the steel. If we regard the thermal stresses only and reckon them as tangential stresses, we obtain a stress distribution as shown in Figure 4.6.5, which applies to a bar 100 mm in diameter and quenched in water from 850 °C. At the point of time T the temperature difference between the surface and the centre is greatest, i.e. some 500 °C. This corresponds to a linear differential of about 0.6%, and gives rise to a stress that exceeds 100 kN/mm². If the body of the material were elastic enough to accommodate this stress, the stress diagram for the surface layer would take the shape of curve a. After temperature equalization between the surface and the centre the stress should fall to zero. Since the yield point of steel is considerably lower at elevated temperatures than at room temperature the material will flow plastically, resulting in a surface stress diagram, such as curve b.
At the centre the stresses will be such as indicated by curve c. When time $T$ has elapsed, contraction in the core will exceed that at the surface. This implies that at time $T$ there is complete stress neutralization and consequently at room temperature there will be compressive stresses in the surface and tensile stresses at the centre.

![Figure 4.63](image)

*Figure 4.63. Development of thermal stresses on cooling. Transformation stresses not taken into account (after Roes).*

When the bar has cooled to room temperature, the stress distribution will be as shown in the right-hand figure, i.e. tensile stresses at the centre and compressive stresses in the surface. The more effective the quenching medium and the larger the dimensions of the bar, the greater will be the stresses.

The phenomena involved when steel hardens produce additional stresses due to the increase in volume that takes place as austenite transforms to other structures. The following increments of volume and linear expansion, Table 4.3, apply to a 1% plain carbon steel. If the main part of the transformation takes place in the centre before the surface starts to transform, which is the course of events in shallow-hardening steels, the same stress pattern as has been described in Figure 4.63 is set up. If the transformation to, say, pearlite has proceeded to 100% completion at the centre before martensite has started to form in the surface, the compressive stresses increase considerably when martensite formation starts.

<table>
<thead>
<tr>
<th>Change in volume</th>
<th>Change in length</th>
</tr>
</thead>
<tbody>
<tr>
<td>%</td>
<td>%</td>
</tr>
<tr>
<td>Austenite to martensite</td>
<td>+4.3</td>
</tr>
<tr>
<td>Austenite to lower bainite</td>
<td>+3.2</td>
</tr>
<tr>
<td>Austenite to pearlite</td>
<td>+2.4</td>
</tr>
</tbody>
</table>

![Figure 4.64](image)

*Figure 4.64. Stress pattern for cooling, transformation and residual stresses in round test bars of different dimensions in steel DIN 22CrMo4 (after Roes).*

When the main part of the transformation starts in the surface, tensile stresses are produced there and compressive stresses at the centre. In steels that become martensitic throughout the entire cross-section, the stress difference increases as the dimension increases. Rose and Bühler have carried out measurements on different specimens. One such series of measurements, shown in Figure 4.64, deals with the quenching in water of bars of various dimensions. We can see how the dimension is instrumental in shaping the stress pattern.

The stress distribution in the 100 mm bar corresponds roughly to that due solely to thermal stresses. If the nose of the pearlite curve had been located farther to the left an additional increment of stress would have resulted.

With regard to the 30 mm diameter bar the start of the bainite formation at the centre coincides approximately with that of the martensite in the surface.
Compressive stresses were measured both in the surface and at the centre and tensile stresses in the intervening zones.

The 10 mm diameter bar is hardened through completely. The temperature difference between the surface and the centre is smaller than that obtaining in larger-diameter bars. Furthermore, the main part of the martensite formation in the centre takes place when the surface-zone martensite reaction is nearly complete. Both of these factors contribute to the creation of tensile stresses in the surface and compressive stresses at the centre.

The development of similar stress patterns may be envisaged if the alloy content of the steel is increased or if the hardening temperature is raised, the dimensions of the bars being kept constant (see Figure 4.65). There are endless variations possible along these lines. Let us limit our studies to two cases only.

1. Hardening of a plain-carbon steel so that the result will be compressive stresses in the surface and tensile stresses at the centre.
2. Hardening of an alloy steel so that the result will be tensile stresses in the surface and compressive stresses at the centre.

In the first case the steel hardens at the surface and in the other case it hardens through entirely. The stress distribution is shown in Figure 4.66.

We shall discuss these cases with regard to the incidence of crack initiation on bending, assuming that the strength of the surface layer is the same for both bars and the hardening depth is at least 10% of the bar diameter. On subjecting the bars to bending forces, as illustrated in Figure 4.66 tensile stresses are induced in the lower layers and compressive stresses in the upper.
ones. A crack is more easily initiated in the layer subjected to tensile stresses. In order to balance the compressive stresses in the underside of the surface-hardened bar a definite bending force must be applied. This force, if applied to the through-hardened bar, will simply cause the tensile stresses on the underside to increase. On increasing this bending force additional tensile stresses are induced in the underside of this bar which will therefore be the first one to fracture, always assuming that there is an ample depth of hardening in the surface-hardened bar. What has been said above holds in a higher degree for fatigue and impact stresses.

In practice it has often been observed that surface hardening is an excellent means of increasing the endurance limit, whether by making use of shallow-hardened steels or by special methods of surface hardening, such as induction hardening or case hardening. Similarly, a long service life has been given by surface-hardened steels when used as tools subjected to impact stresses. A necessary condition is, however, that all working faces have been hardened.

The example of the pilger rolls discussed in the preceding section constitutes a typical case of a tool that has had compression stresses induced into its surface layers. As a consequence it has gained enhanced resistance to fracture which would otherwise be caused by the tensile stresses developing in the surface during the rolling work. Oil quenching can produce tensile stresses in the surface of a through-hardening steel. These stresses may be reduced by resorting to martempering instead.

To sum up, it may be said that whenever practicable from the point of view of the heat-treatment process, surface-hardened or alternatively not completely through-hardened tools and machine parts are to be preferred, due consideration of course being given to the service that the parts are subsequently to be subjected to.

REFERENCES

6. Lamont, J. L., 'How to Estimate Hardening Depth in Bars', Iron Age, 152, 64–70 (October 1943)
5
Heat treatment—general

The preceding chapters have dealt with the theoretical background of the heat treatment of steel. The following chapters are based mainly on practical experience and applications. This general section describes the standard heat-treatment techniques on which are based the special methods described in the next chapter.

5.1 ANNEALING

The general purpose of an annealing treatment is to reduce the hardness of steel or to produce a structure that facilitates the progress of subsequent manufacturing operations. The term annealing by itself usually implies full annealing, which in most cases involves solution treatment in the austenite phase followed by slow cooling. For other types of annealing prefixes that indicate the nature of the process are generally used.

5.1.1 Annealing for maximum softness or Spheroidizing Anneal

The theory of this apparently simple heat-treatment operation is still rather obscure but on the basis of practical experience and on data derived from isothermal (TTT) and continuous cooling (CCT) transformation diagrams, almost ideal annealing practices have been developed. The classic spheroidizing annealing temperatures for carbon steels are shown in Figure 5.1, where it can be seen that for hypo-eutectoid steels annealing below A1 is suggested. However, annealing above A1 followed by transformation just below A1 is to be recommended in many instances as outlined below.

When a steel with a pearlite structure is annealed the cementite lamellae after a short time assume the fanciful shape of a dog’s leg. On continuing the annealing the lamellae form globules at their ends and split up into spheroids; hence the name spheroidizing. As the temperature is increased the process speeds up. Figure 5.2 is typical of the fully spheroidized structure of a tool steel. If the annealing treatment is carried out below A1 the rate of cooling does not influence the final hardness at room temperature. If
annealed above \( A_1 \) and followed by slow cooling, the spheroidized structure is still largely maintained provided that the temperature was only a few tens of degrees above \( A_1 \). As the steel cools, the carbon dissolved in the austenite will separate out on the carbide spheroids. If the annealing temperature is higher, a greater amount of carbide will dissolve and the cementite will separate out as lamellae. The actual temperature below \( A_1 \) at which the transformation takes place has an important bearing on the appearance of the structure obtained. The closer the temperature of transformation lies to \( A_1 \), the coarser and softer will be the spheroidized structure. The farther away below \( A_1 \) the transformation takes place the finer, more lamellar and harder will be the structure (see Figure 5.3).

The major rule to apply when undertaking a spheroidizing anneal of plain-carbon and low-alloy steels may be summarized in the following lines:

Austenitize at a temperature not more than 50 °C above \( A_1 \), and transform at a temperature not more that 50 °C below \( A_1 \).

For example: Steel S15 2244 (BS 708A42) the \( A_1 \) temperature of which is about 750 °C, is annealed, according to the above rule, as follows:

1. Partial austenitization at 775 °C (2 h).
2. Cooling for 5 h down to 725 °C (10 °C/h).
3. Continued cooling to room temperature.

In Figure 5.4 the cooling curve is traced in the TTT diagram (curve 1). Curve 2 traces the cooling process when carrying out an isothermal anneal, as discussed in Section 5.1.4. Once the austenite has transformed, the cooling structure may be obtained, as mentioned above, by increasing the annealing temperature. This is clearly illustrated in the series of micrographs, Figure 5.5.

Hyper-eutectoid steels must be annealed above \( A_1 \) in order to spheroidize the grain-boundary cementite within a reasonable time. Even so, to obtain acceptable annealing results in a steel with a pronounced network of grain-boundary cementite may be quite a problem. A prior normalizing treatment may help to dissolve the network. After the steel has cooled from the normalizing temperature the carbides are present in a more finely dispersed form.

The spheroidization of plain-carbon and low-alloy steels containing carbon in the range 0.50-0.80% may be facilitated by a pre-annal at about 25 °C below \( A_1 \), prior to the annealing proper. The spheroidized carbides formed during the pre-annal are more resistant to dissolution at the higher annealing temperature and hence during the subsequent cooling they will promote the
formation of spheroidized rather than lamellar pearlite. For the majority of steels the annealing time is between 2 and 6 hours.

**FULL ANNEALING OF HIGH-SPEED STEEL**

With regard to high-speed and other high-alloy steels, where $A_1$ temperatures are around 800°C, the annealing temperature must be in the region of 875°C in order that the steels may obtain their lowest hardness. The grain size resulting from the hardening of high-speed steel is dependent on the temperature of the preceding annealing process. A low annealing temperature gives rise to a fine-grained structure and consequently a greater hardness, as is shown in Figure 5.6. The annealing temperature has a very definite influence on the grain size, which is also affected by other factors such as the degree of reduction during hot working. In general, the smaller the dimensions of the finished steel, the finer is the resulting grain size.

In principle, most steels can be rehardened without any intervening annealing treatment. Each successive hardening operation may be regarded as a normalizing treatment that has a grain-refining effect. However, this does not apply to high-speed steels nor to steels with high tungsten contents, e.g. hot-work steels. If such steels are rehardened without an intervening and thorough full annealing treatment they are liable to become coarse-grained. This abnormal grain growth depends on the hardening temperature. As is shown in Figure 5.7, for each hardening temperature above approximately 1040°C there corresponds, during the second hardening operation, a critical temperature above which abnormal grain growth will start. If the first hardening treatment is carried out at a low temperature, say 1150°C, then during the second hardening treatment, grain growth will already have started as the temperature approaches 1100°C. If, however, the first hardening takes place from 1290°C there is no grain growth during the second solution treatment for hardening until the temperature exceeds about 1230°C.

The reason for this anomaly has not been definitely accounted for. One possible explanation is that the dissolution of carbides (mainly $M_23C_6$),

![Figure 5.5. Influence of temperature of annealing on structure and hardness of a steel 0.40% C, 0.60% Cr (Bolens R 13), Holding time: 2 h. Rate of cooling: 15–15°C to 620°C: (a) Annealing temperature 820°C, hardness 285 HB; (b) Annealing temperature 800°C, hardness 195 HB; (c) Annealing temperature 775°C, hardness 185 HB; (d) Annealing temperature 750°C, hardness 174 HB. All 250×](image)

![Figure 5.6. High-speed steel M2. Influence of annealing temperature and time on hardness after annealing and on grain size after hardening. Diameter of test piece 18 mm.](image)
which started during the first hardening treatment, is continuing during the second heating for hardening but that now the amount of carbide remaining is too small to prevent grain growth. A form of carbide precipitation occurs around 550 °C and the higher the preceding hardening temperature has been, the more pronounced is this precipitation. Assuming that the first hardening

![Diagram of grain size vs hardening temperature](image)

Figure 5.7: Dependence of grain size on hardening temperature when double-hardening high-speed steel, type 1 from Cohn and Kato.

temperature has been relatively high carbides will begin to be precipitated during the heating-up for the second hardening as a temperature of 550 °C is approached. This precipitation might constitute an impediment to grain growth and as a consequence the critical temperature should be displaced upwards. If an annealing treatment is introduced at this stage it should cause the carbides to be precipitated more completely, thus impeding grain growth during subsequent hardening treatments. The phenomenon described above, however, does not manifest itself in other medium- or high-alloy steels such as H13 or D6.

Annealing for optimum machinability

The above heading appears in certain specifications. Apparently the implication is that the steelmaker is to carry out some annealing treatment that, along with giving the steel its best machinability is also the most economic one. The specification may imply, too, that there exists an understanding between the steelmaker and the consumer to the effect that a certain structure or a certain annealing treatment has been agreed upon.

An example of the above is the case-hardening steel S1S 2511–98 (En 35), for which the maximum hardness has been set at 217 HB. Should this hardness be exceeded after rolling, an annealing treatment at 650 °C only

is normally carried out. If the annealing temperature is higher the hardness may become so low that the steel will be 'sticky' in certain machining operations. Similar steels with somewhat higher carbon contents possess the best machinability, e.g. in gear and spline shaping when they have been isothermally annealed (see later sections).

The phrase 'annealing for improved machinability' is also used. This treatment is applied to low-alloy steels that are to be subjected to forging operations after rolling, and usually consists of an anneal at 650 °C. In such cases the maximum hardness must be stipulated, often 250 HB, in order that the steel may be cold sheared or sawn. Several grades of low-alloy structural steels possess optimum machinability when the pearlitic constituent consists of roughly equal parts of lamellar and spheroidized forms. For alloy steels it is a maxim that the lower their hardness, the better is their machinability. The structure should consist of completely spheroidized carbides in a ferritic matrix. The presence of grain-boundary bands of carbide globules or traces of lamellar pearlite drastically impairs machinability.

5.1.2 Recrystallization annealing

When steel is subjected to cold work its hardness increases and its ability to endure continued cold work diminishes. By annealing the steel above 600 °C recrystallization occurs, i.e. new, stress-free grains are formed which grow at the expense of the deformed original grains. A holding time of 1–2 h at the recrystallization temperature is normal. If continuous annealing is used the time may be shortened considerably.

The grain size after the recrystallization anneal depends on the original grain size. The amount of cold work before recrystallization has a decisive influence on the grain size of plain-carbon steels containing carbon less than 0.2%. On subjecting them to the so-called critical reduction of area, i.e. 6–15%, reduction, the steels become coarse grained on being subsequently annealed.

Annealing for recrystallization is most commonly applied to cold-rolled, low-carbon sheet or strip steel. Austenitic steels, such as the 18/8 stainless and the 13% manganese (Hadfield) steels, also recrystallize on being annealed after cold work (see Section 5.5.1, Quench annealing).

5.1.3 Stress-relief annealing

When steel is machined or deformed plastically, stresses are induced in to the cold-worked surfaces. These stresses may give rise to increased hardness which is apt to render continued working of the steel increasingly difficult. In addition they may cause the steel to distort during the subsequent heat-treating operation and should therefore be reduced or eliminated, which can be done by a 1–2 h stress-relief anneal. For plain-carbon and low-alloy steels a temperature of 550–650 °C is required, for hot-work and high-speed steels, 600–750 °C. This treatment will not cause any phase changes, but
recrystallization may take place. In order that thermal stresses are not induced during cooling, it is good practice to allow the parts to cool slowly in the furnace to approximately 500°C after which they may be taken out and allowed to cool freely in air.

When very large tools or machine components requiring the greatest possible freedom from residual stresses are being stress-relieved, the cooling rate at the start must be very slow, i.e. a few degrees C only per hour. As the temperature gets lower the cooling rate may be increased but not until a temperature of about 300°C is reached is it advisable to cool freely in air. The reason for recommending an initial slow cooling when the steel is at its highest temperature is that the yield point is low and there is a danger that it might be exceeded by the stresses induced if too great a temperature difference exists between the surface and centre. If the thermal stresses cause permanent deformation, new stresses will remain after cooling to room temperature.

Hardened and tempered steel may be stress relieved at a temperature about 25°C below that used for tempering. Tools and machine components that are to be heat treated should be left with a machining allowance sufficient to compensate for any warping resulting from stress relieving. Forgings should also be stress relieved if internal stresses are induced in the steel on account of rapid or uneven cooling, or if the forging has been subjected to severe straightening. In some instances a stress-relief treatment after normalizing is stipulated, for example for castings and weldments, specially those of alloy steel.

If the greatest possible freedom from internal stresses is required the annealing treatment should be carried out near the upper limit of the appropriate temperature range indicated above. Unfortunately this might not always be practicable since a high temperature may give rise to an unacceptably large amount of superficial oxidation or to too much softening in a hardened and tempered steel. In such cases it is necessary to choose a lower stress-relief temperature but which is as high as is compatible with a scale-free surface. However, a certain amount of residual stress must then be accepted.

Rosenstein\(^2\) has suggested a method of assessing the effectiveness of the stress-relief treatment. The method has been tested on 13 steels having yield points ranging from 25 to 125 kpsi at temperatures between 482 and 593°C and for times up to 24 h. The specimens were inserted in a creep-testing machine and after being heated to the testing temperature they were loaded, to their approximate yield points at the temperature concerned. As stress relaxation proceeded progressively at the testing temperature the machine automatically reduced the applied stress so that the gauge length remained constant. The type of diagrams obtained is shown in Figure 5.8. Steel HY-100 has the following composition:

<table>
<thead>
<tr>
<th>C</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.18</td>
<td>1.65</td>
<td>2.91</td>
<td>0.42</td>
</tr>
</tbody>
</table>

The specimens were hardened and tempered at 620°C to give a yield strength of 74 kpsi. During the first hour, according to Figure 5.8, stress relaxation occurs very rapidly but after that it slows down considerably.

Figure 5.8. Stress as a function of time after stress relieving at various temperatures (after Rosenstein\(^4\))

In Chapter 1, p. 22 it was shown how the relation between hardness and temperature and time of tempering may be represented by means of an expression derived by Hollomon and Jaffe. Rosenstein has suggested that this expression (which is also known as the Larsson-Miller parameter) can be applied equally well to stress relaxation when annealing for stress relief. Rosenstein's original diagram which contains values for all the steels tested.
is shown in Figure 5.9. In Figure 5.10 the diagram has been reconstructed and has been incorporated with a new type of tempering diagram showing the interdependence of time and temperature, since this type of diagram is easier to interpret. For further information see p. 242.

![Diagram of Stress as a function of time and temperature when stress relieving]

According to the mean curve in the diagram it can be seen that an anneal of 1 h at 450°C results in about 50% stress relief only. For complete stress relief it would be necessary to anneal at, say, 650°C for 1 h or at 600°C for 15 h. Rosenstein's results agree reasonably well with stress measurements carried out at Bofors on steel before and after annealing for stress relief.

When carrying out stress relief or tempering subsequent to a hardening treatment, a considerable stress reduction is obtained by heating to some lower temperature, provided that there is a decomposition of martensite. When a hardened and tempered steel is to be stress relieved the stress-relief diagram may be combined with the tempering diagram for the steel concerned, since both diagrams are functions of time and temperature. In this way it is possible to optimize the time and temperature for stress-relief annealing without jeopardizing the hardness.

5.1.4 Isothermal annealing

This treatment involves heating the steel at two different temperatures: first an austenitization treatment at a suitable temperature above A1 and then a rapid lowering to the transformation temperature below A1. When the transformation is complete the steel may cool freely to room temperature. An isothermal annealing curve is drawn in Figure 5.4. For practical reasons it is not possible to carry out such a heat treatment in a conventional muffle furnace since its large heat capacity does not permit the required rapid fall in temperature. Instead, use is made of two muffle furnaces or of continuous furnaces with different temperature zones in each of which comparatively small portions only of the charge are present at the one time. By resorting to isothermal annealing, the time of treatment can be shortened considerably in comparison with the time taken for conventional full annealing. However, the resulting hardness is higher.

Alloy case-hardening steels such as BS 637M17 (En 352) are often subjected to an isothermal anneal. The austenitization is carried out at 930–940°C, i.e. at a temperature somewhat higher than that subsequently used for the carburization. This treatment is intended to reduce any distortion that may occur during the actual carburization treatment. The transformation is made to start at 610–680°C and takes 2–4 h to completion. The structure obtained consists of ferrite and pearlite which is entirely suitable for most machining operations.

Isothermal annealing is also resorted to during the various stages of steel processing. If an ingot or rolled billet of alloy (air-hardening) steel is allowed to cool freely to room temperature, cracks are very likely to appear if the core transforms to martensite. Should this happen, tensile stresses are set up in the surface layers which by that time have already been transformed. The warm ingots or billets are therefore placed in isothermal annealing furnaces at about 700°C, this being the temperature at which transformation to pearlite takes place. When this is complete the steel may cool freely to room temperature.

5.1.5 Quench annealing

This treatment is used for austenitic steels only and consists in fact of a homogenizing anneal or, in certain cases a recrystallization anneal. Slow cooling or heating will cause carbidic precipitation in austenitic steels. This precipitation generally takes place at the grain boundaries. For stainless steels (type 18/8) this implies a reduction in the chromium content in the proximity of these grain boundaries. Hence the corrosion resistance there
will diminish; thereby facilitating the incidence of what is called intergranular corrosion. This carbide precipitation can also take place by heating the steel in the temperature range 500–800 °C. The austenitic manganese steel, which is generally used for castings, is subject to intense carbide precipitation on being allowed to cool in the mould from the casting temperature. This gives the steel low impact strength.

The above-mentioned ill effects are eliminated by solution heat treatment and quenching, i.e. quench annealing. This consists of heating the steel to a temperature at which the austenite becomes homogeneous, usually around 1000–1100 °C, followed by rapid cooling (quenching), in water for instance. For light sections, air cooling will suffice to give the steel a satisfactory structure.

The stages of carbide dissolution and homogenization of the structure for a manganese steel containing 18% Mn are shown in Figures 5.11a–d.

The dependence of the hardness of this steel on the quenching temperature is shown in Figure 5.12.

Neither on heating nor cooling does the matrix of austenitic steels undergo any phase transformation that would result in grain refinement. Solution treatment may instead give rise to some grain coarsening, but this has generally no detrimental effect on the toughness of the steel. Solution

Figure 5.12. Hardness as a function of quenching temperature for a Hadfield steel (Bofors N N 4 R).

The treatment of austenitic manganese steel causes its toughness to increase considerably. This improvement is more noticeable in heavy dimensions in which the carbide precipitation is more pronounced on account of the slow cooling. Even in light sections the improvement can be quite apparent, as is shown in Figure 5.13.

Figure 5.13. Bend test on a manganese sheet strip 6 mm in thickness. Left: as rolled, followed by slow cooling. Right: as quench annealed.

On quench annealing cold-worked austenitic steels it is found that they recrystallize, which gives them their lowest hardness. If they have been subjected to a critical strain level, an abnormally large grain growth results.
5.1.6 Homogenizing annealing

When the steel, after teeming, has solidified, its structure is inhomogeneous. The subsequent heating and hot working equalizes the structure to a large extent. This process of homogenization can be speeded up by a so-called diffusion or homogenizing anneal which takes place at a high temperature, approximately 1100 °C, and for a rather long time in order to produce the desired result. Section 1.5 discusses an instance of diffusion annealing, the purpose of which was to even out concentration differences of manganese. The high temperatures and long treatment times entail very high costs and this makes the treatment applicable only to very special cases.

5.1.7 Hydrogen annealing

The hydrogen dissolved in the molten steel may give rise to internal cracks, the so-called 'flakes', in the steel after it has cooled from the hot-working operation. By subjecting the steel to an annealing treatment at 600-650 °C for several days, followed by slow cooling, the hydrogen content can be reduced and the danger of flakes formation removed. Hydrogen annealing was discussed in more detail in Section 5.2.1.

5.1.8 Hydrogen expulsion

While being pickled in non-oxidizing acids, or being cathodically cleaned or plated, steel is liable to pick up hydrogen. This hydrogen may cause the steel to fail under a stress considerably lower than that normal for the steel in question. This phenomenon is called hydrogen embrittlement. After ordinary pickling a large amount of the hydrogen diffuses out of the steel at room temperature but after cathodic electroplating with Cr, Zn, Cd, Ni, Sn, Pb, and Ag the hydrogen leaves much less readily. The current practice, which is to heat to 170 °C for 15 h or to 200 °C for 5 h, causes the hydrogen to be driven off. In some instances a higher temperature, e.g. up to 300 °C, is required, but then the treatment must take place in a controlled atmosphere to protect the plating from oxidation. At a steel hardness of around 350 HB the danger of fracture from hydrogen embrittlement is already apparent; at a hardness of more than 400 HB hydrogen removal is recommended; at hardnesses around 60 HRC hydrogen must be expelled without delay, directly following the surface treatment, otherwise the steel may crack in a few hours' time.

5.2 NORMALIZING

This treatment is also called normalizing anneal. It consists of heating the steel to a temperature corresponding roughly to its hardening temperature, holding it there for some 10-20 min and then allowing it to cool freely in air.

Figure 5.14. Influence of normalizing anneal on grain size. Carbon steel 0.30% C: (a) As rolled or forged. Grain size 3 ASTM; (b) Normalized. Grain size 6 ASTM. Both 300 ×
HEAT TREATMENT—GENERAL 193

The normalizing treatment refines the grain of a steel that has become coarsely-grained as a result of its having been heated to a high temperature, for example for forging or welding. In Figure 5.14 is shown the grain-refining effect produced by normalizing a steel containing 0.50% C.

Normalizing is applied mainly to plain-carbon and low-alloy steels. The hardness resulting from this treatment depends on the dimensions and composition of the steel. The difference in the rate of cooling between the surface and centre during normalizing is but slight. Figure 5.15 shows cooling curves for various dimensions at different depths below the surface. See also Figure 4.50. In order to calculate the tensile strength of normalized plain-carbon and low-alloy steels, use can be made of several current expressions for this purpose but these do not take any account of the dimensions of the steel. The expression used at Bofors is as follows:

\[ C_p = C \left[ 1 + 0.5 \times (C - 0.20) \right] + Si \times 0.15 + Mn \times 0.125 + 0.25 \times (C - 0.20) + P \times 1.25 - 0.5 \times (C - 0.20) + Cr \times 0.2 + Ni \times 0.1 \]

\( C_p \) is the sum of the carbon potentials.

194 HEAT TREATMENT—GENERAL

The ultimate tensile strength in kp/mm² after normalizing is:

- \( 27 + 56 \times C_p \) for hot-rolled steel
- \( 27 + 50 \times C_p \) for forged steel
- \( 27 + 48 \times C_p \) for cast steel

In some instances a double normalizing treatment is stipulated. This treatment consists of first heating the steel to a temperature some 50–100°C above the normal temperature, whereby a more complete dissolution of the constituents is obtained. The second normalizing anneal is carried out near the lower limit of the temperature range and is solely aimed at giving the steel a fine-grained structure. This double treatment is sometimes stipulated for steels to be used under subzero conditions. Generally speaking, a fine-grained steel has a lower transition temperature range than a coarse-grained one. Plain-carbon or low-alloy steel castings should always be normalized.

5.3 HARDENING

The theoretical background of the process of hardening has been discussed in detail in Section 1.3. The present section will deal mainly with the practical aspects of hardening. The structure of a steel as a result of hardening may be regarded as a starting structure which is capable of modification by subsequent tempering. The result of the hardening depends on several factors, each of which may be of decisive importance. Hence it is of the utmost importance that each of the various steps of the hardening process is given the same care and consideration.

5.3.1 Heating media

During the heating-up stage it is necessary to provide against any unintentional carburization or decarburization. Should this happen the superficial hardness measured after hardening will be misleading, which may result in the choice of an incorrect tempering temperature. Variations in the carbon content of the surface layers may cause the steel to crack during hardening. If the surface oxidation is too severe it is possible that the tolerances set for the tool might not be kept after descaling and grinding. To protect the surface of the tool the necessary precautions to be observed depend on the type of furnace used.

SALT-BATH FURNACES

These offer good protection against variations in the superficial carbon content, owing to the short heating time required and the neutral character of the bath. The bath must of course be maintained in good condition. As a rule, manufacturers of heat-treatment salts do not specify the composition of the salts. On the other hand they give detailed information on their properties and uses. For purposes of information, however, some typical
compositions and working temperatures for neutral salt baths are given below:

- 45% NaCl + 55% KCl 675-900 °C
- 20% NaCl + 80% BaCl₂ 675-1060 °C
- 100% BaCl₂ 1025-1325 °C

During use the bath takes up iron from the material being heat treated. This iron is oxidized since the bath is in contact with the atmosphere. The iron oxide thus formed has a decarburizing action on the steel charge and hence the bath must be regenerated. Follow the recommendations given by the salt manufacturers! Barium chloride baths with certain additions are used as heating media for high-speed steel and certain other high-alloy steels. Such baths may be regenerated by means of a few pieces of silica brick, which combines with the iron dissolved in the bath. The sludge thus formed must be removed regularly. By means of a graphite rod immersed into the salt bath it is easy to check whether iron is present. Any iron oxide in the bath will be reduced by the graphitic carbon and small bright beads of metallic iron will form on the rod. The carburizing or decarburizing action of a salt bath may be tested by means of steel foil which is dipped into the bath for a few minutes and then quenched. If the foil is soft after quenching, the bath is decarburizing; if more brittle than it normally should be the bath is carburizing. Foil of varying carbon contents is available.

GAS-OR OIL-FIRED MUFFLE FURNACES

These allow the amount of air of combustion to be controlled. Hence the atmosphere inside the muffle can be adjusted so that for the treatment temperature and grade of steel concerned the superficial carbon content of the charge will not be appreciably affected.

ELECTRICALLY-HEATED MUFFLE FURNACES

These, if not operated with controlled atmospheres, should be used with annealing boxes in which the steel charge is packed with some protective material, which should be as neutral as possible. Unfortunately there is no general-purpose substance so versatile that it is suitable as a packing material for all grades of steel since these materials are only neutral towards steels of a definite carbon content and at a definite temperature. One purpose of the packing material is to prevent the ingress of air to the steel. For this purpose cast-iron chips are very suitable since the air that is sucked into the box on account of temperature variations preferentially oxidizes the chips. Practical tests and scientific investigations have shown that cast-iron chips are practically inert towards plain-carbon steel and low-alloy steel containing carbon between 0.6 and 1%. They are also inert towards high-alloy chrome steels of the type D 2 and D 6, provided the customary hardening temperatures are employed. With regard to hot-work tool steel containing carbon around...
0.3% cast-iron chips have a slightly carburizing action. At about 1000°C there is already a danger that cast-iron chips may begin to sinter and they should not be used at temperatures above 1050°C. If the tool is wrapped in a thin layer of newsprint before being packed in cast-iron chips the surface is also protected against mechanical damage.

Wood-charcoal or coke fines (but not carburizing compounds) act as decarburizers on plain-carbon and low-alloy steels containing medium to high carbon, but act as carburizers on high-alloy chrome steels and hot-work steels. However, it is the temperature and not the composition of the steels that is the factor that determines whether carburization or decarburization will take place. The result of an investigation into the influence of packing media on the hardness, carbon content and amount of retained austenite in some tool steels is shown in Figures 5.16 to 5.19. The normal hardening temperature for each steel was used; heating-up time was about 2 h and holding time, 15 min.

Figure 5.20. Trials with protective paste Marcel CR 20 on steel H 33 (before R P 19).

Hardening temperature: 1000°C.

Holding times: 30 min and 120 min.

Cooling: in air

Under certain conditions borax may be used as a protective agent against decarburization. The borax powder is sprinkled over the surface to be protected and on melting, it forms a covering film. Before hardening, this film should be brushed off, otherwise it retards the rate of cooling of the steel. Borax is often used to protect the engraved surface of coining dies.
HEAT TREATMENT—GENERAL

when they are being hardened. On water-hardening steels the borax film spills off when the steel is quenched and leaves a bright clean surface.

There are also some commercially available protective pastes on the market. These pastes are applied to the protected surface prior to heat treating. The protective efficiency of these pastes is best assessed by testing them on a piece of steel of the grade for which they are recommended. One half of the test-piece is coated with the paste, placed in the hot furnace and held at the normal hardening temperature for, say, 1 h. After hardening, the Vickers hardness in the surface, both treated and untreated, is measured, using varying loads, or using, say, a 5 kp load on a taper section. Examples of such tests are shown in Figures 5.20 and 5.21.

![Figure 5.21: Trials with protective pastes on steel D 2 (Before ROP 57).](image)

Hardening temperature: 1050 °C.
Holding time: 2 h.
Quenching: in oil.
Tempering: 180 °C, 2h

Heat-resisting steel foil has also been put to use as a protective measure against decarburization. The foil is supplied as sheet or bags in which the steel parts are wrapped.

CONTROLLED ATMOSPHERES

These specially produced protective gases, have shown themselves, both technically and economically, to be highly suitable when used for the heat treatment of costly tools and component parts. This is particularly so for large-scale production hardening. The gas is made in a separate plant, whence it is led into the hardening furnace which must be gas-tight to prevent any leakage. Protective gas for hardening is made according to the following three main principles:

1. Exothermic gas is produced by the exothermic combustion (i.e. without the addition of heat) of gas and air; the combustion not going to completion. As the gas contains large volumes of CO₂ and H₂O the greater part of these must be removed by chemical absorption and drying, respectively. If both H₂O and CO₂ are removed, the gas contains CO, H₂ and N₂. This is called monoxets.

2. Endothermic gas is produced by endothermic combustion, i.e. energy must be supplied. The gas, which is usually propane or propylene, is mixed in carefully balanced proportions with air and passed through a gas generator in which a so-called carrier gas is produced. The production of endothermic gas will be discussed in Section 6.3. Gas carburizing on p. 355. The carrier gas usually has a carbon potential of about 0.40%. This means that a great number of steels can be heated in direct contact with this gas without any risk of changes taking place in their surface carbon contents. When treating higher-carbon steels the carbon potential of the gas may be increased by introducing undiluted propane or propylene direct into the furnace. A requirement for a successful result is that a close check is kept on the carbon potential. In many cases a satisfactory check may be obtained by measuring the dew point. To obtain a more exact check and hence a better control of the carbon potential it is recommended that an instrument be used capable of measuring the CO₂ content by means of infrared radiation and converting direct to a figure indicating the carbon potential. (The definition of carbon potential is given on p. 340.) The required carbon potential of the gas should be about the same as that of the carbon dissolved in the steel at the hardening temperature. Table 5.1 gives a selection of carbon potentials that have been used with various steels with good results.

3. Inert gas is a protective gas that, as regards its carbon, oxygen and nitrogen contents, remains unreactive to the steel. From a chemical point of view it is only the Inert Gases, such as argon, that truly satisfy this criterion. The most commonly used unreactive gas consists mainly of nitrogen. On a large scale it is produced by the exothermic combustion of crude gas and air and then cleaned according to various methods. For further information, readers are referred to the literature, for example an article by Grassel and Wynning[3] in which different methods of producing inert gas are described.

A fairly simple method applicable to air-hardening steels is to place the tools in a gastight box fitted with an inlet and an outlet tube. Nitrogen gas from a nitrogen cylinder is led through the box during the whole treatment cycle. When the box is taken out of the furnace the rate of cooling may be increased by placing the box in a current of air from a fan. A large number of such hardening treatments have been applied to tools made from steel H 13.
This method is a simplified form of 'hardening in a protective atmosphere' and the necessary equipment can be installed without considerable capital expenditure.

**Vacuum Hardening**

This has become increasingly popular during the last decade. The charge is placed in a cold furnace which is then evacuated by means of vacuum pumps. For most cases an operating pressure of $10^{-2}$ torr is acceptable. Should a higher vacuum be required, diffusion pumps are necessary which will increase the cost of the furnaces. After the charge has been heated and austenitized, the furnace and contents are cooled to room temperature, usually in a nitrogen atmosphere but in special cases argon is used. New furnace types have been developed in which the charge can be quenched straight into oil while under vacuum. For this purpose special oils have been developed. These oils can be heated to about 200°C, which enables an austempering treatment to be carried out. Figure 5.22 shows one type of vacuum furnace in which the charge can be cooled with gas or quenched in oil in the underlying tank. Figure 5.23 shows the profile of the heating and cooling curves of a cube-shaped tool, the cube length of which was 380 mm (15 in) and weighing 330 kg (750 lb).

The greatest benefit resulting from vacuum hardening is that the surfaces of the parts comprising the charge are not altered by the hardening treatment, i.e. the parts are completely free from surface decarburization or oxidation. There are now available advanced furnace designs that can compete successfully with protective gas furnaces from the points of view of both economy and speed of operation. Extensive literature on vacuum hardening is now available.

### Table 5.1 SELECTION OF CARBON POTENTIAL FOR VARIOUS STEELS

<table>
<thead>
<tr>
<th>AISI BS</th>
<th>Grade of steel</th>
<th>SIS</th>
<th>Before</th>
<th>Carbon content %</th>
<th>Carbon potential %</th>
</tr>
</thead>
<tbody>
<tr>
<td>W 1</td>
<td>1880</td>
<td>B 20 V</td>
<td>1-0</td>
<td>0-80</td>
<td></td>
</tr>
<tr>
<td>O 1</td>
<td>2592</td>
<td>SR 1855</td>
<td>1-0</td>
<td>0-80</td>
<td></td>
</tr>
<tr>
<td>O 1</td>
<td>2140</td>
<td>RT 1733</td>
<td>0-90</td>
<td>0-75</td>
<td></td>
</tr>
<tr>
<td>2550</td>
<td>2560</td>
<td>HKO 1245</td>
<td>0-55</td>
<td>0-45</td>
<td></td>
</tr>
<tr>
<td>A 2</td>
<td>2360</td>
<td>ROP 21</td>
<td>1-0</td>
<td>0-70</td>
<td></td>
</tr>
<tr>
<td>D 2</td>
<td>2310</td>
<td>ROP 37</td>
<td>1-50</td>
<td>0-70</td>
<td></td>
</tr>
<tr>
<td>D 6</td>
<td>2310</td>
<td>RT 60</td>
<td>2-00</td>
<td>0-80</td>
<td></td>
</tr>
<tr>
<td>H 13</td>
<td>2242</td>
<td>ROP 19</td>
<td>0-40</td>
<td>0-30</td>
<td></td>
</tr>
</tbody>
</table>

*Figure 5.22. Vacuum furnace. Charge cooled with gas or oil (after Hayes)*

### 5.3.2 Rate of Heating

If the rate of heating to the hardening temperature is very fast the tool may crack or warp, since sections having different dimensions heat up at different speeds. These differences may be diminished by preheating the tool to a temperature usually lying close below the transformation temperature $A_1$ of the steel. When heating for hardening in a muffle furnace, preheating is not so necessary as when a salt bath is used. Even in this latter case preheating may be omitted if the parts to be treated are small and symmetrical. However, if a salt bath is used, a preheat to about 100°C is always advisable in order to remove any moisture adhering to the steel.

During the initial heating-up stage the surface of the steel is at a higher temperature than the centre. The closer the temperature of the steel approaches the preset temperature, the smaller is this temperature difference, which is shown in Figures 5.24 and 5.25. In actual practice it can therefore be assumed
heat treatment: general

that when the surface has reached the temperature of the furnace, the steel is heated right through.

This assumption runs contrary to the generally widespread belief that the surface of the steel reaches the temperature of the furnace considerably faster than the centre. This belief is based on the observation that the corners of steel objects become warm more quickly than other sections. The diagrams in Figures 5.24 and 5.25 were obtained by inserting thermocouples at the centre of test specimens, thereby ensuring that the heating effect from the end surfaces would not influence the readings. That the ‘corner effect’ really does exist has been shown in a heating trial carried out on a large die block. One thermocouple was located at a corner, 50 mm below the surface; another at the middle of the die, also 50 mm below the surface and a third thermocouple right at the centre. As is seen from Figure 5.25 the corner heated up much more quickly than did the other parts of the block. As regards the other surfaces it is seen that even in this large section the temperature as registered in the middle followed closely that at the surface, once the latter had reached the temperature of the furnace.

The heating-up time may be calculated from standard expressions. Based on trials and with the help of these expressions, the heating-up time necessary in an oil-fired muffle furnace has been calculated for various steel dimensions and plotted in a diagram as a function of the cross-sectional area, Figure 5.27. This diagram is valid for plain-carbon and low-alloy steels. A presupposition is that the length of the part is at least twice the size of the diameter or diagonal. The dimension 950 × 500 mm (Figure 5.26) is equivalent to a round bar 700 mm in diameter. According to Figure 5.27 the heating-up time for a round bar of this diameter is 20 h, which is in quite good agreement with the 22 h indicated in Figure 5.26.

The time of heating-up depends not only on the dimensions of the steel but also on the heating capacity of the furnace and the degree of packing.
Hence it is obviously necessary to gain experience with specific furnaces and to compile tables or diagrams for bars of various dimensions. An example illustrating this is shown in Figure 5.28 which traces the heating curve of the centre of a 75 mm diameter bar made from steel D 2. Preheating was carried out in an 80 kW convection furnace measuring 800 mm diameter by 2000 mm. The final heating took place in a 160 kW furnace of the same dimensions. Note the rather long time to reach the preheating temperature (1.5 h) compared with the time taken to reach the hardening temperature.

Figure 5.27. Time for heating to hardening temperature, 850°C, for bars of varying diameters up to 1000 mm. Length 2 x diameter. Starting temperature 20°C. Oil-fired muffle furnace, calculated values.

Figure 5.28. Heating curve for a 75 mm diameter bar preheated in a furnace 800 mm diameter x 2000 mm, rated at 80 kW. Final heating in a furnace of the same dimensions, rated at 150 kW.

5.3.3 Hardening temperature

For each grade of steel, on the basis of a series of practical trials, a range of temperatures has been established to which the steel is to be heated for hardening. This temperature range, also called the quenching range, is chosen so as to give maximum hardness and at the same time maintain a fine-grained structure in the steel. Figure 5.30 illustrates how to determine the hardening temperature for steel S15 2092 (Bolors SR 1855). The specimen dimension in this case being 30 mm diameter x 100 mm and a holding time of 20 min at the hardening temperature. The micrographs were taken 2 mm below the surface of the specimen and show that 850°C is the lowest permissible temperature that can be used along with the rate of cooling actually employed. A lower hardening temperature would give rise to the formation of pearlite and bainite and the hardness at the surface would be inadequate.

As the temperature increases, so does the grain size and also the amount of retained austenite. The trial run shown in Figure 5.31 illustrates this fact. The specimens were tempered at 200°C in order to bring out more clearly the retained austenite. At 920°C and 970°C the retained austenite may be discerned as light angular areas. An untrained observer may confuse the retained austenite with carbides which in Figure 5.31a are clearly visible as round or oval grains. The acicular areas in b and c are martensite needles. When these are distinctly visible in the microscope, it usually implies that the fracture may be assessed as No. 7 or coarser, on the JK scale.
Figure 5.30. Microstructure of test-pieces of steel 15S209 (Bohm, S.R. 1855) after hardening from 890 to 850°C. Dimensions of test-pieces: 30 x 100 mm. All 400 x

<table>
<thead>
<tr>
<th>Hardening temperatures</th>
<th>(a)</th>
<th>(b)</th>
<th>(c)</th>
</tr>
</thead>
<tbody>
<tr>
<td>JK Fracture number</td>
<td>10</td>
<td>10</td>
<td>10</td>
</tr>
<tr>
<td>Hardness (HRC)</td>
<td>55</td>
<td>61.5</td>
<td>66</td>
</tr>
</tbody>
</table>

Figure 5.31. Microstructure of test-pieces of steel 15S209 (Bohm, S.R. 1855) after hardening from 870 to 970°C, followed by tempering at 200°C. Dimensions of test-pieces: 30 mm diameter x 100 mm. All 400 x

<table>
<thead>
<tr>
<th>Hardening temperatures</th>
<th>(a)</th>
<th>(b)</th>
<th>(c)</th>
</tr>
</thead>
<tbody>
<tr>
<td>JK Fracture number</td>
<td>9</td>
<td>7</td>
<td>5</td>
</tr>
<tr>
<td>Hardness (HRC)</td>
<td>62.5</td>
<td>62</td>
<td>61</td>
</tr>
<tr>
<td>Retained austenite, %</td>
<td>12</td>
<td>20</td>
<td>28</td>
</tr>
</tbody>
</table>
Based on actual trials the quenching range for this steel has been fixed at 850–880 °C. The hardening temperatures for the most commonly occurring steels are given in Chapter 6 in connection with each type of steel.

As has been previously mentioned, by raising the hardening temperature the hardenability of steel can be increased. This increase is due to the greater amount of carbide going into solution and to increased grain size. When high-speed steel is being hardened, a temperature of some tens of degrees below the melting range is used. Only a slight increase above the ordinary temperature will cause incipient fusion round the carbides. This molten phase will 'permeate' through the grain boundaries and hence reduce the toughness. A further increase in temperature may lead to pronounced melting. The steel is then said to be burnt, Figure 5.32. The danger of melting increases with the degree of segregation. The larger the dimension the greater the segregation. For this reason a lower hardening temperature should be chosen for high-speed steel tools made from heavy sections. Even for dimensions corresponding to 100 mm diameter a hardening temperature near the lower limit of the customary temperature range should be chosen.

5.3.4 Holding time at temperature

When the steel has reached the hardening temperature it is austenitic provided that the temperature has been correctly chosen. The time of holding at the hardening temperature depends on the desired degree of carbide dissolution. Since the amount of carbide is different for different types of steel the time of holding is also dependent on the grade of steel.

Plain-carbon and low-alloy structural steels which contain easily dissolved carbides require only a few minutes' holding time after they have reached the hardening temperature. In order to make certain that there has been sufficient carbide dissolution, a holding time of 5–15 min is quite sufficient.

For medium-alloy structural steels a holding time of 15–25 min is recommended, irrespective of the dimension.

For flame and induction hardening which enable the setting of accurately determined heating times, considerably higher hardenability temperatures are used than for conventional hardening. Besides, the time of holding can be reduced to a few minutes and in some instances even to a second or so.

For low-alloy tool steels, on the other hand, a definite holding time is necessary to ensure that the required hardness will be attained. A suggested figure for the holding time is 0.5 min per millimetre of section thickness; the minimum and maximum time, however, should be 5 min and 1 h respectively. Why the holding time should increase with section thickness is due to the fact that the size of the carbide grains and hence their reluctance to go into solution increases as the section dimension increases. Besides, most low-alloy steels generally have insufficient hardenability which, however, may be improved, particularly in heavy sections, by promoting the increased dissolution of carbides. This may be brought about by increasing the time of holding at temperature.

High-alloy chrome steels require the longest holding time of all the tool steels. To a very great extent, however, the holding time for these steels is dependent on the hardening temperature, see Figure 5.33, which applies to
HEAT TREATMENT—GENERAL 211

steel D2. Each hardening temperature requires a definite holding time in order that maximum hardness should be obtained on hardening. Too short a time results in a lower hardness, caused by an insufficient amount of carbide going into solution; too long a time also results in a lower hardness due to the presence of retained austenite. A suggested figure for the holding time is 0.5–0.8 min per millimetre of section thickness, with minimum and maximum times of 10 min and 1 h respectively. The 0.5-min factor is used when the hardening temperature lies near the upper limit of the hardening range and the 0.8-min factor when the temperature lies near the lower limit.

Hot-work steels contain carbides that do not go into solution until about 1000 °C. At these high temperatures grain growth is fairly rapid and for this reason the time of holding must be limited. If the heating is carried out in a salt bath, in which it is easy to see when the steel has reached the desired temperature, 20–30 min is a normal holding time which is by and large independent of the dimensions of the part being treated. If the tool is packed in a box it is more difficult for the heat treater to judge when it is at the right temperature. Under such circumstances the best temperature to use is the one at the lower limit of the hardening range since a longer holding time may then be used, provided it does not exceed 1 h.

The high-speed steels are the most highly alloyed of all tool steel. Their hardening temperatures lie between 1200 and 1300 °C. In order to avoid excessive grain growth, which may easily occur at such high temperatures, the current practice is to use holding times of a few minutes only. In the case of high-speed steel, the normal heat treatment of which is preheating to 800–1000 °C followed by immersion in the salt bath, the time of holding at the hardening temperature is taken as the time of immersion in the salt bath. The times given in Table 5.2 apply to tools heated to conventional hardening temperatures after having been preheated to 850 °C.

Table 5.2 HARDENING OF HIGH-SPEED STEEL. INFLUENCE OF SECTION THICKNESS ON IMMERSION TIME

<table>
<thead>
<tr>
<th>Section thickness</th>
<th>Immersion time</th>
</tr>
</thead>
<tbody>
<tr>
<td>mm</td>
<td>min</td>
</tr>
<tr>
<td>Less than 10</td>
<td>3</td>
</tr>
<tr>
<td>10–20</td>
<td>4</td>
</tr>
<tr>
<td>20–30</td>
<td>5</td>
</tr>
<tr>
<td>30–40</td>
<td>6</td>
</tr>
<tr>
<td>40–50</td>
<td>7</td>
</tr>
</tbody>
</table>

The times given in Table 5.2 can be exceeded by 50% without adverse consequences. When high-speed steel is hardened from lower temperatures, e.g. 1000–1100 °C, the times of immersion should be trebled.

HOLDING TIME IS HARDENING TEMPERATURE

In Figures 1.7 and 1.9 the interrelation between time of holding, hardening temperature and amount of carbide dissolved was shown. Diagrams of this type for each grade of steel would be most useful to the heat treater. The diagrams should be supplemented with data on grain size, hardness on hardening and amount of retained austenite. In Figure 5.33 a start has been made along these lines. Additional data concerning other tool steels including grain size and retained austenite is presented in Chapter 6.

HOLDING TIME BY HEATING-UP TIME

For the sake of simplicity the times given in Table 5.3 refer to holding times, after preheating, in a salt bath—short heating-up time—and in a muffle furnace—very long heating-up time, specially if pack heating. The heating-up time necessary, in the case of a muffle furnace, to raise the temperature of the steel through the final ten or so degrees increases sharply with increasing section thickness. Carbide dissolution, however, begins some 50–100 °C below the ordinary hardening temperature and this fact should be taken into consideration when using long heating times. For the large die block referred to in Figure 5.26 the heating-up time between 825 and 850 °C is about 4 h. Hence, when heating in a muffle furnace it should be possible, on economic grounds, to reduce the time of holding to half of that previously reported; always remembering to keep to the stipulated minimum number of minutes.

When heating-up a piece of steel in a muffle furnace the heat treater might well make use of the very rough and ready approximation of the holding time for all steels, except high-speed steels, viz. 20 min. This time is quite independent of the section thickness and the hardening temperature should be the one in the middle of the recommended temperature range. Therefore it is important for reasons of economy to study the characteristics of the furnaces at the disposal of the heat treater and, as suggested previously, to draw up recommended times for the heating-up of different dimensions and batch weights.

5.3.5 Methods of cooling

The cooling methods used for hardening depend on the grade of steel, the shape of the part and the properties to be imparted to the steel. The three following methods are described below: direct quenching to martensite, martempering and austempering (Figure 5.34).
Direct quenching was doubtless the original hardening method and it is still the commonest one. According to this method the steel is cooled or quenched straight from the hardening temperature to room temperature or to the slightly higher temperature of the cooling medium being used. This medium may be water, oil or air.

**Martempering**

The principle underlying martempering is most clearly understood by referring to a TTT diagram (see Figure 5.34). In this hardening treatment the cooling takes place in two stages. First the tool is quenched in a molten salt or metal bath kept at a temperature somewhat above that at which the austenitic structure of the steel starts to transform to martensite (i.e. the $M_s$-temperature). The steel is held there until temperature equalization is complete. Next, the tool is allowed to cool freely in air, which results in the formation of martensite. Due to this method the temperature difference between the surface and the centre during the martensite formation is much smaller than it is in conventional hardening. Consequently the austenite transforms to martensite practically simultaneously throughout the various parts of the tool even if the latter has widely different section thicknesses. As a result there is less residual stress and a minimum of distortion. The final structure and properties obtained by martempering the steel are, in principle, the same as those obtained from conventional hardening but the amount of retained austenite is generally larger after martempering.

In order to carry out martempering successfully, the composition of the steel must be such that the steel obtains a martensitic structure—if this is what is required—when it has been quenched in the martempering bath and then air-cooled. Therefore this heat-treatment process is confined mainly to the oil-hardening and air-hardening grades. Table 5.4 lists the temperatures of quenching and martempering as applicable to some popular tool steels. Included in the table are also the largest dimensions suitable for martempering, i.e. such as will develop full superficial hardness on being quenched in a clean, well-stirred martempering bath. 300 mm is set as the maximum diameter. In some instances larger dimensions can be martempered but in that case large-capacity baths and effective stirring equipment are required.

The heating to the temperature indicated in Table 5.4 is carried out in the same way as for conventional hardening. When sufficiently austenitized, the tool is transferred to the martempering bath, the capacity of which should be large enough to take up the heat from the tool without appreciably raising the temperature of the bath. The tool is kept moving and submerged in the bath long enough for the core to reach the temperature of the bath. However, the time of holding should not be so long that the austenite starts to transform to bainite, for if this should happen the tool will not attain full hardness. A suitable holding time is usually 2–4 min per each 10 mm of section thickness.
The shorter time is applicable to low hardening temperatures, say around 850 °C and the longer time for higher temperatures, ca 1000 °C.

Since the steel is still in the austenitic state when it is removed from the bath it can be subjected to any necessary straightening operation during the

<table>
<thead>
<tr>
<th>AISI BS</th>
<th>Designation of steel</th>
<th>Austenitizing temperature °C</th>
<th>Martempering temperature °C</th>
<th>Maximum diameter mm</th>
<th>in</th>
</tr>
</thead>
<tbody>
<tr>
<td>2102</td>
<td>SR 1855</td>
<td>850-880</td>
<td>225-250</td>
<td>75</td>
<td>3</td>
</tr>
<tr>
<td>D 1</td>
<td>2140</td>
<td>810-840</td>
<td>225-250</td>
<td>60</td>
<td>24</td>
</tr>
<tr>
<td>D 2</td>
<td>2310</td>
<td>900-980</td>
<td>225-250</td>
<td>150</td>
<td>6</td>
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<tr>
<td>D 6</td>
<td>2312</td>
<td>960-1000</td>
<td>225-300</td>
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<tr>
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<td>820-840</td>
<td>225-300</td>
<td>125</td>
<td>5</td>
</tr>
<tr>
<td>S 1</td>
<td>2710</td>
<td>880-920</td>
<td>350-400</td>
<td>25</td>
<td>1</td>
</tr>
<tr>
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<td>2242</td>
<td>1000-1050</td>
<td>300-500</td>
<td>100</td>
<td>4</td>
</tr>
<tr>
<td>H 10 A</td>
<td>QRO 45</td>
<td>1100-1150</td>
<td>300-500</td>
<td>100</td>
<td>4</td>
</tr>
</tbody>
</table>

subsequent air cooling. This capacity is particularly valuable when hardening long tools such as reamers, mandrels, etc. The straightening is commenced as soon as the tool has been taken out of the bath but generally not above about 350 °C and should be completed before the temperature has fallen to about 150 °C. After the tool has been removed from the bath and straightened, if necessary, it is allowed to cool in air. It is during this cooling that the hard martensite is formed. Then, as in conventional hardening, the tool is tempered when the temperature has fallen to 75 °C or 50 °C. The curve in Figure 5.35 shows the profile of the complete heat treatment programme for a roll made from steel D 2, diameter 103 mm, length 1000 mm. The thermocouple that indicated the temperature was located at the centre of the roll. After being preheated to 550 °C the roll was heated in a protective atmosphere of nitrogen gas in a pit furnace rated at 160 kW. In this instance the hardening temperature was high, viz. 1030 °C, and the time of holding at this rather high temperature was 30 min. The roll was quenched in a salt bath at 220 °C, held there for 45 min and then allowed to cool freely in air. The total heat-treatment time was some 8 h. To this should be added the time required for tempering plus the time taken for straightening, if the latter operation should have been deemed necessary.

Figure 5.36 shows how a pressure die-casting die, made from steel H 13, is being transferred to the martempering bath. This tool is a rather special case. Its predecessor had suffered a large amount of distortion and after but a few shots it had begun to crack. The die shown in the photograph had been transported some 1500 km to be hardened by heat-treatment experts, who

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**Figure 5.35**: Heating and cooling programme for hardening of a roll, diameter 103 mm, made from steel D 2 (Before ROP 57)

**Figure 5.36**: Martempering of a pressure die-casting die, made from H 13 (Before ROP 19)
had it martempered. The result was negligible distortion and an above-average service life for the die.

AUSTEMPERING

By applying a heat treatment termed austempering it is possible, in certain cases, to obtain greater toughness than that obtained after conventional hardening and tempering to the same hardness. There are no excessively large or harmful stresses induced in the tools and, consequently, distortion is small. As in the case of martempering, the parts to be hardened are quenched in a salt bath, the temperature of which is kept above that of the martensite formation. The holding time in the bath, however, is so long that the austenite is practically completely transformed to bainite (see Figure S.34). The resulting hardness depends on the temperature of the salt bath; a high temperature

![Graph showing Temperature vs. Hardness with labels A, P, B, Ms.]

Figure S.37(a). Austempering of steel O 1 (Before RT 1733)

218 HEAT TREATMENT—GENERAL

gives a low hardness. Diagrams of isothermal transformations (TTT diagrams) often indicate the hardness obtained at different temperatures after the formation of 95–98% bainite. Conventional punching tools made, for example from steel O 1 are seldom able to utilize fully the increased toughness resulting from austempering. Very thin-walled tools, such as hollow punches austempered to 58–60 HRC, have given remarkably good service. This also applies to the type of punching tools used for making tin cans. When worn, the

![Graph showing Temperature vs. Hardness with labels A, P, B, Ms.]

Austenitizing temperature 850°C

Temperature

800
700
600
500
400
300
200
100

Hardness

HRC

0
10
20
30
40
50
60
70
80
90
100

A = Austenite
P = Pearlite
B = Bainite
Ms = Start of martensite formation

Figure S.37(b). Austempering of steel S1S 2092 (Before SR 1855)

tool is renewed by upsetting and then ground to size. Figures S.37a and b exemplify the austempering temperature-time programme for two steels.

All steels and dimensions that are capable of being martempered as described in the preceding section can also be austempered. In addition, it is
common practice to austemper certain low-alloy structural steels, e.g. spring steel, up to about 10 mm section and plain-carbon steel up to a maximum diameter of 5 mm. The length of time necessary for complete transformation restricts the application of the austempering treatment to high-alloy steels, since the time needed will often be uneconomically long. Cooling from the austempering temperature takes place freely in air. No tempering is required.

Austempering usually improves the toughness, particularly in that hardness range in which conventionally hardened and tempered steels are susceptible to a reduction in their impact strength. In ordinary low-alloy steels this reduction generally occurs when they are tempered around 300 °C.

PATENTING

The actual patenting treatment may be regarded as a variant of austempering. It is applied mainly to unalloyed and low-alloy steel wire with a carbon content of 0.6-1.1%. The treatment takes place at about 500 °C in a molten lead bath through which the wire passes continuously. At this temperature very fine lamellar pearlite is formed. In spite of its high hardness this structure has shown itself to be well suited for wire drawing since it can accommodate large reductions of area without fracture.

5.3.6 Quenching media

When hardening tool steel the heat treater generally aims at obtaining a martensitic structure, at least in the surface layers of the steel. Hence the rate of cooling must be controlled so that pearlite or bainite formation is avoided. Since the dimensions of the tool are usually decided in advance the required depth of hardening must be governed by a correct choice of steel and cooling medium. For plain-carbon steels the cooling medium is practically always water. The more highly alloyed the steel, the milder is the cooling medium required, and the milder the medium, the smaller is the distortion that will occur. In the following is presented a survey of the commonest cooling media and the cooling capacity of each under different conditions.

WATER

Water is probably the oldest cooling medium used for hardening and it has remained the major coolant throughout the ages. Pure water, however, is rather unsuitable as a cooling medium since its greatest cooling efficiency occurs at 300 °C, i.e. the temperature at which martensite formation starts in many steels. By adding 10% of common salt or soda to the water, its cooling capacity is increased very considerably and at the same time its greatest heat-extracting capacity now occurs at 500 °C. These points are illustrated in Figure 5.38 which also shows the rapid fall in cooling capacity as the temperature of the water rises above 60 °C. The over-all best result is obtained when the water temperature is between 20 °C and 40 °C. The great drawback with water, as mentioned above, is that the rate of cooling is high in the temperature range of martensite formation. This exposes the steel to the simultaneous influence of transformation stresses and thermal stresses, the combined effects of which will increase the risk of crack formation.

Figure 5.38. Cooling capacity of pure water and brine. Test carried out with 20 mm ball of steel (after W. Peter)

The danger of cracking occurring during water quenching may be reduced if the steel is removed from the water when it has cooled to some 200-400 °C and is rapidly transferred to an oil bath. Incidentally, this is a very neat way of increasing the depth of hardening of oil-hardening, low-alloy steels.
Figure 3.39: Cooling curves for round bar stock, diameter 50-2000 mm, water cooling.

Figure 3.39 shows the cooling curves for different dimensions on being water quenched.

Oil.

Oil cooling is much slower than water cooling (see Figure 3.40). The rate of cooling is greatest at about 600 °C and is relatively slow in the range of martensite formation. Since oil has a rather low capacity for heat extraction relative to water, its use as a coolant for medium to low-alloy steels is restricted to light sections. For example, on quenching steel O1 in still oil, it is not certain that full superficial hardness is obtained in dimensions around 100 mm. A reliable and tested way of increasing the cooling capacity of oil is by agitating vigorously the bath or the charge. This simple technique should not be

Figure 3.40: Calculated cooling curves for round bar stock, diameter 100-900 mm, oil cooling. B = radius, r = distance from centre (after Hambler, Simidic and Brenz)

Figure 3.41: Hardening of a 0.50% carbon steel (Before B 10). Influence of agitation of oil on depth of hardening. Temperature 660 °C
underrated. Figure 5.41 shows the effect the agitation of the oil has on the hardness increase in a 0.45% C steel.

Another way of increasing the cooling potential is to raise the temperature of the oil to 50–80°C. This increase in temperature makes the oil more fluid and hence increases its cooling capacity. Figure 5.42 shows how the cooling capacities of a conventional mineral oil and a fast-quenching oil are influenced by their respective temperatures.

**Mixtures of Water and Oil (Emulsions)**

By emulsifying water and ‘water-soluble’ oil in various proportions it is possible to obtain cooling media of various cooling capacities. In this respect, however, such media are usually inferior to the oil itself. Figure 5.43 shows that an emulsion of 90% oil and 10% water has cooling properties inferior to those of unadulterated oil. Even an emulsion containing 90% water and 10% oil is inferior to oil in this respect. Around 300°C the cooling capacity of this emulsion begins to outstrip that of oil but the consequences of this may be rather unfortunate since the martensite transformation has usually started about this temperature.

Roughly the same result makes itself felt if water is inadvertently added to ordinary quenching oil (Figure 5.44). This may lead to cracking on

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**Figure 5.41. Comparison of water, oil and emulsions of water and soluble oil.** Centre cooling curves for still-quenched 18/8 stainless steel specimens 2 in diameter x 2 in long, indicating comparative cooling characteristics of plain water, conventional oil and two emulsions of water and soluble oil, at temperatures of 75°F (24°C) and 125°F (52°C). (Reproduced by permission from Metals Handbook, Vol. 2, American Society for Metals, 1954)

**Figure 5.44. Effects of water contamination on the quenching power of fast oil at 130°F (55°C).** In quenching type 304 stainless steel specimens 2 in diameter x 2 in long. Oil was not circulated and specimens were not agitated. Thermocouples were located at the geometric centre of specimens. (Reproduced by permission from Metals Handbook, Vol. 3, American Society for Metals, 1964)
hardening particularly if the steel is a deep-hardening one because then the martensite formation will start at the surface considerably in advance of the centre which, when it transforms, will increase the stresses in the surface. If the water and oil are not properly emulsified the former having collected at the bottom of the tank instead, a rapid heating-up of the water followed by steam generation may give rise to an explosion.

The presence of water may be ascertained by dipping a narrow-bore tube, preferably of glass and temporarily closed at its upper end, into the tank at its lowest point. On opening the tube the operator lets the quenching fluid rise in the tube. The opening is then again closed, the tube withdrawn and its contents examined.

**WATER WITH SPECIAL ADDITIVES**

The ideal quenching medium should cool the steel quickly to the $M_T$-temperature and then rather slowly to room temperature or to the temperature of the bath. By the addition of certain substances to water it is possible to obtain a quenching medium possessing nearly ideal properties. Mention may be made of an additive, the trade name of which is Aquaquench A, which on analysis, is found to consist mainly of 8% hydroxyl-substituted polyether and 92% water (Figure 5.45). With just a moderate addition (about 2%) of this substance to the water it has been shown that during the quenching of unalloyed steel the risk of cracking is reduced without any loss of hardness being incurred. An addition of 15% produces a quenching medium having almost the same cooling properties as a quenching oil, with the added advantage that the danger of fire is completely eliminated. During quenching, the temperature must not rise above 30°C in still baths or above 45°C in agitated baths. When in use it is found that the pH of the bath decreases and hence it must be adjusted by means of neutralizing additions to keep the pH from falling below 7.5, otherwise there is the corrosion danger to contend with.

**AIR**

Low-alloy steels in light sections and high-alloy steels may be successfully hardened by means of compressed air or still air. The advantages of using air are that distortion is negligible and that the steel can easily be straightened during the cooling process. So that it should cool evenly the part is rotated in a steady current of compressed air. One drawback here is that the surface may be oxidized during the cooling. Among the group of high-alloy steels, D6 has a sufficiently high hardenability to enable it to be hardened in quite heavy dimensions. In Figure 5.46 a comparison is made between D6 and D3 with different manganese contents. The most popular of the air-hardening steels is H13. This steel can be hardened in still air in sections up to 200 mm
Figure 5.47. Calculated cooling curves for round bar stock, diameter 100–1000 mm, air cooling. R = radius, r = distance from centre (after Hinzerer, Strålida and Brezna).

in diameter. Figure 5.47 shows the air-cooling curves for various dimensions (cf. Figure 4.30).

SALT BATHS

The most popular salt baths used for cooling purposes consist of approximately equal parts of sodium nitrite and potassium nitrate. They are used in the temperature range 160–500 °C. There are also salts for 500–600 °C. A salt bath is the ideal quenching medium for a steel having a reasonably good hardenability and not too heavy a section. The cooling capacity down to about 500 °C is high and then decreases as the temperature of the steel continues to fall. The lower the temperature of the bath and the greater the agitation, the better is its cooling capacity (see Figure 5.48). In Figure 5.49 it illustrated the effect of agitation on the rate of cooling and on the hardness of a plain-carbon steel containing 0.5% C.
HEAT TREATMENT—GENERAL 229

If the salt bath becomes contaminated its cooling efficiency is very much reduced and when the bath is stirred the foreign particles are kept in suspension in the molten salt. They are apt to attach themselves to the parts being treated and hence impede heat transfer. In such cases it is better to let the impurities sink to the bottom and instead move the parts smartly about in the clear supernatant molten salt. A recommended holding time in the salt bath is 2–4 min per centimetre of section thickness. The shorter time applies to low hardening temperatures and light sections; the longer time to high temperatures and heavy sections.

The cooling capacity may be increased by adding water to the salt bath. The amount of water must be kept within close limits and should amount to 0.1–0.5%. The cooling capacity will then be approximately doubled. The water, as steam, leaves the surface of the bath continuously and should therefore be replenished, either all the time or as periodic additions. A convenient way of adding water is to introduce it into the vortex created by a circulation impeller. It will then immediately dissolve in the salt without any danger of explosion. The water content may be checked by measuring the density of the bath by means of a hydrometer which has been calibrated in a subsequently solidified salt, the water content of which is determined by weighing. The presence of water may be detected by the characteristic sizzling sound that is caused by the evolution of steam when the charge is lowered into the salt. This steam also sets up a stirring action at the surface of the bath.

230 HEAT TREATMENT—GENERAL

cooling. To complete this section, Figure 5.50 shows a graphical compilation of the cooling capacities of various cooling media.

5.3.7 Quenching equipment

The result of the hardening operation by quenching in the above-mentioned liquid media is dependent on their temperature and degree of agitation.

TEMPERATURE

The best temperature range for water is 20–40 °C and for oil, 50–80 °C. The temperature should be checked continually. In order to raise the temperature of the oil it is usual, before the first quenching treatment proper, to cool some hot pieces of steel in the bath. Another possibility is to have an electric heating element immersed in the bath when it is not in use. There are also ready-made quenching tanks commercially available, completely with temperature control. This control is effected by letting hot or cold water flow through heat exchangers fitted in the bath.

AGITATION

When quenching large pieces, the heat treater is generally not able to create the required stirring action by moving the piece. Modern quenching tanks

![Figure 5.50. Cooling curves for centre of an 18% steel specimen, 12.7 mm diameter x 37.5 mm, on being quenched in some conventional quenching media (after Gad)](image)

Parts that have been heated in a cyanide bath containing more than 10% cyanide must not be quenched in a nitrite-nitrate bath on account of the danger of explosion.

A heavy layer of scale on the surface of the steel reduces the rate of cooling. This applies also to salt from the salt bath clinging to the steel. The very thin oxide layer present on parts after their having been pack heat-treated in cast-iron chips appears to have the least retarding influence on the rate of
have a built-in impeller or pump which provides the agitation of the liquid medium. Illustrations of such equipment are shown in Figures 5.51 and 5.52.

COOLING IN QUenchING JIG

This method is used for several reasons. Two common quenching practices may be mentioned. The first one, called press quenching, is used for rings, gears and similar cylindrical flat or elongated parts which are very liable to undergo distortion. After having been heated, the part is clamped in a jig and placed in a press, usually hydraulically operated. Simultaneously with the closing of the press, oil is made to flow over the part which thereby maintains its plane shape during the quenching. The jig should be preheated to about 50°C lest the rate of cooling at the surfaces in direct contact with the jig be too rapid and hence give rise to superficial cracks.

The other type of jig quenching is applied to tools and other parts containing holes or cavities which for design purposes are to be hardened from the inside only. This hardening method is often used for upsetting dies made from plain-carbon steel and intended for the production of cold-headed bolts. The quenching principle is shown in Figure 5.53.

OTHER EQUIPMENT

Tools made from plain-carbon or low-alloy steel may have such a form that their cooling is impeded. Such obstruction to the cooling may be caused, for instance, by undercut cavities or recessed holes in which the cooling medium cannot circulate adequately or in which air bubbles prevent rapid cooling. In such cases it is necessary to increase the rate of flow of the coolant if a

Figure 5.53. Jet quenching of cold-upsetting tool

232 HEAT TREATMENT—GENERAL

martensitic structure is to be obtained, otherwise cracking may occur during service or even during the hardening operation.

EXAMPLE I

A coining die, made from steel W 1 had cracked through the engraving after some time in use. The die was returned for examination but the engraving

Figure 5.54. Duplex structure consisting of martensite and pearlite. 300 x

Figure 5.55. Submerged jet quenching of engraved work of coining die. Steel W 1 (Before B 20 V)
had been almost completely ground off, which somewhat hampered the investigation. Cracks were found at the bottom of the deepest engraving marks where very finely lamellar pearlite was observed in the otherwise martensitic structure (see Figure 5.54). The tensile stresses set up as the discs were being coined caused the die to crack since the pearlite has not the same tensile strength as martensite. Faults of this type may be avoided if, on quenching, the tool is immersed into the water bath and held above a tube which delivers a powerful current of water, as illustrated in Figure 5.55.

**Example 2**

A mandrel made from steel O 1 had cracked while it was being heat treated, or more specifically during the tempering operation. The location of the cracks and the hardened surface zone is shown in longitudinal section in the sketch in Figure 5.56. Owing to an insufficient cooling effect in the hole, no martensite was formed there. The martensite in the outer zone of the mandrel had induced large tensile stresses at the bottom of the hole where the fillet radius was practically zero. The cracks had formed during the tempering, i.e. when the mandrel was being lowered into a vigorously agitated salt bath. Since the outer zone of the mandrel was the first part of it to expand, the tensile stresses on the inside were still further intensified. This failure could have been avoided if quenching oil had been circulated in the mandrel hole by means of some arrangement such as that shown in Figure 5.55. What further aggravated the situation was the insufficient fillet radius at the bottom of the hole as well as the sudden heating to the tempering temperature.

### 5.4 Tempering

When steel is hardened to a martensitic structure the toughness is rather low. On tempering, i.e. heating it to some temperature between 160°C and 650°C, the toughness increases considerably. In Section 1.4 the theoretical background has been described.

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234. **Heat Treatment — General**

5.4.1 Heating to Temperature

Tempering should be carried out in close proximity to the quenching operation as soon as the steel has cooled to 50–75°C. If it is allowed to cool to room temperature before being tempered the steel may crack. Heating for tempering can take place in a conventional muffle furnace, a convection-type furnace (Figure 5.57), i.e. a tightly closed furnace with a circulation fan, or in a salt-bath furnace of the same type as that used for martempering. The heating can also be done either by means of a gas flame or by electric induction. These last methods are used for local tempering.

The heating of complicated tools must take place slowly. Rapid heating, as in an agitated salt bath, will cause the surface layers to increase in volume and this can induce unfavourable stress conditions that may result in cracking. Hardened parts that have been heated or quenched in a salt bath must be freed from any adhering salt before being tempered in a muffle or convection furnace. This salt is most conveniently removed by rinsing the charge in warm water. If the salt residues are not removed they may cause corrosion during the tempering but if this tempering is carried out in a salt bath there is no need to rinse the parts beforehand. Any adhering hardening salt may be very difficult to remove by rinsing but if the tempering is done in a salt bath of the nitrate-nitrite type, the hardening salt is dissolved. The tempering salt is very soluble in warm water.

5.4.2 Rate of Heating

The superficial colour of unalloyed steel changes as the tempering temperature increases, according to the colour scale given in Table 5.5. When tempering is carried out in a salt bath at some temperature above 220°C
it is easy to decide from the clearly visible colour when the steel has taken on the temperature of the bath. When the tempering is carried out in a muffle furnace it is also possible to tell the temperature from the colour but in order to be able to distinguish the shade of colour the surface of the steel must be ground clean, which, however, is not the usual practice with parts that are to be tempered. From the dull red colour produced by temperatures around 600°C it is possible to judge when the steel has reached the desired temperature.

There is an old rule-of-thumb that says that the tempering time should be 1–2 h for each inch of section thickness, but it does not say when to begin the timing, i.e. if it is the moment the charge is introduced into the furnace or when the charge is warm. To settle this question specific tests were performed. The test material consisted of round bars, the lengths of which were at least twice the diameter. The main part of the investigation was carried out in an 18 kW convection furnace, the internal dimensions of which were 360 mm diameter by 600 mm. Tests involving various steel grades, low-alloy as well as high-alloy, showed that the alloying elements had negligible effect on the time to reach the required temperature.

Furthermore it was observed that when tempering, the time taken to reach the tempering temperature is more or less independent of the temperature, whether it be high or low. Where high temperatures are concerned the rate of heating was somewhat more rapid on account of the effect of radiation. Another fact brought to light was that the centre of the test-piece reaches the temperature of the furnace at about the same time as the surface. In Figures 5.58 and 5.59 are shown some of the results of the tests performed.

Table 5.5

<table>
<thead>
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<th>Temperature °C</th>
<th>Temper colour</th>
<th>Temperature °C</th>
<th>Temper colour</th>
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</thead>
<tbody>
<tr>
<td>220</td>
<td>Straw yellow</td>
<td>310</td>
<td>Light blue</td>
</tr>
<tr>
<td>240</td>
<td>Light brown</td>
<td>325</td>
<td>Grey</td>
</tr>
<tr>
<td>270</td>
<td>Brown</td>
<td>350</td>
<td>Grey-purple</td>
</tr>
<tr>
<td>285</td>
<td>Purple</td>
<td>375</td>
<td>Grey-blue</td>
</tr>
<tr>
<td>295</td>
<td>Dark blue</td>
<td>400</td>
<td>Dull grey</td>
</tr>
</tbody>
</table>

Figure 5.58. Time for heating to various temperatures. Steel specimen BS 817 M40 (En 24) 50 mm diameter × 100 mm. 18 kW convection furnace, internal dimensions 360 mm diameter × 600 mm

Figure 5.59. Time for heating to various temperatures. Steel specimen BS 817 M40 (En 24) (Before CRO 861) 150 mm diameter × 300 mm. 18 kW convection furnace, internal dimensions 360 mm diameter × 600 mm
When the furnace is charged with several test-pieces at the same time the temperature obviously rises more slowly than if there is only one test-piece in it. Of course, the ratio of the furnace input to the weight of the charge is decisive in this case. In the tests reported in Figures 5.60 and 5.61 the heating time to temperature was not prolonged by more than 15 min when, in the first instance the number of test-pieces increased from 1 to 15 (diameter 50 mm x 100 mm), and in the second, from 1 to 4 (diameter 150 mm x 300 mm). Similar tests have been carried out in convection furnaces of various sizes, and reasonably good agreement with the above-mentioned tests has been obtained.

Often the parts are put into the furnace when its temperature is below the tempering temperature. The time taken to heat up will then be longer (see Figure 5.62) and in such cases the tempering time is taken from the moment the furnace has reached the preset temperature.

By making use of the data obtained from the tests performed, a diagram has been constructed showing the time for heating as a function of the diameter, up to 150 mm of the bar stock. The time may be taken either from the moment when the charge is loaded into the preheated furnace or from the moment when the indicating instrument shows the preset temperature (Figure 5.63). This diagram cannot be expected to hold for all cases but it

![Figure 5.60. Time for heating to 250°C. Specimen size 50 mm diameter x 100 mm. 1 and 15 specimens respectively. 10 kW convection furnace](image)

![Figure 5.61. Time for heating to 250°C. Specimen size 150 mm diameter x 300 mm. 1 and 4 specimens respectively. 10 kW convection furnace](image)

![Figure 5.62. Dependence of heating-up time on starting-off temperature of furnace. Specimen size 150 mm diameter x 300 mm. 2 specimens. 18 kW convection furnace](image)
240 HEAT TREATMENT—GENERAL

should not be out by more than some 30 min, which is generally of no
practical significance.

The die block described in Section 5.3 has been used for time-temperature
measurements and the values obtained have gone to the drawing up of
Figure 5.64. The heating was carried out in an electric muffle furnace without

forced circulation. Figure 5.65 contains the results of theoretical calculations
regarding the time taken to heat up bars of different dimensions. Round
bars shorter than twice their diameter heat up more quickly, but the heating-
up time is very little affected when the bar length is increased beyond three
times its diameter.

5.4.3 Holding time

According to the description of the progress of the tempering process as
given in Section 1.4, martensite and retained austenite are transformed to
various products. This transformation is not only dependent on the tem-
perature but also on the time factor which in some instances has a very great
effect. Hollomon and Jaffe, amongst others have investigated this time-
temperature factor which has led to the development of the expression for
the tempering parameter $P$ as described in Section 1.4, viz.

$$ P = T(k + \log t) $$

Extensive tests performed at the Bofors Company have shown that this
expression can be applied to all the steels that have been investigated. A
figure of 20 for the constant $k$ is valid for all the steels.

The dependence of the progress of tempering on time is best shown in the
diagram in Figure 5.66. For temperatures up to 450°C, time has very little
influence on the hardness of this steel. At higher temperatures, however, the
time factor is important. The values obtained from Figure 5.66 can be con-
densed into a so-called 'master curve' which is shown in Figure 5.67. Since
curves of this type are rather cumbersome for everyday use, a new type of
The diagram has been evolved by the Bofors Company and is shown in Figure 5.68. The upper part of the diagram is a conventional tempering curve and as such is valid for a 1 h tempering time. Should the effect of a longer tempering time be required, move vertically downwards to the lower diagram and follow the appropriate sloping temperature line until it intersects the time.

Figure 5.67. The so-called master curve (hardness as a function of the tempering parameter) for steel H 13 (Bofors ROP 19).

Figure 5.69. Time-dependent tempering diagram for steel H 13 (Bofors ROP 19). Oil quenched or air cooled from 1005°C.

horizontal concerned. From this point of intersection move vertically upwards into the tempering diagram and read off the hardness in the usual way. The following example which applies to steel H 13 and to the diagram in Figure 5.68 gives an actual illustration of the procedure. After a tempering treatment of 1 h at 500°C the hardness is 56 HRC. If we wish to temper for 10 h at the same temperature, follow the 500°C line in the lower diagram.
until it intersects the 10-h horizontal. Proceed straight up to the top diagram and read off the hardness, viz. 55 HRC. In the same way, tempering for 100 h at 500 °C gives a hardness of 53 HRC.

Tempering for 1 h at 600 °C results in a hardness of 49 HRC. If the time is increased to 3 h the hardness falls to 46 HRC. This latter hardness is also obtained by tempering at 625 °C for 1 h. By visualizing a temperature line at 625 °C and following it down to the 10-h horizontal we obtain a hardness of 35 HRC, i.e. a hardness reduction of 11 units relative to the 1 h tempering.

Notice the large hardness reduction obtained by tempering for 100 h at this temperature compared with the small reduction at 500 °C after the same tempering time.

The high-alloy hot-work tool steels that show a sudden fall in hardness around a tempering temperature of 600 °C are obviously very sensitive to a few hours' difference in the holding time. There are untold instances of failures with such steels owing to the heat treaters' not being aware of the very strong influence of the time factor. Such failures are now avoidable if the new Bofors-diagrams are used. Chapter 6 contains further examples of these new tempering charts. From the charts it is seen how the hardness is dependent on temperature and time. The question now crops up: how long should the holding time be? There are many and varying opinions among heat-treatment experts on this question. For example, should it be a short holding time (1 h) at a high temperature or a long holding time (10 h) at a low temperature or is the more usual holding time of 2 h to be preferred? Results from investigations of a number of heat-treated steels have shown that figures for the yield point, ultimate tensile strength, elongation, reduction of area and impact strength have not been influenced in any special direction when, without going to extremes, the temperature and holding time have varied as permitted by the tempering parameter.

Figure 5.69 shows a diagram for steel En 29 B (Bofors RO 7155), the mechanical properties of which have been obtained by various combinations of treatment temperature—between 525 °C and 700 °C—and time—10 min to 8.5 h. Even longer holding times would yield the desired hardness but for some steels, impact strength may be considerably reduced (temper brittleness). In the absence of more detailed investigations a holding time of 1–2 h is recommended. For very large components the holding time may be prolonged by an hour or so in order to be certain that they have reached the desired temperature. Note that the holding time is regarded as beginning when the surface of the steel has reached the required temperature.

### 5.4.4 Double tempering

As was stated earlier, tempering should be carried out with a minimum of delay after the hardening operation and usually when the heat-treated part has cooled to between 75 °C and 30 °C. The martensite that forms between 75 °C and room temperature may induce cracking. In some instances even at room temperature some martensite formation takes place isothermally. The temperature at which to interrupt the quenching process can be selected...
by referring to the position of the \( M_s \) and \( M_f \) temperatures in the TTT diagram, or better still, the \( M_s \) temperature if it is included in the TTT diagram. The higher the alloy content of the steel, the lower are the \( M_s \) and \( M_f \) temperatures and hence the lower the temperature at which the cooling may be interrupted without risk of hardening cracks appearing.

If the cooling is interrupted at, say, 80 °C and immediately followed by a tempering treatment at 170 °C, the formation of martensite ceases and a fairly large amount of untempered martensite is produced as the steel cools to room temperature. For this reason a steel that has been subjected to interrupted cooling at some intermediate temperature should be tempered a second time. For example, a case was experienced when surface-hardened bars cracked during transportation. Metallographic examination showed that the surface layers contained about 30\% untempered martensite—see Figure 5.70a. When tempered at 200 °C the structure shown in Figure 5.70b was obtained.

![Figure 5.70](image)

(a) (b)

Figure 5.70: Microstructure resulting from insufficient hardening immediately followed by tempering (a), and the same specimen after another tempering treatment (b).

(a) 30\% Martensite as a result of inadequate hardening practice.
(b) Same specimen after tempering at 200 °C. Both 300×

Steels that form bainite on tempering or that already have a bainitic structure after hardening, such as heat-treatable alloy steels in heavy sections, need to be tempered once only. In high-alloy chrome and high-speed steels the retained austenite is transformed to martensite on cooling from a tempering temperature around 500 °C. Therefore such steels should be retempered, which toughens the newly formed martensite. If the required hardness has been reached after the first tempering treatment the second tempering must take place at a lower temperature lest the hardness of the steel should be reduced. In practice, this second tempering is carried out some 10–30 °C below the first tempering temperature. A suitable retempering temperature may be deduced from the temperature-time tempering diagrams. If the tempering treatment of a costly tool cannot follow hard on the quenching, it should be kept between 50–100 °C in a heating oven during the waiting period.

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246 HEAT TREATMENT—GENERAL

6. TEMPERING AFTER GRINDING OR ELECTRICAL SPARK MACHINING

When a hardened and tempered tool is subjected to grinding or spark erosion machining operations, the heat developed in the surface of the steel may lead to structural transformations and an unfavourable stress pattern which may result in cracks and chipping after an exceptionally short time. This ill effect can be avoided by an extra tempering treatment introduced after the grinding or spark erosion machining operation.

5.4.5 Temper brittleness

When steel is tempered it passes through a series of transformations that affect not only the hardness but also the toughness. The conspicuous reduction in toughness that occurs on tempering at certain temperatures is called temper brittleness. Due in part to the composition of the steel, this brittleness makes its appearance at different temperatures and by judging from the manifestations of embrittlement at various temperatures the four following regions may be distinguished.

1. 170–350 °C
2. 400–450 °C
3. 475–525 °C
4. 500–570 °C

1 270–350 °C BRITTleness

This britleness range is found in most unalloyed and low-alloy steels, both tool steels and constructional steels. The temperature range coincides with the range in which retained austenite is transformed to bainite and it is highly

![Figure 5.71](image)

Figure 5.71: Hardness and impact strength as functions of tempering temperature for steel S15 2550 (Before H R O 1243) C, 1% Cr, 3% Ni, 0.3% Mo

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probable that some mechanism involved in this transformation is responsible for the loss of toughness. Tests with steels containing very low contents of P and N do not show any particularly conspicuous impact-strength minimum. Consequently these elements may be suspected of taking part in the development of temper brittleness within this temperature range.

An example of the 300°C brittleness is shown in Figure 5.71 which gives the impact values for steel S15 2550 (Bofors HRO 1243). There are many instances when an incorrect tempering temperature has been the cause of costly failures. Further examples are given in Chapter 6.

400–550°C BRITTLINESS

This brittleness range is generally the one referred to when temper brittleness is mentioned in connexion with alloy constructional steels. The development of this embrittlement has been found to depend on both tempering temperature and time. Curves resembling conventional TTT diagrams have been drawn up and one such series of curves is shown in Figure 5.72. The curves represent the impact transition temperatures. A low transition temperature implies

![Image](http://example.com/image.png)

**Figure 5.72.** Dependence of transition temperature on tempering temperature and time for steel SAE 3400 of the following composition: 0.40% C, 0.65% Mn, 0.65% Cr, 1.0% Mo, 0.2% Ni. Water quenched from 960°C. Double tempering: first tempering for 1 h at 765°C, water quenching. Second tempering—time and temperature as in diagram (after Jaffe and Baffum).

that the steel is tough (see Section 3.2). According to Figure 5.72, if the tempering time is less than 10 h, a tempering temperature of 450°C results in a transition temperature of about 60°C, and if raised to 500°C the transition temperature is about 30°C. Improved impact strength (i.e. lower transition temperature) is again obtained as the tempering temperature is raised above 550°C.

This brittle range differs from the foregoing one in that the degree of brittleness is dependent on the rate of cooling after tempering. Even if the steel has been tempered at a higher temperature a slow cooling through this critical range may result in a deterioration of the impact strength. For this reason heat-treatable steels in heavy sections are particularly susceptible to this type of embrittlement which, however, is fairly easy to avoid simply by letting the steel cool in water or oil from the tempering temperature. An alloy steel that has become embrittled may be restored to its tough condition by being reheated above the brittle range. The toughness may even be restored to a certain extent when the steel is heated to the critical range but it must be quenched therefrom.

Of the elements that tend to promote temper brittleness, the most significant is P when present in amounts of more than about 0.015%. At this P level, Cr and Mn each have an unfavourable influence too, which is intensified if both elements occur together and in amounts of more than about 1% each. If Ni also is present the susceptibility to embrittlement is further increased although Ni by itself has no deleterious effect. The presence of small amounts of As has in some instances been found to be harmful. Susceptibility to temper brittleness can be reduced by alloying the steel with Mo in amounts from 0.15 to 0.5%.

475°C BRITTLINESS

The 475°C brittleness is the embrittlement that, accompanied by a hardness increase, appears in ferritic and semi-ferritic chrome steels, the Cr content of which lies over 13%. The brittleness is brought about by holding the steel in the temperature range 400–550°C. The brittleness is most noticeable at 475°C. In steels with Cr contents over 25%, the embrittlement may occur after a few minutes heating at 475°C. There are several theories about the cause of this embrittlement which disappears on heating to a higher temperature than 600°C. The hardness increase is said to be due to the coherent precipitation of a complex Cr rich compound. This in turn causes a Cr-depletion of the matrix, resulting in reduced resistance to corrosion (see Sections 5.6 and 6.3).

500–570°C BRITTLINESS

In high-alloy steels, such as heat-worked steels and high-speed steels, finely dispersed carbides are precipitated in this temperature range and during cooling martensite is formed. Both these factors increase the hardness and at the same time the impact strength is reduced. This brittleness is graphically illustrated in Figure 5.73.
5.5 TRANSFORMATION OF RETAINED AUSTENITE

Austenite may remain in the structure after hardening in appreciable quantities and is called 'retained austenite'. Applied stresses may transform this austenite to martensite which in turn may cause the steel to crack. The retained austenite is transformed to martensite either by subzero treatment, i.e., cooling to below -70°C, or by tempering. In the latter bainite or martensite is formed, depending on which temperature range is used for the treatment (see Figure 1.20). As mentioned previously, the amount of retained austenite increases with increasing hardening temperature. It also increases if an interruption in the cooling occurs in the vicinity of the Ms temperature. Since the subject of retained austenite and its transformation is of paramount interest, when tool steel is concerned it will be treated along with the individual tool steel grades in Chapter 6.

MECHANICAL TRANSFORMATION OF RETAINED AUSTENITE

In the introduction to this section it was stated that retained austenite can be transformed to martensite by the application of high stresses. For this reason, tools and machine components that are liable to be subjected to high stresses should be tempered in order to transform the retained austenite. Recommendations for suitable tempering temperatures for each individual steel grade are given in the following chapter.

Hadfield steel, i.e., manganese steel containing 12–14% Mn and 2–2.4% C is completely austenitic after a quench annealing treatment. By subjecting the surface to cold work the austenite is transformed to martensite, resulting in a very considerable increase in surface hardness. According to Schumann, the cold work gives rise to dislocations which create stacking faults which then act as nuclei for α-martensite. This constituent later changes to α-martensite.

In order to increase the hardness of manganese steel intentionally by cold working, a charge of explosive paste may be detonated on it. Shot-blasting by steel sand will also produce a considerable increase in surface hardness (see Figure 5.74). The formation of martensite also in this case results in a volume increase.

STABILIZATION OF RETAINED AUSTENITE

If after hardening, the steel is kept at room temperature for some time or is heated to the temperature range corresponding to the first tempering stage (see Section 1.4) the austenite is stabilized which implies that it has become more difficult to transform when subjected to a subzero treatment. The stabilizing effect increases as the tempering temperature and time increase. The diagrams in Figures 5.75a, b and c show how much of the retained austenite is transformed by a subzero treatment at -180°C after various pre-treatments.

After quenching from, say, 840°C there is 18% retained austenite. If the
subzero treatment is carried out within 5 min after the temperature of the steel has reached 20°C, i.e. almost immediately after quenching, about 70% of the retained austenite will be transformed. If a lapse of 40 min is allowed before the treatment, 60% is transformed and after 50 h only 30% of the retained austenite will respond to the subzero treatment. If the steel is heated to 120°C for 10 min after quenching and then subzero treated, only 30% of the retained austenite will be transformed.

The initial amount of retained austenite is dependent to a very large extent on the hardening temperature. As this temperature is raised there is an increase in the amount of retained austenite that will be transformed after it has been given a specific stabilizing treatment. This is clearly shown in the diagram, Figure 5.76, which shows the stabilizing effect at 20°C. From the diagram

![Diagram showing stabilizing temperature and time on amount of retained austenite](image)

**Figure 5.75. Influence of stabilizing temperature and time on amount of retained austenite that transforms on being subzero treated at -180°C. Ball-bearings steel AISI 52100 (after Lichteurstünz et al.)**

(a) Temperature of quench: 780°C. Retained austenite after quenching: 9.4%.

(b) Temperature of quench: 840°C. Retained austenite after quenching: 12.8%.

(c) Temperature of quench: 900°C. Retained austenite after quenching: 15.2%.

it can be seen that, for example, after holding it for 1000 min at 20°C the retained austenite, on subzero cooling to -180°C, will undergo about 10% transformation if the hardening temperature had been 780°C. For temperatures 840°C and 900°C the corresponding figures are 40% and 60% respectively. By raising the hardening temperature it is thus possible to transform increasing amounts of retained austenite by a subzero treatment. By estimating from Figures 5.75 and 5.76 the amounts of retained austenite remaining after the subzero treatment it is found that for the hardening temperatures investigated, they are practically the same, viz. at 6-10%, for stabilizing times between 10 and 1000 min.

If the hardening quench is interrupted somewhere around the \( T_s \)-temperature a similar stabilization of retained austenite is obtained. If the cooling to room temperature is then continued the same effect, in principle, results as that obtained by the subzero treatment in the example given above, viz. a transformation of retained austenite to martensite. The amount of martensite will however in this case be less than after a direct quench to room temperature. The numerical figures quoted in the above description are valid only for the steel examined.
5.6 PRECIPITATION HARDENING

This type of hardening involves two treatments, viz. solution treatment and ageing. It can be applied to low-carbon unalloyed steels, albeit for such it is of hardly any practical use. To introduce the subject, however, the example below is taken from this group of steels. The solution treatment consists of heating the steel to a temperature just below $A_1$ where the solubility of C in ferrite is greatest, 0.09%. (see Figure 5.77). Any N present is also dissolved, a similar diagram being applicable to N. When the steel is quenched to room temperature, C and N are still in solution. After the lapse of some time (ageing) at this or at some slightly higher temperature a finely dispersed precipitation of e-carbide and carbo-nitride—if N is present—takes place. This precipitation induces an appreciable hardness increase, as shown in Figure 5.78. By tempering the steel at some moderate temperature a hardness reduction is obtained.

It is believed that this relatively high increase in hardness is due to the occurrence of the precipitate at the dislocations. Hereby their mobility is impeded which manifests itself as an increase in hardness. As the temperature is raised the precipitated particles increase in size at the expense of their numbers. The dislocations become increasingly mobile and as a consequence the hardness falls. Useful properties resulting from precipitation hardening can be conferred on nickel-base alloys by additions of Al, Ta, Ti or Mo. Subsequent ageing gives considerable hardness increases. Since this ageing takes place at rather high temperatures these alloys are well adapted to high-temperature service.

With the aid of the electron microscope it is possible to follow the progressive growth, with increasing time and temperature, of the precipitated precipitation-hardening alloys are A 286 and the nickel-base alloys designated Nimonic. These alloys are solution treated and quenched from about 1000°C, followed by ageing at about 700°C for 16 h. The hardness increases from 200 H B to 300 H B. Fe-Ni alloys with less than 25%, Ni, on cooling from the

264 HEAT TREATMENT—GENERAL

particles. Figure 5.79 illustrates this for an austenitic alloy designated Pyromet 86012 which has the following composition:

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>Co</th>
<th>Ti</th>
<th>Al</th>
<th>B</th>
<th>Fe max.</th>
<th>max.</th>
<th>max.</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.10</td>
<td>1.0</td>
<td>1.0</td>
<td>5.0</td>
<td>1.0</td>
<td>3.5</td>
<td>2.5</td>
<td>0.75</td>
<td>0.008</td>
<td>0.012</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

A solution treatment at 1095°C for 2 h followed by water quenching gave the alloy a hardness of 160 H B. Ageing it for various times and temperatures gave the hardnesses indicated in Table 5.6. Other examples of austenitic

![Figure 5.77. Solubility field for carbon in n-iron (after Hartschla)]](image)

![Figure 5.78. Hardness variations obtained by artificially ageing a steel containing 0.09% C. Prior treatment: heating to 700°C and quenching)](image)

<table>
<thead>
<tr>
<th>Ageing time</th>
<th>2 h</th>
<th>50 h</th>
</tr>
</thead>
<tbody>
<tr>
<td>75°C</td>
<td>100</td>
<td>150</td>
</tr>
<tr>
<td>800°C</td>
<td>120</td>
<td>170</td>
</tr>
<tr>
<td>850°C</td>
<td>130</td>
<td>180</td>
</tr>
</tbody>
</table>

Table 5.6 HARDNESS HRC

<table>
<thead>
<tr>
<th>Ageing time</th>
<th>745°C</th>
<th>775°C</th>
<th>800°C</th>
<th>830°C</th>
<th>855°C</th>
</tr>
</thead>
<tbody>
<tr>
<td>5 h</td>
<td>35-36</td>
<td>36-37</td>
<td>36</td>
<td>36</td>
<td>33-34</td>
</tr>
<tr>
<td>15 h</td>
<td>38-39</td>
<td>37-38</td>
<td>35-36</td>
<td>35-36</td>
<td>31-33</td>
</tr>
<tr>
<td>50 h</td>
<td>38-39</td>
<td>37-38</td>
<td>35-36</td>
<td>33-34</td>
<td>30-31</td>
</tr>
<tr>
<td>100 h</td>
<td>38-39</td>
<td>37-38</td>
<td>34-35</td>
<td>33-34</td>
<td>30-31</td>
</tr>
</tbody>
</table>
solution temperature, produce a martensite called Nickel-martensite\(^1\) (see micrographs Figure 1.14).

The effect of nickel on the transformations is shown in Figure 5.80. This diagram holds for normal rates of heating and air cooling. If a ferritic alloy containing 18% Ni, for example, is heated, it is found that at 600°C, 10% austenite is formed and at 630°C there is 90% austenite. On cooling, the transformation to α-iron—in this instance to martensite—is suppressed to a lower temperature so that when 315°C is reached, 10% α-iron has formed. At 260°C, 90% of the austenite has been transformed to martensite.

![Figure 5.80](image)

*Figure 5.80. Temperature-transformation diagram for Fe-Ni alloys. N.B. not an equilibrium diagram (after Jones and Pamphrey)*

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Figure 5.79: Growth of precipitated particles with time and temperature in a solution-treated, precipitation-hardened alloy. 10,000 x (after Manier et al.\(^{16}\))

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Figure 5.81: Coherence between the (110) crystallographic plane of martensite and the (0001) plane of Ni\(_3\)Ti in a maraging steel (after Fisher and Antilli)
If the steel is alloyed with, say, Ti for precipitation-hardening purposes, it can be aged at temperatures up to about 600 °C without any appreciable transformation of martensite to austenite taking place.

Nickel-martensite is fairly soft and is probably not subject to the type of tempering effect associated with the more familiar type of martensite. When Ni-martensite is aged, Ni₃Ti is precipitated. Since the dislocation density in this martensite is very high the precipitate is highly dispersed, which explains the large hardness increase that is produced on ageing these so-called maraging steels. Yet another contribution to the hardness increase comes from the coherency strains that arise owing to the similarities in atomic structure between the martensite and the Ni₃Ti precipitate, Figure 5.61[2].

The maraging steel, grade 250, is supplied solution treated, hardness 280 H.B. (The treatment consists of annealing at 820 °C for 1 h followed by air cooling.) The precipitation hardening temperature is 480 °C, holding time 3 h. The mechanical properties given below are obtained:

\[\begin{array}{cccc}
\sigma_0 & \sigma_H & \psi & H B \\
kp/mm² & kp/mm² & % & % \\
175 & 180 & 10 & 50 & 470
\end{array}\]

The maraging steels are discussed in more detail in the next chapter.

5.7 Straightening

Even if all reasonable precautions are taken during the hardening process it is not always possible to avoid some distortion or warping. This is particularly true of long thin components or those having a complicated shape. Slight distortion often disappears during the grinding operation that follows but if not, the part must be straightened. This may be carried out in several different ways.

By peening, which consists of giving the concave side of the warped part a few sharp blows with a peen hammer, the material is stretched on this side and the tool can be straightened in that way.

When press straightening, the component is placed on two supports and pressed on the convex side. The chances of successfully straightening tools harder than 50 HRC are rather poor. Surprisingly, case-hardened as well as nitrided parts can be straightened to a very large extent owing to the lower core hardness. Nitried parts are more easily dealt with if the material is heated to 400 °C and straightened at this temperature.

The most efficient way of carrying out the straightening operation is to take it in combination with the hardening process and while the steel is still in the wholly austenitic state. The most convenient hardening process from the point of view of subsequent straightening is martempering, i.e. when the steel has assumed the temperature of the bath. The straightening is started immediately the steel is taken out of the bath and should be completed when the temperature has fallen to about 150 °C. Straightening during air hardening is also convenient.

Most of the high-alloy steels are easy to straighten as quenched, since in their untempered state they have a high content of retained austenite and their yield points are relatively low. Straightening may also be performed during the actual tempering process. Warped tools can be clamped in a jig and become straightened as a result of the stress relaxation that sets in during the tempering. If there are several warped tools it may be practicable to clamp them together in pairs, placing washers between them if necessary. Straightening by means of local heating with a welding torch requires much experience and a high degree of discrimination on the part of the operator. Such straightening should be avoided if the steel is harder than 50 HRC. It should preferably have been tempered to at least 400 °C. The bent part is heated on the convex side, the torch being moved in one or several straight passes at right angles to the direction of curvature of the bent part. The heated length expands thermally, thereby becoming upset. When it cools, tensile stresses are induced and these tend to straighten the part but there is the risk that the stresses may cause cracking.

5.8 Machining Allowances

Tool Steel

Even if tools do not require clean surfaces from a functional point of view it is a general rule that if the tool is to be hardened all its surfaces must be machined clean. The reason for machining before hardening is to remove scale, surface cracks and decarburization. Surface cracks may initiate further crack growth. A decarburized surface may easily give rise to cracks during hardening since martensite formation starts in the surface layers containing the lower carbon content. Subsequently, as more and more martensite continues to be formed in the core the latter increases in volume and may eventually cause the initially formed martensite case to rupture. For high-speed steel the stipulated maximum acceptable depth of decarburization is 1% of the bar diameter + 0.1 mm per surface. According to AISI the maximum allowable decarburization per surface for hot-rolled tool steel is about 1% of the bar diameter + 0.2 mm.

Constructional Steel

Structural components that are required to meet exacting demands of high strength, in particular fatigue strength, must be machined clean before hardening. Sometimes this machining is confined to the most highly stressed surfaces only. Structural steels generally have lower carbon contents than tool steels and are therefore not so susceptible to quenching cracks as the latter.

Nitriding Steel

Nitriding steels, alloyed with aluminium, require a larger machining allowance than other steels since if there is any surface decarburization it invariably
HEAT TREATMENT—GENERAL

Table 5.7 MACHINING ALLOWANCE PER SIDE FOR ROUND BARS

<table>
<thead>
<tr>
<th>Specified dimension mm</th>
<th>Hot rolled</th>
<th>Allowance per side mm</th>
<th>Forged</th>
<th>Rough machined</th>
<th>Cold drawn</th>
</tr>
</thead>
<tbody>
<tr>
<td>≤ 13</td>
<td>0.4</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>— 0.4</td>
</tr>
<tr>
<td>&gt; 13-25</td>
<td>0.8</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>— 0.8</td>
</tr>
<tr>
<td>&gt; 25-50</td>
<td>1.2</td>
<td>1.8</td>
<td>0.5</td>
<td>0.6</td>
<td>— 1.2</td>
</tr>
<tr>
<td>&gt; 50-75</td>
<td>1.6</td>
<td>2.4</td>
<td>0.5</td>
<td>1.6</td>
<td>— 1.6</td>
</tr>
<tr>
<td>&gt; 75-100</td>
<td>2.2</td>
<td>3.0</td>
<td>0.6</td>
<td>2.2</td>
<td>— 2.2</td>
</tr>
<tr>
<td>&gt; 100-125</td>
<td>2.8</td>
<td>3.7</td>
<td>0.6</td>
<td>2.8</td>
<td>— 2.8</td>
</tr>
<tr>
<td>&gt; 125-150</td>
<td>3.8</td>
<td>4.3</td>
<td>—</td>
<td>1.0</td>
<td>— 3.8</td>
</tr>
<tr>
<td>&gt; 150-200</td>
<td>5.1</td>
<td>5.1</td>
<td>1.2</td>
<td>—</td>
<td>— 5.1</td>
</tr>
<tr>
<td>&gt; 200-250</td>
<td>5.1</td>
<td>6.1</td>
<td>1.8</td>
<td>—</td>
<td>— 5.1</td>
</tr>
</tbody>
</table>

Table 5.8 MINIMUM MACHINING ALLOWANCE (A AND B) PER SIDE FOR HOT-ROLLED SQUARE OR FLAT BARS

<table>
<thead>
<tr>
<th>Specified width mm</th>
<th>&lt; 13</th>
<th>13-25</th>
<th>25-50</th>
<th>50-75</th>
<th>75-100</th>
<th>100-125</th>
<th>125-150</th>
<th>&gt; 150</th>
</tr>
</thead>
<tbody>
<tr>
<td>Hot rolled</td>
<td>A</td>
<td>0.6</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>&gt; 13-25</td>
<td>A</td>
<td>0.6</td>
<td>1.1</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>&gt; 25-50</td>
<td>A</td>
<td>0.8</td>
<td>1.1</td>
<td>1.7</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>1.0</td>
<td>1.3</td>
<td>1.7</td>
<td>—</td>
<td>—</td>
<td>—</td>
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</tr>
<tr>
<td>&gt; 50-75</td>
<td>A</td>
<td>0.9</td>
<td>1.3</td>
<td>1.7</td>
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<tr>
<td></td>
<td>B</td>
<td>1.3</td>
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<td>1.8</td>
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<td>—</td>
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<tr>
<td>&gt; 75-100</td>
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<td>1.4</td>
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<td>2.9</td>
<td>—</td>
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<td>B</td>
<td>1.7</td>
<td>1.9</td>
<td>2.2</td>
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<td>1.7</td>
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<td>2.9</td>
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<td>4.8</td>
</tr>
<tr>
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<td>B</td>
<td>2.4</td>
<td>2.9</td>
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<tr>
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<td>1.8</td>
<td>1.9</td>
<td>2.3</td>
<td>2.9</td>
<td>3.8</td>
<td>4.8</td>
</tr>
<tr>
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<td>B</td>
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<td>3.3</td>
<td>3.7</td>
<td>3.9</td>
<td>4.3</td>
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<td>2.0</td>
<td>2.5</td>
<td>3.2</td>
<td>3.8</td>
<td>4.8</td>
</tr>
<tr>
<td></td>
<td>B</td>
<td>3.0</td>
<td>3.8</td>
<td>4.2</td>
<td>4.6</td>
<td>4.8</td>
<td>4.8</td>
<td>4.8</td>
</tr>
<tr>
<td>&gt; 200-225</td>
<td>A</td>
<td>1.5</td>
<td>1.9</td>
<td>2.5</td>
<td>2.5</td>
<td>3.2</td>
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<td>4.8</td>
</tr>
<tr>
<td></td>
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<td>4.8</td>
<td>4.8</td>
<td>4.8</td>
<td>4.8</td>
</tr>
</tbody>
</table>

References

8. Uhlitzsch, H., 'Mathematical Treatment of the Rate of Change of Temperature on Heating and Cooling a Finite Solid Cylinder and its Application to the Heat Treatment of Large Forgings', Neue Härte, No. 5, 277-287 (May 1959) (in German)
13. Schumann, H., 'Investigation into the Cause of High Rate of Hardening of High-manganese Steel', Neue Härte, 12, No. 4, 220-226 (1967) (in German)
6 Heat treatment—special

This chapter is specifically written for designers and heat treaters actively engaged in practical work. It contains recommendations for suitable heat treatments for various types of steel as well as suggested applications for a number of well-known steels. The chapter recapitulates some of the material of the preceding one and also extends its contents in the context mentioned above. Each section contains an introductory survey of the steels discussed in the chapter.

6.1 HARDENING AND TEMPERING OF TOOL STEELS

Reckoned on a tonnage basis, tool steel represents only a few percent of the total quantity of steel produced but its importance to the industry as a whole is immense. Regrettably this fact is seldom sufficiently appreciated. Perhaps in greatest measure this applies to the heat treatment of tool steel. The cost of the steel and its heat treatment amount generally to less than a quarter of the total cost of the whole tool. A wrong choice of steel or faulty heat treatment may give rise to serious disruption of production and higher costs. It is the author's hope that the contents of this section will increase his readers' knowledge of tool steels and their heat treatment.

In the text, tool steels are designated only by the type letter and numeral as used in the USA and the UK for standardized tool steels, e.g. H 13, O 1. These designations are so well known by steel consumers all over the world that no qualifying institutional designations are necessary. Steels for which there are no AISI or BS specifications are designated according to DIN or SIS standards.

6.1.1 Carbon steels and vanadium-alloyed steels

The hardening of these steels, which are made with carbon contents between 0.80% and 1.20%, is quite straightforward. Since the rate of carbide dissolution proceeds rapidly, the holding time, as a consequence, is short and therefore the heating of small tools can often take place without any extra precautions.
| Hardening treatment | Temperature range °C | Cooling medium | Dimension mm | Hardness Rockwell C after hardening and tempering (experimental results) | Tempering temperature °C | 100 | 150 | 200 | 250 | 300 | 400 | 500 | 600 | AISI (B) |
|---------------------|----------------------|----------------|--------------|---------------------------------------------------------------------|--------------------------|-----|-----|-----|-----|-----|-----|-----|-----|-----|---------|
| 770-800             | Water                | 25             | 770          | 67          | 65          | 63          | 59          | 56          | 47          | 38          | -          | W1          |       |     |     |     |       |     |     |     |     |         |
| 770-800             | Water                | 25             | 790          | 66          | 64          | 63          | 60          | 57          | 48          | 38          | (P171)      |             |       |     |     |     |       |     |     |     |     |         |
| 800-840             | Oil                  | 25             | 800          | 64          | 63          | 62          | 60          | 58          | 53          | 47          | 40          | G1          |             |       |     |     |     |       |     |     |     |     |         |
| 850-890             | Oil                  | 25             | 860          | 66          | 64          | 63          | 62          | 61          | 58          | 50          | 42          | (SR 1855)   |             |       |     |     |     |       |     |     |     |     |         |
| 810-840             | Oil                  | 100            | 820          | 58          | 58          | 57          | 55          | 53          | 48          | 44          | 40          | (HR 1243)   |             |       |     |     |     |       |     |     |     |     |         |
| 830-850             | Air                  | 100            | 840          | 50          | 50          | 49          | 48          | 47          | 44          | 41          | 38          |             |             |       |     |     |     |       |     |     |     |     |         |
| 860-880             | Water                | 20             | 880          | 62          | 60          | 58          | 57          | 56          | 53          | 49          | 45          |             |             |       |     |     |     |       |     |     |     |     |         |
| 880-920             | Oil                  | 20             | 900          | 59          | 58          | 57          | 57          | 56          | 53          | 49          | 45          |             |             |       |     |     |     |       |     |     |     |     |         |
| 950-980             | Oil,                  | 25             | 960          | 64          | 63          | 61          | 60          | 58          | 58          | 57          | 52          | A2          |             |       |     |     |     |       |     |     |     |     |         |
| 950-980             | salt bath            | 25             | 1000         | 63          | 63          | 62          | 60          | 59          | 59          | 58          | 50          | D2          |             |       |     |     |     |       |     |     |     |     |         |
| 950-980             | or air               | 25             | 970          | 64          | 63          | 62          | 61          | 60          | 57          | 54          | 50          | D3          |             |       |     |     |     |       |     |     |     |     |         |
| 960-1000            | or air               | 25             | 980          | 64          | 63          | 62          | 61          | 60          | 59          | 59          | 50          | D6          |             |       |     |     |     |       |     |     |     |     |         |
| 800-910             | Oil                  | 100            | 900          | 52          | 52          | 52          | 52          | 51          | 47          | 39          | (R O 8155)  |             |       |     |     |     |       |     |     |     |     |         |
| 1000-1010           | Oil                  | 50             | 1025         | 54          | 55          | 53          | 51          | 48          | 44          | 38          | H11         |             |       |     |     |     |       |     |     |     |     |         |
| 1000-1010           | salt bath            | 50             | 1040         | 51          | 52          | 52          | 51          | 49          | 46          | 48          | (Q R 45)    |             |       |     |     |     |       |     |     |     |     |         |
| 1050-1110           | or air               | 50             | 1100         | 52          | 53          | 54          | 53          | 52          | 47          | 44          | (Q R T 71)  |             |       |     |     |     |       |     |     |     |     |         |
| 1260-1280           | Oil                  | 25             | 1270         | 64          | 64          | 65          | 65          | 64          | 63          | 58          | 53          | T 1         |             |       |     |     |     |       |     |     |     |     |         |
| 1260-1280           | salt bath            | 25             | 1210         | 65          | 64          | 66          | 65          | 64          | 61          | 57          | M 2         |             |             |       |     |     |     |       |     |     |     |     |         |
| 1260-1280           | or air               | 25             | 1210         | 65          | 64          | 66          | 65          | 63          | 59          | 54          | M 7         |             |             |       |     |     |     |       |     |     |     |     |         |
| 1160-1200           | air                  | 25             | 1180         | 62          | 63          | 66          | 67          | 66          | 65          | 63          | M 42        |             |             |       |     |     |     |       |     |     |     |     |         |

FIGURE 6.1: The successful heat treatment of an expensive tool steel via the proper cooling. The quenching temperature is about 770°C. It is a critical and very intricate process. Quick quenching produces a non-metallic compound (cementite), which may cause quench cracking. Proper quenching and tempering procedures can significantly reduce the risk of quench cracking. The depth of the hardened layer can be controlled by adjusting the quenching temperature and time. For high-speed steels, the depth of hardening is typically measured using a linear scale. In industrial practice, the depth of hardening is usually measured using a linear scale, which is standardized by various national and international standards. The depth of hardening can be influenced by various factors, such as the composition of the steel, the quenching medium, and the tempering conditions. The depth of hardening is crucial for the performance and longevity of high-speed steels.
266 HEAT TREATMENT—SPECIAL

hardening is considerably greater. Owing to the high content of V the steel remains fine-grained even when hardened from exceptionally high temperatures. The very considerable toughness inherent in a plain-carbon steel, due to its shallow-hardening properties, is forfeited if the tool through-hardens locally at some sections because the cross-sectional area there is too small. For shearing tools or small tools generally, such as scissors, knives or letter die punches, which are not subjected to heavy impact blows, this drawback is of less importance. Tools operating under heavy blows, e.g. upsetting dies for cold-heading of bolts, must not be through-hardened. Coining and striking punches are other examples of carbon tool steels that require high wear resistance. Such tools may also be subjected to bending stresses and should therefore not be through-hardened. The tempering temperature normally used for tools belonging to this group lies in the range 170–220°C, the hardness being generally about 60–64 HRC. Representative examples of tools made from grade W 1 are shown in Figure 6.4.

6.1.2 Low-alloy cold-work steels

Only two steels have been chosen from this group for discussion, viz. grade O 1 (BR 1733) and S15 2092 (SR 1855).

When carbon steel is used for punching dies or cold hobbing tools the dimensions of the tool are bound by a ruling section that is determined by the load on the tool. A punch or a die, made from carbon steel, having a diameter of, say, 50 mm, will show rather poor resistance to sinking on account of the shallow depth of hardening. Should this resistance not suffice, another steel will have to be chosen, in this case grade O 1 or S15 2092. From the point of view of heat treatment, these two steels differ somewhat since their hardening temperatures are different. Steel S15 2092 requires 850–890°C whereas grade O 1 requires 800–840°C. Owing to its lower
hardening temperature, O 1 has somewhat greater dimensional stability. This property makes it a first choice for blanking dies and other tools requiring a high degree of dimensional stability (Figure 6.5).

In both steels the depth of hardening decreases by roughly the same amount as the thickness of the section increases. In the diagram in Figure 6.6 the hardening temperature was raised as the cross-sectional area increased in order to increase the hardenability of the steel. Tools having diameters greater than about 80 mm or equivalent sections in flat dimensions are difficult to harden to full hardness if there are re-entrant corners. For such

<table>
<thead>
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<tr>
<td>0</td>
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<td>10</td>
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</tr>
<tr>
<td>45</td>
<td>25</td>
</tr>
<tr>
<td>50</td>
<td>20</td>
</tr>
</tbody>
</table>

Figure 6.6. Curves showing depth of hardening for steel O 1 (Bofors RT 1731). Specimen 25 mm diameter oil quenched from 800 °C. Specimen 50 mm diameter oil quenched from 820 °C. Specimen 100 mm diameter oil quenched from 840 °C.

The difference in hardenability between the two steels has been extensively discussed in Chapter 4. The above-cited example is to be regarded as a practical assertion of the possibility of estimating the depth of hardening from the Jominy diagram. However, it should be emphasized once again that a 'contour-hardened' tool is tougher than a through-hardened one.

Figure 6.8 shows a section through a 'contour-hardened' Pilger roll. As a rule both steels are oil quenched. For heavy sections, e.g. dimensions greater than 100 mm in diameter, it is best to use water quenching when dealing with SIS 2092 (Bofors SR 1855). When the surface temperature of the steel has fallen to between 400 °C and 300 °C the water quenching is interrupted by transferring the tool to an oil bath.

The tempering temperature for both steels is generally in the range 170–200 °C which gives a hardness of more than 60 HRC. As can be seen from Figure 6.9, SIS 2092 has a greater resistance to tempering than grade O 1.
On being tempered in the range 250–350 °C the steel suffers a reduction in its impact strength which in turn increases the risk of chipping. For this reason tools that are subjected to impact stresses should not be tempered in this temperature range. The higher impact strength manifested after tempering at 170–200 °C is due to the presence of retained austenite, viz. about 10%.

This soft retained austenite can accommodate impact stresses better than the harder constituents. Retained austenite is decomposed when it is tempered at about 300 °C. For the steel DIN 105 WCr 5 (1.05% C; 1% Mn; 1% Cr; 1.2% W) which closely approximates to grade O 1, the relation between tempering temperature and torsional fracture strength, torsional yield strength and energy of plastic deformation in torsion is shown in Figure 6.10. If a specimen is subzero treated before annealing to 200 °C it is found that the work required for plastic deformation (toughness) is less and the torsional yield strength is higher than if the specimen were not subzero treated.

If, during service, some areas of the tool have to support excessive pressures, for example the shearing edge of circular slitting knives (see Figure 6.11), retained austenite may be transformed to martensite, with spalling at the edge as a result. Should this occur, tempering at 300–400 °C is recommended. After such a treatment the hardness of SIS 2092 still remains around 60 HRC. Since the wear resistance of SIS 2092 is as much as 25%, greater than that of grade O 1, the former is very popular when a wear-resisting steel is required that can give a better performance than grade O 1. Compared with this steel, SIS 2092 has been shown to have a considerably longer service life, particularly as a drawing die steel. Another interesting application is as...
cane-slitting tools (see Figure 6.12). The requirements of this type of tool are both high wear resistance and toughness in its thin walls. Of all the steels tested the best results were obtained with S1S 2092 (Bofors SR 1855).

In recent years S1S 2092 has increasingly been used for so-called Pilger rolls which are in part made as rings and in part as dies. Figure 6.13 shows one of the world's largest Pilger rolls, designed for cold-rolling 10 in tubes. The only steel suitable for this tool was S1S 2092. Another field of application is for what are known as Yoder rolls. A sketch showing the principle of operation and tube manufacture is shown in Figure 6.14. For this mill unit, the wear resistance of rolls made from S1S 2092 has shown itself to be on a par with that of grade D 2, in fact in some instances it has outlasted this grade.
6.1.3 Low-alloy cold-work and hot-work steels

The hardenability of SIS 2550 (Bofors HR0 1243) is considerably better than that of either of the two steels just discussed. SIS 2550 is air hardening in fairly heavy sections, which is of advantage where dimensional stability is concerned. Due to the lower carbon content the toughness is greater than that of the previously mentioned steels. When used for cold-work tools, the steel is tempered at 170-250 °C, the resulting hardness then being 55-58 HRC. With regard to impact strength this steel, too, is susceptible to tempering treatments around 300 °C (see Figure 6.15). In Figure 6.16 is shown a tool that failed during use on account of its having been tempered at 300 °C. A similar tool that was tempered at 200 °C functioned satisfactorily.

SIS 2550, after hardening and tempering at 200-250 °C possesses very high tensile strength and good impact strength. The values given below have been obtained on tensile test specimens that were oil quenched from 830 °C and tempered at 250 °C.

\[
\begin{align*}
R_{p0.2} &= 140 \text{ kp/mm}^2 (1370 \text{ N/mm}^2) & A_j &= 8\% \\
R_{p0.5} &= 155 \text{ kp/mm}^2 (1520 \text{ N/mm}^2) & Z &= 33\% \\
R_{p0.8} &= 166 \text{ kp/mm}^2 (1630 \text{ N/mm}^2) & \text{HRC} &= 54 \\
R_{	ext{im}} &= 208 \text{ kp/mm}^2 (2000 \text{ N/mm}^2) &
\end{align*}
\]

Such favourable mechanical properties make the steel suitable for tools subjected to large static and dynamic forces. Some typical applications are dies for table-wafer, shear blades for heavy plate, dies for plastic moulds for which is required a steel possessing a high degree of dimensional stability and excellent polishability.

SIS 2550 is also used for hot-work tools working at moderate temperatures, e.g. drop-forging dies. Such tools are tempered between 400 °C and 600 °C, the exact temperature depending on the hardness required and the working temperature of the tool. For working temperatures above approximately 400 °C the hardness of the steel falls relatively quickly; this is shown in Figure 6.17. If higher working temperatures are involved it is...
recommended to use the special hot-work steels that are discussed in later sections.

Grade S 1 (Bofors R T O 912) has both high wear resistance and high impact strength. The hardenability is inferior to that of the Cr–Ni–Mo steel SIS 2550. This implies that for dimensions greater than 50 mm in diameter, this steel is contour-hardening which, in fact, further increases its toughness. The normal hardening temperature is about 900 °C but it may be raised to 950 °C without any risk of grain growth being incurred. If a hardness higher than 50 HRC is required in dimensions up to about 60 mm in diameter the steel should be quenched in oil. For heavier dimensions a combined water–oil quenching procedure may be necessary. As regards the tempering temperature ranges the same principles as are applicable to the foregoing steels are valid in this case also.

Of the many cold-work applications for tool steel, special mention should be made of the cold punching of plate having a thickness greater than about 3 mm. If a plate of increasing thickness is being punched and consequently the thickness measurement of the plate is approaching the diameter of the hole, the punches used show an increasing tendency to break if they are made from, for example, grades O 1, A 2 or D 2. For this type of punching work, grade S 1 has been shown to possess the best combination of toughness and wear resistance. A suitable hardness is 56–58 HRC. Wear resistance is further increased if, during the course of the hardening treatment, the tools are heated for some 20 min in a cyanide bath. After this treatment no further finishing is required; at the most a very light finish grinding is permissible. Another example is the use of this steel as the impact hammer in nail guns, used for driving nails into concrete.

Owing to its high toughness in comparatively large dimensions, grade S 1 can successfully be used for table-ware dies which, depending on their dimensions should either be quenched in oil or be heat treated according to the combined oil–water quenching procedure previously mentioned. Another

field of application is shear blades for cold shearing of heavy plate. Owing to its rather good resistance to tempering, grade S 1 may also be used for hot shears, a suitable hardness for this latter use being about 45 HRC.

In the field of hot-work, grade S 1 has been superseded by other grades, e.g. H 13. However, mention should be made of an interesting and successful field of application for grade S 1 (Bofors R T O 912), viz. as chisels used in connexion with the electrolytic reduction of aluminium from bauxite. The function of the chisels is to break up the hard alumina-containing crust which forms on the metal bath. During use the chisels also come into contact with the bath itself and are thus subjected to both high temperatures and impact stresses. A suitable chisel hardness is about 350 HB. Figure 6.18 shows a worn-out chisel.

6.1.4 High-alloy cold-work steels

This group comprises the following four steels:

<table>
<thead>
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<th>Grade</th>
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</tr>
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<tbody>
<tr>
<td>A 2</td>
<td>Grade D 2</td>
</tr>
<tr>
<td>Grade D 2</td>
<td>(Bofors R P 57)</td>
</tr>
<tr>
<td>Grade D 3</td>
<td>(Bofors R 60)</td>
</tr>
<tr>
<td>Grade D 6</td>
<td>(Bofors R T 60)</td>
</tr>
</tbody>
</table>

D 6 is no longer an AISI standard grade but the steel is standardized in a number of countries other than the USA. These steels, except grade D 3, are air-hardening in dimensions up to ~100 mm in diameter. Grade D 3 air hardens up to about 50 mm in diameter. Grade D 6 (Bofors R T 60) has a hardenability superior to that of grade D 3, partly owing to its higher content of Mn and partly to the presence of W. The results of air-hardening trials with 12% Cr steels of various compositions are shown in Figure 6.19. The degree of hardenability can be controlled by means of the hardening

![Figure 6.18: Worn-out chisel made from steel S 1 (Bofors R T O 912). Chisel used in connexion with aluminium smelting](image)

![Figure 6.19: Hardenability of steel D 3 (Bofors R 60) containing various Mn contents, and steel D 6 (Bofors R T 60). Steel specimen 100 mm (4 in) diameter hardened from 1000 °C in still air](image)
Heat Treatment—Special

Temperature since these steels are very rich in carbides. Since progressively higher hardening temperatures bring about increasing grain growth and increasing amounts of retained austenite in the hardened steel, the upper temperature limit of the hardening range is determined by how much grain growth and retained austenite, respectively, is acceptable. Grade A 2 is more susceptible to grain growth than either of the grades D 2 or D 6 (see Figure 6.20).

Figure 6.21 shows the hardness obtained in specimens, 30 mm in diameter, on quenching in oil from various temperatures after a holding time of 30 min. One series of specimens was subzero treated at -30°C, another at -180°C. The conclusions reached from the series of experiments described above were that maximum hardness was obtained after quenching from the following temperatures:

- A 2: 950°C
- D 2: 1010°C
- D 6: 980°C

These temperatures are also the ordinary hardening temperatures, respectively, for the steel grades listed above. If the steels are treated at subzero temperatures after hardening, maximum hardness is obtained when the hardening temperature is somewhat higher. Figure 6.22 shows how the amount of retained austenite in grade D 2 increases with increasing hardening temperature! As has been pointed out in a previous section, an air-hardened specimen contains a larger amount of retained austenite than other, more
Figure 5.23. Steel A 2 (Before B.O.P 21). Influence of tempering temperature on hardness of specimens (1) after oil quenching, from various temperatures, and (2) after oil quenching followed by subzero treatments at −80 and −100°C, respectively. Specimen 30 mm in diameter. Holding times: at hardening temperatures, 30 min; at tempering temperatures, 1 h.

Figure 5.24. Steel D 2 (Before B.O.P 37). Influence of tempering temperature on hardness of specimens (1) after oil quenching from various temperatures, and (2) after oil quenching followed by subzero treatments at −40 and −180°C, respectively. Specimen 30 mm in diameter. Holding times: at hardening temperatures, 30 min; at tempering temperatures, 1 h.
drastically hardened, specimens owing to the stabilizing effect produced in the austenite as it slowly cools through the range of martensite formation. By far the greater number of tools made from these steels are tempered at 180–200 °C, which gives them a hardness of 61–63 HRC, provided that a normal hardening temperature has been used and that the section is not too heavy. The relatively high toughness shown by the steels is due to the presence of retained austenite, which usually amounts to some 20%. The hardness falls as the tempering temperature is raised but if the hardening temperature has been sufficiently high the tempering may produce a secondary hardening peak in the tempering curve. The temperature at which this peak occurs increases as the hardening temperature is raised. The results obtained on tempering the specimens that were hardened as described in Figure 6.21 are shown in Figures 6.23, 6.24 and 6.25.

A similar investigation was carried out by Rapatz who used a steel containing 1.75% C, 0.22% Mn and 13% Cr. The steel specimens were oil quenched from temperatures ranging from 900 to 1300 °C. One test series was subzero treated in liquid air. After tempering at 450 °C, 500 °C, 550 °C...

**Figure 6.25. Steel 25 C (Refors 5760). Influence of tempering temperature on hardness of specimens (1) after oil quenching from various temperatures, and (2) after oil quenching followed by subzero treatments at −80 °C, respectively. Specimen 30 mm diameter. Holding time: at hardening temperatures, 30 min; at tempering temperatures, 1 h.**
and 600 °C the hardness and amount of retained austenite were determined (see Figure 6.26).

Tempering at 450 °C. The amount of retained austenite is the same as after quenching (see Figure 6.22). When a subzero treatment is introduced before tempering, the amount of retained austenite is reduced, which results in a hardness increase.

Tempering at 500 °C. The amount of retained austenite in the specimen quenched from 1000 °C is reduced from 30% to 10%. There is also some decomposition of the austenite that is retained in specimens quenched from temperatures approaching 1200 °C. Quenching from 1100 °C produces maximum hardness in the specimen. A subzero treatment preceding the tempering does not influence the hardness until quenching temperatures of 1150 °C and higher are used.

Tempering at 550 °C. All the retained austenite resulting from the 1000 °C quench has been decomposed. The maximum feasible hardness, viz. 62 HRC, is obtained by stepping up the hardening temperature to 1175 °C, the retained austenite content then being about 50% after quenching. If a subzero treatment precedes the tempering, the maximum hardness is only 57 HRC.

Tempering at 600 °C. The retained austenite is completely decomposed in all the specimens. A quenching temperature of 1150 °C gives a hardness maximum when the subzero treatment is omitted before the tempering.

The high hardness, around 64 HRC, found in light sections after quenching and tempering at 180 °C is seldom obtainable in heavy sections, i.e. in sections more than 75 mm in diameter, when quenched in oil or in a marten tempering bath. This is because the amount of retained austenite increases as the cooling rate decreases. Figure 6.27 shows diagrammatically the expected hardnesses in sections ranging from 10 to 150 mm in diameter when they have been quenched in an oil or marten tempering bath and then tempered. The upper line refers to the smallest dimensions, the lower line, the largest.

If grades D 2 and D 6 are quenched from a high hardening temperature, say about 1050 °C, and then tempered around 525 °C, the resulting hardness is more or less independent of the dimensions of the specimen (see Figures 6.28 and 6.29). The advantage of using such a treatment is that the greater part of the retained austenite is decomposed during tempering. Hence high working pressures may be applied to the steel without risking the manifestation of the adverse effects that result when untempered martensite is formed. Since the steel is heated to a high hardening temperature a large amount of carbide dissolution takes place and this increases the homogeneity of the steel as well as its toughness. Owing to the increased resistance to tempering imparted to the steels on their being quenched from high hardening temperatures they can also be used as hot-work steels, provided that the requirements of impact strength are not too severe. Their improved resistance to tempering, just mentioned, is an additional advantage if these steels are to be nitrided. See following chapters.

Figures for impact strength, as recorded from the Charpy U test, are very low for these high-alloy steels. Moreover, different heat treatments produce

![Figure 6.27. Tempering curves showing interdependence of time and temperature for: Steel A 2 (Before ROP 27), Hardening temperature 960 °C. Steel D 2 (Before ROP 27), Hardening temperature 1000 °C. Steel D 6 (Before RT 66), Hardening temperature 900 °C]
Figure 6.28. Tempering curves showing interdependence of time and temperature for steel D2 (Bosfor ROP 57), quenched from 1030°C in still salt bath at 225°C. Diameter of specimen 30 mm

Figure 6.29. Tempering curves showing interdependence of time and temperature for steel D2 (Bosfor ROP 57), quenched from 1030°C in still salt bath at 225°C. Diameter of specimen 150 mm

Figure 6.30. Influence of tempering temperature on hardness and impact strength of steel A2 (Bosfor ROP 21). Oil quenched from 976°C
only insignificant differences in impact values. These facts are brought out in Figure 6.30, which applies to grade A 2. Consequently, tests have also been carried out with unnotched specimens, 8 mm in diameter x 55 mm, held in supports 30 mm apart. From Figures 6.31 to 6.33 it may be seen that an impact strength maximum is obtained after a tempering treatment at 300–400 °C and a minimum at around 500 °C. If there is a possibility that tool failure can occur on account of heavy impact blows, this risk may be anticipated by avoiding a tempering treatment for the tool concerned in the region around 500 °C. However, actual trials with punching tools, tempered at 500 °C, have demonstrated that this temperature range need not be regarded as particularly risky for grades D 2 and D 6.

![Graph of Hardness vs Tempering Temperature](image1)

Figure 6.31. Influence of tempering temperature on hardness and impact strength of steel D 6 (Bofors RT 60), oil quenched from 950 °C. Unnotched impact test specimen, 8 mm diameter x 55 mm

The probable explanation of the relatively high impact values obtained in the 200–400 °C range is that a favourable intermix of retained austenite and tempered martensite is present in the steel. This is borne out by the fact that if specimens are subzero treated before being tempered at temperatures up to 450 °C (see Figure 6.33) their impact strength is consistently lower than if they were not subzero treated.

Trials with punching and blanking tools made from various grades of steel have shown, according to Bühler, Pollmar and Rose that grade D 3 (DIN X 210 Cr 12) suffers 40% more wear than grade D 6 (X 210 CrW 12). Figure 6.34 contains results from this investigation. Steel 105 WCr 6 which corresponds to grade O 1 with an addition of 1% W, suffers about twice as much wear as the steels containing 12% Cr. The high-alloy Cr steels are used for blanking tools and drawing tools when large production series are concerned. A blanking tool for 1 mm mild steel sheet, if made from

![Graph of Hardness vs Tempering Temperature](image2)

Figure 6.32. Influence of tempering temperature on hardness and impact strength of steel A 2 (Bofors ROP 21), oil quenched from 970 °C. Unnotched impact test specimen, 8 mm diameter x 55 mm

![Graph of Hardness vs Tempering Temperature](image3)

Figure 6.33. Influence of tempering temperature on hardness and impact strength of steel A 2 (Bofors ROP 21), oil quenched from 970 °C and subzero treated at −40 °C. Unnotched impact test specimen, 8 mm x 55 mm
HEAT TREATMENT—SPECIAL

grade D 6 (Bofors RT 60) can make about 100,000 stampings before being redressed. For grade O 1 the corresponding number of stampings is about 25,000. However, the hardness of the tool must be suited to the thickness of the sheet. The curves in the diagram (Figure 6.35) may be of help in choosing the best hardness.

To a large extent the steels are used for tools for shaping and forming. The so-called Sendzimir rolls, i.e. rolls forming a cluster-roll mill, constitute an example of such use. The hardness of these rolls is in the range 58–64 HRC. Another example is mandrels for tube rolling by means of Pilger rolls. A suitable hardness for mandrels in grade D 2 is 52–54 HRC. Figure 6.36 illustrates such mandrels. As mentioned above, the steels may also be used for hot work. This applies in the first instance to grade D 2 which finds an application as forging dies for stainless-steel knives. Valve seats for motorcar internal combustion engines are best made of grade D 6 steel which has been found to be superior to D 3 for this purpose.

6.1.5 Hot-work steels

Among the hot-work tool steels may also be included, as mentioned previously, the Cr-Ni-Mo steel S15 2550. A similar steel, much used for drop-forging dies, is the slightly lower-alloy Cr-Ni-Mo-V steel, designated DIN 56 NiCrMoV 7 to which, however, there is no equivalent grade in either the BS or AISI standards. This steel is normally supplied heat treated to a hardness of 350 HB. The remaining hot-work tool steels included in Table 6.1 on p. 262 have Cr and Mo as their principal alloying elements.

BOFORS RO8155

This steel has no direct equivalent among the standardized steels. It owes its existence primarily to its serving as a substitute for the more expensive grade
Figure 6.37. Curves showing depth of Bofors RO 8155. Specimen 200 mm diameter, oil quenched from 900°C. Tempered at 400, 500, and 600°C for 1 h.

Figure 6.38. Curves showing various mechanical properties of Bofors RO 8155, oil quenched from 900°C and tempered. Specimen 200 mm diameter. Location of test-piece: 68 radius.

292 HEAT TREATMENT—SPECIAL

H13 which it has successfully replaced in a number of instances. The steel is normally supplied heat treated to a hardness of 320-360 HRC. As may be seen from the curves in Figure 6.37 it has excellent hardenability, i.e., on oil quenching, the steel is practically through-hardening in dimensions up to 200 mm in diameter. The mechanical properties of the steel, both as oil quenched and tempered and as air hardened and tempered, respectively, are shown in Figure 6.38. As may be seen from the curves in Figure 6.39 the steel has high strength at elevated temperatures, i.e., up to about 500°C.

Figure 6.39. Mechanical properties at elevated temperatures of Bofors RO 8155. Air hardened from 900°C and tempered to various hardnesses.

One of the uses of the steel is for drop-forging dies (see Figure 6.40). In this capacity it often has a longer service life than grade H13. This is probably due to the superior thermal conductivity of RO 8155. Another use to which the steel is put is as material for supporting dies in the extrusion of aluminium. The hardness should be 44–47 HRC. The steel is well suited to nitriding.
The strength at elevated temperatures increases as the hardening temperature is raised but this increase in strength takes place at the expense of the toughness. This fact is illustrated for steel H 13 in Figure 6.41 in which the elongation and reduction of area are the properties taken as a measure of the toughness (actually ductility). The higher the hardening temperature the more susceptible is the steel to grain growth. A coarse-grained steel is less tough than a fine-grained one. The recommended holding times that were given in a previous section must be adhered to (see p. 211). To illustrate this point, Figure 6.42 shows the grain size as a function of the hardening temperature and time of holding. Owing to its high V-content, grade H 13 is not unduly susceptible to grain growth. As has been pointed out before, the grain size number after hardening should be at least 7 A.S.T.M. When the heating is carried out in a neutral salt bath the holding time is easily kept under control. If the tool is pack-heated it is difficult to
judge the actual holding time and therefore when this method of heating is used the lower temperature of the hardening range is recommended.

As the hardening temperature increases, more and more carbides go into solution. For grade H 13 the consequences of this are that the higher the hardening temperature the sooner is the start of the carbide precipitation

that precedes the pearlite and bainite formation when quenching the steel. This fact is brought out by the continuous-cooling-transformation (CCT) diagram reproduced in Figure 6.43. These diagrams as well as the one in Figure 6.44 date from a period earlier than the one from which the diagram shown previously in Figure 4.44 originates and therefore they differ somewhat from the latter. However, from an educational point of view the older diagrams are fully acceptable.

Impact strength determinations carried out on specimens of steel H 13, after having been hardened from various temperatures, show that toughness decreases with increasing hardening temperature. In this instance grain growth cannot have been the only factor concerned since the steel is relatively insensitive to grain growth. Tests involving different rates of cooling have shown that the lowest cooling rate resulted in the lowest impact strength. It has also been demonstrated that there is some connexion between low impact values and the presence of precipitated grain-boundary carbides in the steel. Hence, if the highest possible impact strength is to be imparted to grade H 13, a moderately high hardening temperature combined with a high cooling rate should be used.

Figure 6.42. Influence of hardening temperature and holding time on grain size of steel H 13 (Before ROP 19)

Figure 6.43. Continuous-cooling-transformation (CCT) diagrams for grade H 13, austenized at 1030 and at 1100 °C, respectively, for 15 min (from Atlas zur Wärmebehandlung der Stähle)

A = Austenitic, B = Bainitic, K = Carbide, M = Martensite, P = Pearlite,
Y = Retained austenite, P = Pearlite, M = Start of martensite formation
From the point of view of carbide precipitation Figures 6.44a, b and c indicate which dimensions are the crucial ones when the steel, after having been austenitized at 1030°C for 15 min, is cooled in different cooling media.

When a 100 mm diameter bar is quenched in oil the curve representing the rate of cooling at the centre, as traced in the diagram, just touches the curve of incipient carbide precipitation. This state of affairs will hardly cause any deterioration in the impact strength of the steel. However, if the above-mentioned bar is quenched in a salt bath, carbides will be precipitated both in the core and the surface layers. With air cooling it is not possible to avoid carbide precipitation in any of the dimensions indicated in the diagrams. Hence, if high impact strength is required, a rapid rate of cooling is necessary. Since oil quenching may result in more distortion than salt-bath quenching or air cooling (see Chapter 7) it is recommended that oil quenching be resorted to only when the requirements of dimensional stability are not of primary importance.

In practice it has been found that with regard to large and complicated dies made from grade H 13, such as die-casting dies for aluminium, the longest service life is obtained if they have been martenpered. Figure 6.45 shows such a die in the process of being martenpered.

Figure 6.44. a, b and c. Continuous-cooling-transformation (CCT) diagrams for grade H 13, austenitized at 1030°C for 15 min. Cooling curves for specimens of various dimensions and different cooling media superimposed on diagrams (b and c overlap)
Figure 6.45. Martempering of die-casting die for aluminium. Die made from steel H 13 (Bofors ROP 19)

Figure 6.46. Hardness and impact strength of steel H 13 (Bofors ROP 19) as functions of tempering temperature

Figure 6.47. Tempering curves showing interdependence of time and temperature for steel H 13 (Bofors ROP 19). Oil quenched or air cooled from 1050°C
On tempering hot-work steels in the neighbourhood of 500°C a secondary hardening is obtained (see Figure 6.46). Practical experience has shown that since there is a reduction in impact strength on tempering the steel at this temperature, it is better to forgo maximum hardness and only in exceptional cases to resort to hardnesses greater than 50–52 HRC. However, since wear resistance increases with hardness, the highest attainable hardness is usually aimed at. A general rule states that the smaller the cross-sectional area of the tool the greater is the obtainable hardness. In an analogous way a tool of simple design may, without jeopardizing its functioning, be given a hardness near the upper limit of the recommended hardness range.

The significant influence of time when these steels are tempered above 500°C has been described in detail in the preceding chapter. For the sake of completeness the tempering diagram for grade H 13 is reproduced in this chapter also (see Figure 6.47). In order to estimate a suitable hardness for hot-work tools subjected to known pressures and temperatures it is assumed that the yield strength of the steel at the temperature concerned should be greater than the working stress. The diagrams in Figure 6.48 show the mechanical properties at elevated temperatures of grade H 13 after specimens have been air hardened and tempered to various hardnesses. For working temperatures up to 500°C the initial hardness has a very marked influence on the tensile strength and yield point. As the temperature increases, this influence diminishes and at 700°C the values are the same for all the specimens.

Despite the high ultimate tensile strength of this steel at room temperature the figures for elongation and reduction of area are also very high. However, as the dimensions of the steel increase, the impact strength decreases, particularly in the transverse direction. By remelting the steel, making use of the VAR or ESR consumable electrode process (see Chapter 3) the transverse impact strength is improved but in many cases, not enough. For this reason the Bofors Company since the middle of the 1960s has been applying a heat
treatment process called 'structure treating'. This treatment consists of an austenitization followed by cooling prior to the annealing treatment proper. The effect of this treatment on the structure is shown in Figures 6.49a and b. The values illustrated by the bar chart in Figure 6.50 are the result of tests carried out on specimens taken from 200 mm diameter bar stock that had been processed according to different methods.

The structure treatment has been found to be particularly beneficial to H 13 steel used for die-casting dies for aluminium. The dies shown in Figure 6.51, which is an example taken from actual practice, show the importance of a correct initial structure and hardness of a forging die made from grade H 13. The die, marked a, cracked under the very first forging blow. Its hardness was 43 HRC which is normal for a die of this size. The microstructure was much inferior to that shown in Figure 6.49a. The die, marked b, cracked after 333 blows. In this instance the microstructure was ideal. The hardness was 53.5 HRC which is too high for a die of this type. The die, marked c, was removed from service owing to its beginning to show heat checking after 18,859 blows. The structure was normal and the hardness 45 HRC.

Further examples of hot-work dies and tools made from grade H 13 are shown in Figures 6.52 and 6.53.

The excellent mechanical properties of grade H 13 have resulted in the
use of this steel for highly stressed structural members operating at room temperature. This field of application is discussed in a later section, entitled ‘High-duty steels’. Like the other hot-work steels mentioned above, grade H 13 can be nitrided with excellent results. Data concerning this process are given in Section 6.4.

**BOFORS QRO 45**

This steel has been developed at the Bofors Company and is patented in several countries. The first patent application was lodged in 1956. The composition of the steel corresponds to BS BH 10A. It is a relatively low-alloy steel: the low-alloy content being the result of stringently optimizing the composition with reference to the overall properties of the steel. It possesses good hot-wearing properties, low susceptibility to fatigue cracking and high elevated-temperature strength. In the great majority of instances QRO 45 serves as a wholly satisfactory substitute for the previously very popular 5% and 9% tungsten steels, respectively. The hardening temperature is the same as for grade H 13, viz 1000-1050 °C; 20–30 min is a satisfactory holding time at the hardening temperature. Longer times or temperatures higher than 1050 °C tend to make the steel coarse-grained (see Figure 6.54). Compared with grade H 13, QRO 45 has a greatly superior resistance to tempering, which can be seen in Figure 6.55. At temperatures around 600 °C and above, the elevated-temperature strength of QRO 45 is higher than that of grade H 13 (see Figure 6.56).

Typical fields of application for QRO 45 are dies for die casting and hot forging of brass, mandrels and dies for the extrusion of copper, tools for the hot-upsetting of steel bolts and nuts (Figure 6.57). For this last example the wear resistance of the steel can be greatly improved by nitriding or by case hardening (see Sections 6.3 and 6.4). Owing to the low hardness of QRO 45
Figure 6.55. Tempering curves showing interdependence of time and temperature for Bofors 450. Oil or air hardened from 1050°C.
as annealed, cold hobbing is a feasible alternative process for shaping tools of suitable design (see Figure 6.58).

**BOFORS QRT 71**

This steel has also been developed by the Bofors Company. It is used for the same purposes as QRO 45 when the latter steel has had inadequate elevated temperature strength. The hardening temperature for QRT 71 is 1050–1100 °C. Figure 6.59 shows the influence of hardening temperature and holding time on the grain size. When hardening, a temperature of 1100 °C should not be exceeded. The holding time is the same as that for QRO 45, viz. 20–30 min. Figures 6.61 and 6.62 show that for this steel both resistance to tempering and elevated-temperature strength are greater than for QRO 45. QRT 71 has given very good performance as dies for the extrusion of copper
Figure 6.61. Tempering curves showing interdependence of time and temperature for Bifors QRT 71. Oil or air hardened from 1000°C.

Figure 6.62. Mechanical properties at elevated temperatures of Bifors QRT 71. Air hardened and tempered to various hardnesses.
6.1.6 High-speed steels

The first high-speed steel, which was of the T 1 type, was developed at the beginning of this century. This steel, designated 18-4-1, was the precursor of modern high-speed steel. Grade T 1 has for many years kept its position and it is still used to some extent. By adding some 5–10% Cr and simultaneously increasing the C and V contents the wear resistance is increased. Common to all 18-4-1 steels is their high hardening temperature, viz. 1260–1280°C. Any danger of overheating the steels by using a hardening temperature of 1280°C is therefore out of the question.

As was the case for hot-work steels it has been possible to replace W in the T 1 grade by Mo. This has led to the development of the M 2 type (6-5-4-2) which, for most purposes, can replace T 1. Yet another variant is M 7 which has a higher content of Mo but less W than M 2. For some applications M 7 is said to possess greater toughness and wear resistance that M 2. Like T 1, both M 2 and M 7 can be alloyed with Co which gives them increased hot wear-resistance. Such a variant of M 7 is designated M 42. For the M steels a suitable hardening temperature is 1200–1220°C. Sometimes 1230°C is given as the maximum hardening temperature and under no circumstances should this temperature be exceeded.

Table 6.1 on page 262 contains only four high-speed steels, viz. the above-mentioned grades T 1, M 2, M 7 and M 42. These four steels constitute a mere tenth of the actual number of high-speed steels on the market. From the heat-treatment angle these steels taken together represent most variants in existence, besides largely covering the fields of application of high-speed steels.

As may be seen from the above, the hardening temperatures for high-speed steels are higher than those for other tool steels. Temperatures of only some tens of degrees below incipient fusion of the steel are used. The chromium carbides go into solution around 1100°C and at the normal hardening temperature for, say, grade M 2 there are undissolved carbides left amounting to some 10%, mainly V carbides and double carbides of Mo and W. The high hardening temperatures employed for high-speed steels are conducive to rapid grain growth and hence the holding time must be carefully controlled (see the following chapter). The hardening temperature must also be accommodated to the original dimensions of the steel stock used for the tool since as the stock dimension increases the amount of carbide segregation increases which, in turn, lowers the temperature of incipient fusion. Therefore, the hardening temperature should be kept near the lower limit of the normal hardening temperature range when the dimension of the original steel stock exceeds about 100 mm.

The hardening temperature is chosen to suit the steel in question, always keeping in mind the use to which the tool is to be put. Tools for machining, e.g. turning and planing tools, or for rough milling, should be hardened from the highest temperature in order to be certain that they obtain the best hot-hardness properties since the cutting edges may reach temperatures as high as 600°C.

Tools to be used at lower temperatures or that require good impact strength, such as cold-upsetting tools, can be hardened from temperatures as low as 1050°C. By this treatment, resistance to tempering is reduced and if the hardening temperature is low enough (below 1000°C) the secondary hardening effect disappears.

As a rule, high-speed steels have good hardenability from which follows that tools made from such steels may be quenched in a salt bath or even air cooled. High-speed steel containing 10% Co has a somewhat reduced hardenability and in order to arrive at maximum hardness by means of air cooling, light sections only (less than 30 mm in diameter) can be treated in this way.

Having been quenched from a normal hardening temperature, high-speed steels contain between 20% and 40% retained austenite. As they cool from a
Figure 6.64. Tempering curves showing interdependence of time and temperature for steel M 2

Figure 6.65. Tempering curves showing interdependence of time and temperature for steel M 42
tempering temperature of about 575 °C there is practically complete transformation to martensite while at the same time the initially formed martensite is tempered. A second tempering treatment is required to give the last-formed martensite its optimum combination of useful properties. In Figures 6.6a, b and c are shown the microstructures in specimens of grade M 42 after hardening, after a single and after a double tempering treatment, respectively.

Retained austenite can also be transformed by subzero treatment. A prior subzero treatment will not affect the amount of retained austenite after tempering over about 575 °C since this constituent is decomposed at the conventional tempering temperature used. At tempering temperatures above 550 °C the length of the holding time produces a pronounced effect on the hardness, as may be seen from the curves in Figures 6.64 and 6.65 which show the interdependence of time and temperature. By increasing the tempering time from 1 to 4 h at 600 °C the hardness of grade M 2 falls from 65 to 63 HRC.

Cutting tools for which the highest hardness is required, are tempered at 550 °C. However, the hardness and tempering temperature must be adjusted to the toughness requirements. The impact strength is highest when the steel is tempered in the range 250–450 °C and lowest at the temperature that gives maximum hardness. As the steel continues to be tempered at increasing temperatures the toughness starts to increase again. Tools that in service are subjected to high pressures give best results if they have been tempered at about 600 °C; the austenite being completely transformed at this temperature.

The torsion-impact test with unnotched specimens yields results that in many cases are in agreement with those from impact strength tests. According to Figure 6.66 the torsion-impact test indicates a minimum value at 650 °C (1200 °F) whereas according to Figure 6.67 the unnotched specimen indicates a minimum value at 510 °C (950 °F). The diagrams in the latter figure also show that grade M 2 has higher impact strength and hardness than grade T 1.

![Image of graphs and diagrams showing the relationship between tempering temperature and hardness, and impact strength.](image)

The hardness of a high-speed steel, quenched and tempered according to current practice, stays at a reasonably high level up to about 500 °C (see Figure 6.68). This high hardness property is of very great importance to cutting tools, the cutting-edge temperature of which may rise to about 600 °C. As was mentioned earlier, hardenability increases as the tempering temperature increases, but the composition of the steel is the determining factor. Co and V are the alloying elements that have the most marked influence, which may be inferred from Figure 6.68.

High-speed steels find their most usual field of application as tools for machining. The most highly alloyed grades, e.g. M 42, are best able to cope with the high cutting speeds and high cutting-edge temperatures which machining tools, such as lathe and planing tools, are subjected to. Unfor-
HEAT TREATMENT—SPECIAL 319

Fortunately the superior wear resistance of this steel cannot always be utilized in milling cutters, for which toughness is imperative. Although the hardness of a turning tool may be about 68 HRC, a figure of 65 HRC should be the maximum hardness for milling cutters.

![Graph](image)

Figure 6.65. Hot-hardness of high-speed steels. Load applied for 7 s

The large amount of carbides in high-speed steels ensures that they have good wear resistance at room temperature also. Hence high-speed steels find use to some extent as punching and blanking tools for which is required a greater degree of wear resistance than that available in grades D 6 or D 2. However, the higher price and the more complex heat-treating process are factors that also should be considered when high-speed steels are being considered for such uses.

Owing to their high strength at elevated temperatures, high-speed steels are also used for hot-work tools, usually for such as perform a shearing operation, for example hot swards and punches in hot-punching machines that produce nuts. The hardness of such tools should be 55-60 HRC. The wear resistance of high-speed steels may be enhanced by carburizing or by nitriding. These processes are discussed further on in this chapter.

6.2 QUenchING AND TEMPERING OF CONSTRUCTIONAL STEELS

6.2.1 Definitions

A constructional steel, quenched from the austenitizing temperature and subsequently tempered in the range 500-700 °C, obtains useful mechanical properties, e.g., a high ratio of yield strength to ultimate tensile strength, high elongation, reduction of area and impact strength. The useful hardness range is 180-400 HB, which corresponds to an ultimate tensile strength of 60-130 kp/mm² (590-1280 N/mm²). This heat treatment—in German called Vergütung—may be described as tough-hardening. Some decades ago a hardness of 330 HB was considered the upper hardness limit that cutting tools could cope with. This limit has been successively raised and at the same time the division between ‘tough-hardened’ steels and what might be termed hard-hardened steels has been erased. ‘Hard-hardened’ constructional steels are such as have been quenched and tempered at approximately 200 °C.

The boundary between the classic ‘tough-hardening’ steels and other steels has also been bridged. The designation ‘High-strength Steels’, which covers several types of steel including tool steel, is currently becoming popular and in the cases under discussion it is applied to steels that have higher tensile strengths than the conventional ‘tough-hardening’ steels.

If it is possible to start out from prehardened steel it is usually more economic in the long run to purchase this than to buy annealed steel and carry out the necessary heat treatment after the machining operations. When the steel is bought as forgings or if mechanical properties are required greater than those available in material from stock the quenching and tempering may sometimes be carried out by the purchaser.

In the chapter on hardening, the interrelationships of composition, dimension and hardness of the steel have been discussed in detail. This information is now being put to good use. Formerly it was customary to use rather highly alloyed grades of steel even in light dimensions and this may sometimes be justified when it is expedient to stock as few different grades as possible. However, the trend of the times is to use the least expensive steel for each function. This calls for detailed knowledge of the hardenability of each steel concerned. For the rest of this chapter the steels are clased in groups, based partly on their chemical composition and partly on their fields of application.

6.2.2 Plain carbon steels

According to ISO/R 683/1-1968 there are eight heat-treatable plain carbon steels. The lowest carbon content is 0.25%, and the highest, 0.60%—see Table 6.3.

Table 6.4 contains the mechanical properties attainable on quenching and tempering the steels. The steels designated C 35 and C 50 have the most general application. It is in light dimensions that steel C 50 in particular is used for parts heat treated to high hardness or high toughness, e.g., axles and bolts. Oil quenching may be used but as shown in Figure 6.69 the steel exhibits a high degree of section sensitivity. A tempering treatment will even out the extreme differences of strength, as shown in Figure 6.70.

Water quenching must be used for heavy sections. This treatment is used for parts such as crankshafts for motor-car engines. The limits of the composition range for such steels must be set very close and the heat treatment
Table 6.4 ISO/R 683/1-1968. MECHANICAL PROPERTIES FOR QUENCHED AND TEMPERED CONDITION

<table>
<thead>
<tr>
<th>Type of steel</th>
<th>$R_p$ min (kgf/mm²; ton/in²)</th>
<th>$R_u$ min (kgf/mm²; ton/in²)</th>
<th>$A$ min (%)</th>
<th>KCU min (kgf.m/cm²)</th>
<th>$R_p$ min (kgf/mm²; ton/in²)</th>
<th>$R_u$ min (kgf/mm²; ton/in²)</th>
<th>$A$ min (%)</th>
<th>KCU min (kgf.m/cm²)</th>
<th>$R_p$ min (kgf/mm²; ton/in²)</th>
<th>$R_u$ min (kgf/mm²; ton/in²)</th>
<th>$A$ min (%)</th>
<th>KCU min (kgf.m/cm²)</th>
</tr>
</thead>
<tbody>
<tr>
<td>C25 and C25e</td>
<td>(23-5) 55-70</td>
<td>(34.5-44.4)</td>
<td>19</td>
<td>31 (19.7) 50-65</td>
<td>21 (31.7-41.3)</td>
<td>7</td>
<td>1</td>
<td>30 (19) 50-65</td>
<td>21 (31.7-41.3)</td>
<td>7</td>
<td>1</td>
<td>30 (19) 50-65</td>
</tr>
<tr>
<td>C30 and C30e</td>
<td>(25-4) 59-74</td>
<td>(37-45-47)</td>
<td>18</td>
<td>34 (21.6) 55-70</td>
<td>20 (34.5-47.4)</td>
<td>6</td>
<td>1 (19) 30 (19) 50-65**</td>
<td>21 (31.7-41.3)</td>
<td>6**</td>
<td>5</td>
<td>5 (34.5-44.4) 50-70</td>
<td></td>
</tr>
<tr>
<td>C35 and C35e</td>
<td>(27-3) 63-78</td>
<td>(40-49-5)</td>
<td>17</td>
<td>37 (23.5) 59-74</td>
<td>19 (37-5-47)</td>
<td>5</td>
<td>1 (29.9) 35 (37-5-47) 50-70</td>
<td>20</td>
<td>5</td>
<td>5 (34-44.4) 50-70</td>
<td></td>
<td></td>
</tr>
<tr>
<td>C40 and C40e</td>
<td>(29-2) 67-82</td>
<td>(42.6-52-1)</td>
<td>16</td>
<td>40 (25.4) 63-78</td>
<td>18 (42-6-52-1)</td>
<td>4</td>
<td>1 (22.2) 35 (37.5-47) 50-74</td>
<td>19</td>
<td>4</td>
<td>4 (37.5-47) 50-74</td>
<td></td>
<td></td>
</tr>
<tr>
<td>C45 and C45e</td>
<td>(31-1) 71-86</td>
<td>(45-54-6)</td>
<td>14</td>
<td>42 (26-7) 67-82</td>
<td>16 (42-6-52-1)</td>
<td>3</td>
<td>1 (24.1) 38 (40-49-5) 63-78</td>
<td>17</td>
<td>3</td>
<td>3 (40-49-5) 63-78</td>
<td></td>
<td></td>
</tr>
<tr>
<td>C50 and C50e</td>
<td>(33) 75-90</td>
<td>(47-5-57-1)</td>
<td>13</td>
<td>45 (28-6) 71-86</td>
<td>15 (45-1-54-6)</td>
<td>5</td>
<td>1 (26) 41 (42-6-52-1) 67-82</td>
<td>16</td>
<td>13</td>
<td>1 (26) 42-6-52-1</td>
<td></td>
<td></td>
</tr>
<tr>
<td>C55 and C55e</td>
<td>(34-9) 80-95</td>
<td>(50-8-60-3)</td>
<td>12</td>
<td>47 (29-8) 75-90</td>
<td>14 (47-6-57-1)</td>
<td>14</td>
<td>1 (27.3) 43 (45-1-54-6) 71-86</td>
<td>15</td>
<td>15</td>
<td>1 (27.3) 45-1-54-6</td>
<td></td>
<td></td>
</tr>
<tr>
<td>C60 and C60e</td>
<td>(36-6) 85-100</td>
<td>(54-0-63-5)</td>
<td>11</td>
<td>50 (31-7) 80-95</td>
<td>13 (50-9-60-3)</td>
<td>13</td>
<td>1 (29.2) 46 (47-6-57-1) 75-90</td>
<td>14</td>
<td>14</td>
<td>1 (29.2) 47-6-57-1</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
carried out with the utmost care. From many points of view water quenching is a convenient method of cooling. Water is a cheap quenching medium and the quenched parts need not be put through any extra cleaning operation as is the case when oil quenching is resorted to. Besides, water quenching produces maximum hardness and maximum depth of hardening. However, owing to the large temperature differences set up in the steel during the water quenching and which may cause cracking and distortion, this treatment is limited to parts of simple design and preferably to such as possess rotational symmetry.

6.2.3 Alloy steels

With regard to the standardization of alloy constructional steels it has been possible to reach agreement, even internationally, on a reduction in the number of steel types. First among the low-alloy steels there are the Cr-Mo steels, as specified by ISO/R 683/II–1968. Their chemical composition is shown in Table 6.5. Some equivalent national standards are given in Table 6.6. Table 6.7 includes the mechanical properties attainable on quenching and tempering the steels. The steels are generally oil quenched but sometimes they may be quenched in water. Steel No. 1 comes through this latter heat treatment best. Figure 6.7I shows the surface hardness obtained in various dimensions of steel No. 1 quenched in oil and water, respectively. When quenched in water and then tempered at 200 °C the steel shows a hardness difference of about 50 HB units between surface and core in bar stock 100 mm in diameter (see Figure 6.7J). When the steel is tempered at 500 °C, which is the temperature used for 'tough-hardening', the hardness difference is reduced to some 50 HB units only. A point in favour of steel No. 1 is its good weldability.

Table 6.5 ISO/R 683/II–1968: TYPES OF STEEL AND CHEMICAL COMPOSITION GUARANTEED

<table>
<thead>
<tr>
<th>Type of steel</th>
<th>C %</th>
<th>Si %</th>
<th>Mn %</th>
<th>P % max</th>
<th>S % max</th>
<th>Cr %</th>
<th>Mo %</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>0.22–0.29</td>
<td>0.50–0.80</td>
<td>0.035</td>
<td>0.035</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>0.30–0.37</td>
<td>0.50–0.80</td>
<td>0.035</td>
<td>0.035</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>3</td>
<td>0.38–0.45</td>
<td>0.50–1.00</td>
<td>0.035</td>
<td>0.035</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Table 6.6 DESIGNATIONS FOR CR–MO STEELS

<table>
<thead>
<tr>
<th>ISO</th>
<th>AISI</th>
<th>BS</th>
<th>En</th>
<th>DIN</th>
<th>SIS</th>
<th>Before</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>4130</td>
<td></td>
<td></td>
<td></td>
<td>25CrMo4</td>
<td>2225</td>
</tr>
<tr>
<td>2</td>
<td>4135</td>
<td>708 A 37</td>
<td>19B</td>
<td>34CrMo4</td>
<td>2234</td>
<td>RO 752</td>
</tr>
<tr>
<td>3</td>
<td>4142</td>
<td>708 A 42</td>
<td>19C</td>
<td>42CrMo4</td>
<td>2244</td>
<td>RO 952</td>
</tr>
</tbody>
</table>

Figure 6.7J shows a tempering diagram for steel No. 3. Note the definite reduction in impact strength after tempering the specimen around 300 °C. To a large extent steels Nos. 2 and 3 are both used for parts that are either flame or induction hardened.

If high mechanical properties in heavy sections are stipulated, higher-alloy steels are used. Examples of such are given in Table 6.8, viz. a Cr–Ni–Mo steel and a Cr–Mo steel. The Cr–Mo steel (SIS 2240) is included in ISO/R
<table>
<thead>
<tr>
<th>Type of steel</th>
<th>$R_y$ min (kgf/mm²)</th>
<th>$R_m$ max (kgf/mm²)</th>
<th>$A$ min (%)</th>
<th>$KCU$ min (kgf/cm²)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>70 (44-4)</td>
<td>90-110 (57-1-69-8)</td>
<td>12</td>
<td>60 (38)</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>80-95 (59-8-60-3)</td>
</tr>
<tr>
<td>2</td>
<td>80 (50-8)</td>
<td>100-120 (63-5-76-2)</td>
<td>11</td>
<td>68 (43-7)</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>90-110 (57-1-69-8)</td>
</tr>
<tr>
<td>3</td>
<td>90 (57-1)</td>
<td>110-130 (69-8-82-5)</td>
<td>10</td>
<td>78 (49-5)</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>100-120 (63-5-76-2)</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Type of steel</th>
<th>$R_y$ min (kgf/mm²)</th>
<th>$R_m$ max (kgf/mm²)</th>
<th>$A$ min (%)</th>
<th>$KCU$ min (kgf/cm²)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>52 (31-6)</td>
<td>75-90 (47-6-57-1)</td>
<td>15</td>
<td>47 (29-8)</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>70-85 (44-4-54)</td>
</tr>
<tr>
<td>2</td>
<td>57 (36-2)</td>
<td>80-95 (59-8-60-3)</td>
<td>13</td>
<td>52 (33)</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>75-90 (47-6-57-1)</td>
</tr>
</tbody>
</table>

* $R_y$ = yield stress (0.2%), proof stress
$R_m$ = tensile strength
$A$ = percentage elongation after fracture ($A_e = 5-6$)
$KCU$ = impact strength with U-notch
Table 6.8 HIGH-ALLOY HEAT-TREATABLE HIGH-STRENGTH STEELS

<table>
<thead>
<tr>
<th>C</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>BS</th>
<th>EN</th>
<th>DIN</th>
<th>SIS</th>
<th>BEFORS</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.25</td>
<td>1.4</td>
<td>1.4</td>
<td>0.2</td>
<td>817 M 40</td>
<td>24</td>
<td>34 Cr-Ni-Mo 6</td>
<td>2541</td>
<td>CHRO 861</td>
</tr>
<tr>
<td>0.10</td>
<td>3.0</td>
<td>0.6</td>
<td>0.4</td>
<td>29B</td>
<td>32 Cr-Mo 12</td>
<td>2240</td>
<td>RO 7155</td>
<td></td>
</tr>
</tbody>
</table>

683/V1–1970 which gives the following mechanical properties for the steel (Table 6.9). The mechanical properties of the Cr-Ni-Mo steel BS 817M40 are practically the same as those of the Cr-Mo steel recommended by ISO. Figure 6.74 shows the tempering diagram for the Cr-Ni-Mo steel. Although this steel is essentially a prehardening steel it has begun to be increasingly used in a condition of high hardness, i.e. quenched and tempered at about 200 °C. With this treatment a hardness of 500–550 HB is obtainable in oil-quenched sections up to about 75 mm diameter.

Some examples of quenched and tempered parts are shown in Figure 6.75. All of them are subjected to high stresses, both static and dynamic and

![Figure 6.73](image)

Figure 6.73. Tempering diagram for SIS 2244 (BEFORS 0952). Diagram applicable to hot stock 30 mm (1/8 in) in diameter, oil quenched from 850 °C.

![Table 6.9](image)

Table 6.9 ISO-RENIS 1070: MECHANICAL PROPERTIES FOR THE QUENCHED AND TEMPERED CONDITION

<table>
<thead>
<tr>
<th>Type</th>
<th>Rmin (0.2% proof stress)</th>
<th>Rm min (yield stress)</th>
<th>Rm max (ultimate tensile stress)</th>
<th>A</th>
<th>% El</th>
<th>Charpy U 0°C</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>150 (600 kg/mm²)</td>
<td>265 (1070 kg/mm²)</td>
<td>360 (1520 kg/mm²)</td>
<td>5</td>
<td>80</td>
<td>70</td>
</tr>
<tr>
<td>2</td>
<td>200 (800 kg/mm²)</td>
<td>320 (1300 kg/mm²)</td>
<td>420 (1760 kg/mm²)</td>
<td>10</td>
<td>60</td>
<td>60</td>
</tr>
</tbody>
</table>

KCU = impact strength with crack.
3.30 Heat Treatment—Special

Therefore the fillets between the changes of section have been burnished. This finish treatment increases the fatigue strength by 25–50%.

6.2.4 Stainless steels

Included among the so-called stainless steels there is a group of martensitic or heat-treatable steels containing about 13% Cr. Table 6.10 gives the carbon content and designations of some of the most frequently occurring steels of this group. Table 6.11 gives the hardness obtainable on quenching from 1000°C in oil or air, followed by tempering at different temperatures. The values hold for 50 mm diameter bar stock. A complete tempering diagram for S1S 2303 (A1S1 420) is shown in Figure 6.76. This steel is generally stocked quenched and tempered to a tensile strength of 70–85 kp/mm² (690–830 N/mm²) but it may also be supplied heat treated to a tensile strength of 90–110 kp/mm² (880–1080 N/mm²) in dimensions up to 200 mm. One of the uses of this steel is for structural members in the chemical and cellulose industries; another for components in steam and gas turbines. It is also used for propeller shafts in ships sailing in fresh water.

Grade A1S1 420 and steels containing higher carbon contents are used when high hardness is required. A popular field of application for such steels is stainless steel cutlery.

The current 13% Cr steels are not readily weldable. During the last ten years or so, steels have been developed that have the same mechanical properties and corrosion resistance as these steels and that, in addition, can be welded cold and need no subsequent heat treatment. The chemical composition and mechanical properties of two such steels are given in Table 6.12. The figures given in the table are obtained in specimens, air cooled from 1000°C and tempered at 600°C.

In the hardened, untempered condition the steel is martensitic. On tempering there is a partial retransformation to austenite, see Figure 6.77.
Table 6.12 NOMINAL COMPOSITION AND MECHANICAL PROPERTIES OF WELDABLE MARTENSITIC STAINLESS STEELS

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>R^s, min</th>
<th>R^m, min</th>
<th>A^s, min</th>
<th>Hardness</th>
<th>KCU, min</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>kpm/\text{cm}^2</td>
<td>kpm/\text{cm}^2</td>
<td>%</td>
<td></td>
<td></td>
</tr>
<tr>
<td>2 R M 2</td>
<td>0.05</td>
<td>12.5</td>
<td>5.5</td>
<td>—</td>
<td>60</td>
<td>590</td>
<td>80</td>
<td>780</td>
<td>15</td>
</tr>
<tr>
<td>7 R M O</td>
<td>0.06</td>
<td>12.5</td>
<td>6.0</td>
<td>1.5</td>
<td>62</td>
<td>610</td>
<td>85</td>
<td>830</td>
<td>15</td>
</tr>
</tbody>
</table>

Figure 6.76. Tempering diagram for AISI 430 (Befors 2 R M 2). Curves applicable to bar stock 50 mm diameter, oil quenched or air cooled from 1000°C

Figure 6.77. Tempering diagram for steel type 13-6 (Befors 2 R M 2)
and this is accompanied by an increase in impact strength. When the steels are welded in this tempered condition they still retain about 25% austenite after completion of the welding operation and it is for this reason they can be welded cold without danger of cracking. This valuable property is of great importance when water turbines are being repaired (see Figure 6.78).

Figure 6.78. Francis and Diagonal turbine runner of cast steel Bofors 2 RM 2

Besides water turbines, feed screws in the cellulose industry and compressor components are made from this steel. For the last application the high impact strength of the steel at low temperatures is most useful. This type of steel has excellent hardenability and hence it is also suitable in heavily dimensioned machine and structural members, such as pump axles in nuclear plant. The two steels, 2 RM 2 and 2 RM 0 are patented by the Bofors Company.

6.2.5 Spring steels

The chemical composition of these steels varies from that of plain carbon steels to high-alloy steels, according to dimensions and applications. Some of the most popular steels of this group are given in Table 6.13.

<table>
<thead>
<tr>
<th>Steel designation</th>
<th>BS</th>
<th>SIS</th>
<th>Bofors</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>V</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>En 42</td>
<td>1770</td>
<td>B 14</td>
<td>0.70</td>
<td>0.3</td>
<td>0.4</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>En 43</td>
<td>2060</td>
<td>S 141</td>
<td>0.55</td>
<td>1.7</td>
<td>0.7</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>En 47</td>
<td>2250</td>
<td>R P 112</td>
<td>0.50</td>
<td>0.3</td>
<td>0.9</td>
<td>1.3</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>0.15</td>
<td>—</td>
</tr>
<tr>
<td>BH 13</td>
<td>2252</td>
<td>ROP 19</td>
<td>0.40</td>
<td>1.0</td>
<td>0.06</td>
<td>5.3</td>
<td>—</td>
<td>1.4</td>
<td>1.0</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>'A 286'</td>
<td>2570</td>
<td>A 286</td>
<td>0.05</td>
<td>—</td>
<td>—</td>
<td>15.0</td>
<td>26.0</td>
<td>1.3</td>
<td>0.3</td>
<td>2.0</td>
<td>—</td>
</tr>
</tbody>
</table>

A large field of application for carbon steel is helical springs. The wire is first hardened by patenting and then drawn to the required strength. Thin-section carbon-steel springs can be quenched and tempered in the conventional way also. The hardness of a spring must match its dimensions.

Figure 6.79. Tempering diagram for S15 2090 (En 45) and S15 2230 (En 47)
In principle, the smaller the dimensions of the spring the higher is the hardness. Watch springs of only a few tenths of a millimetre in thickness are tempered at 160–300°C after being quenched. Leaf springs of thicknesses 1–3 mm are tempered at 300–400°C. The 300°C tempering range is not considered detrimental to springs not subjected to impact blows. On the contrary, tempering in this temperature range sometimes produces beneficial results since the yield point often shows a maximum when the steel is tempered at about 300°C.

The diagrams in Figure 6.79 should be studied when suitable tempering temperatures for SI S 2090 and SI S 2250 are being decided upon. Springs working at temperatures up to 250°C should be made from grade H 13. For springs at elevated temperatures (up to 550°C) grade A 286 should be used.

### 6.2.6 High-strength steels

The designation above is applied to steels that can be heat treated to a yield strength of about 150 kp/mm² (1470 N/mm²), even in very heavy sections. Tables 6.14 and 6.15 give the two types of steel which are discussed.

**Table 6.14 NOMINAL COMPOSITION OF HIGH-STRENGTH STEELS**

<table>
<thead>
<tr>
<th>Type of steel</th>
<th>Nominal composition %</th>
<th>C</th>
<th>Si</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>V</th>
<th>Co</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>H 13</td>
<td></td>
<td>0-40</td>
<td>1</td>
<td>1</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
</tr>
<tr>
<td>Maraging steel,</td>
<td></td>
<td>0-02</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>7.5</td>
<td>0.4</td>
</tr>
<tr>
<td>grade 250</td>
<td>BOP 19</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>MAR 2</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

The heat treatment and properties of grade H 13 have been discussed in detail in Section 6.1.5. In its role as a structural steel it covers a wide spectrum of uses both with regard to fields of application and component size. For example, *Figure 6.80* shows a pawl, 55 mm in length, which was heat treated to 50–52 HRC. The tip of the pawl is subjected to both wear and impact. At the other extreme, *Figure 6.81* shows a component made from the same

**Table 6.15 MECHANICAL PROPERTIES FOR HIGH-STRENGTH STEELS AFTER HARDENING AND TEMPERING**

<table>
<thead>
<tr>
<th>Type of steel</th>
<th>BOP 19</th>
<th>(R_{p0,2})</th>
<th>(R_{p0,5})</th>
<th>(R_{e0,2})</th>
<th>(A_{5})</th>
<th>(Z_{5})</th>
<th>Hardness</th>
<th>Temper-</th>
<th>mm²</th>
<th>mm²</th>
<th>mm²</th>
<th>%</th>
<th>%</th>
<th>%</th>
<th>HB</th>
<th>mm²</th>
</tr>
</thead>
<tbody>
<tr>
<td>H 13</td>
<td>155</td>
<td>1520</td>
<td>180</td>
<td>1730</td>
<td>10</td>
<td>45</td>
<td>520</td>
<td>570</td>
<td></td>
<td></td>
<td></td>
<td></td>
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<tr>
<td>Maraging steel,</td>
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<tr>
<td>grade 250</td>
<td>170</td>
<td>1670</td>
<td>180</td>
<td>1770</td>
<td>10</td>
<td>50</td>
<td>470</td>
<td>480</td>
<td></td>
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</tbody>
</table>

*Figure 6.80. Pawl made from grade H 13 (BOP R 19). Hardness 50–52 HRC, weight 47 g*
steel. It forms part of the largest press (80000 ton) in the world. This member has the same hardness as the pawl.

There is also a steel designated grade H 11 modified, which is a variant of grade H 13 with a somewhat modified chemical composition. This grade is used largely in the aircraft industry which in addition stipulates high cleanliness requirements for this steel. Figure 6.82 shows a structural member for aeroplane landing gear.

The second steel in the table above belongs to what are known as maraging steels. The heat treatment consists of heating the steel to 820°C from which temperature it forms martensite on cooling. However, this martensite is soft and thus the heat treatment is often referred to as annealing. On subsequent ageing the steel at 480°C for 3 h, very high mechanical properties are obtained. There are four types of maraging steel, viz. grade 200, 250, 300 and 350; the number indicating the ultimate tensile strength, in ksi (kilo-pound/square inch). The tensile strength depends on the Ti content which varies between 0.1% and 1.5%.

![Diagram](image)

The maraging steels are less susceptible to hydrogen embrittlement than are steels of type H 13. They have been developed in the USA where they are used for structural members in aircraft and spacecraft designs, principally owing to their high impact properties at low temperatures.

Fatigue tests with unnotched specimens give a somewhat higher endurance limit for grade H 13 than for the 250-grade maraging steel. Tests with notched specimens having a stress concentration factor of 2.8 show that the endurance
limit is dependent on the required service life. These observations are set out in diagram form in Figure 6.83 which holds for vacuum-remelted steel.

6.2.7 Hadfield steel

This rather special steel (Figure 6.84) has been named after its inventor, Sir Robert Hadfield. In its quench-annealed condition the steel is completely austenitic. Its hardness is about 220 HB and its composition 1.25% C, 13% Mn and 0.5% Cr. Owing to its high Mn-content the steel is often merely referred to as a manganese steel. On being subjected to cold work the austenite is transformed to martensite and the hardness rises to about 600 HB. According to Schumann the transformation mechanism is as follows:

Austenite → dislocations → stacking faults → ε-martensite → α-martensite.

The Hadfield steel has been dealt with earlier in this book, see pp. 190 and 250.

6.3 CASE HARDENING

6.3.1 Definitions

Low-carbon steels, which as a rule are readily machinable, can have their surface layers carburized and subsequently hardened. This treatment, called case hardening, gives the steel a hard and wear-resisting surface or case. By virtue of the core remaining comparatively soft and tough the component as a whole shows high impact strength. Owing to the development of compressive stresses in the surface layers during the case-hardening treatment the fatigue strength of the steel is also increased.

Carburization is effected either in solid media (carburizing compounds), in salt baths or in gases at temperatures normally between 825°C and 925°C. The transport of carbon from the carburizing medium always takes place via a gaseous phase, usually carbon monoxide (CO).

The hardening treatment may be carried out as follows:

(a) Direct quenching, i.e. quenching straight from the carburizing medium.

(b) Single quenching, i.e. heating and quenching the carburized parts after first allowing them to cool to room temperature from the carburizing treatment.

(c) Double quenching which usually consists of a direct quench and then a re-quench from a lower temperature. Double quenching may also imply that after carburization the parts are allowed to cool to room temperature and are then subjected to their first quenching or normalizing treatment from 850–900°C. The second quenching is carried out in the same way as in the single quenching, usually from a temperature range between 780°C and 820°C, which is the proper quenching range for the case.

Double quenching, when carried out according to the latter method is applied to steels that have not been fine-grain treated. The other methods, however, are applied mainly to fine-grained steels. The various methods of quenching and tempering will be described in more detail in Section 6.3.3.

By the term depth of carburization is meant either the distance below the surface to a definite carbon concentration, or the total depth of carbon penetration. In addition to time and temperature, the depth of carburization depends on the carbon potential of the carburizing medium and on the composition of the steel. The higher the carbon potential the higher the carbon concentration at the surface of the steel, when equilibrium has been established, and the deeper the carburizing depth.

The carbon potential may be defined as that carbon content which a specimen of a carbon-steel foil acquires when equilibrium conditions have been established between the carbon potential of the carburizing medium and the carbon content of the foil. The dependence of the depth of carburization on the carbon potential and carburizing time is shown in Figure 6.85.

A mathematical model for calculating the carburized concentration profile for carbon has been advanced by Colin, Gunnarson and Thulin. The highest carbon content that the austenite of plain carbon steel can hold may be deduced from the iron–carbon equilibrium diagram. If the carbon concentration is higher, carbides will have formed. This formation is facilitated if the steel contains carbide-forming elements or if it has a composition that favours carbide formation (see Section 3.1.4).

The depth of case hardening, D C, is defined, according to SIS 11 7008, as the distance from the surface to a plane at which the hardness is 550 HV. This definition will also be adopted by ISO (ISO 2639–1973). The depth of case hardening is determined by HV measurements, using a load of 1 kp, on a section through a case-hardened surface and at right angles to it. Figure
Figure 6.86. Schematic diagram showing hardness profile of case in a case-hardened steel. Depth of case hardening, DC, as shown in figure, is 1.15 mm.

Figure 6.86 shows the hardness profile across a section of a case-hardened steel. With the aid of a magnifying lens the depth of case hardening, viewed at a magnification of about 10 times, may be approximately assessed on samples or on actual parts fractured after case hardening. The appearance of such fractured test specimens is shown in Figure 6.87. There exists a fair measure of agreement between the determined and the assessed depths of case hardening, and the latter is widely used for control and check of the carburizing process. Should an exact measure of the depth of case hardening be required, hardness measurements must be resorted to.

Figure 6.87. Test specimens of free-cutting steel SIS 1922, quenched and fractured in order to assess depth of case hardening. Water quenched from (a) 900 and (b) 850°C.

Figure 6.88 has been constructed to serve as an aid to heat-treatment operators who handle steel parts of various grades and dimensions. The curves are based on a large number of case-hardening tests and are sufficiently accurate for practical carburizing work when using a solid compound, a salt bath or a gas. The curves apply to alloy case-hardening steels in dimensions up to about 50 mm in thickness, oil quenched; also to unalloyed case-hardening steels in sections up to about 100 mm, water quenched. The diagram is in good agreement with documented data when the depth of case hardening has been determined in the same way as that described in SIS 11 70 08.

Figure 6.88. Case depth versus duration and temperature of carburization.

The validity of the simple diffusion equation \( x = k \sqrt{t} \) has been checked against the diagrams and the following values have been obtained for the constant \( k \) when \( x = DC \) in mm and \( t = \) time in hours.

<table>
<thead>
<tr>
<th>Temperature °C</th>
<th>875</th>
<th>900</th>
<th>925</th>
</tr>
</thead>
<tbody>
<tr>
<td>Constant ( k )</td>
<td>0.34</td>
<td>0.41</td>
<td>0.52</td>
</tr>
</tbody>
</table>

**Example:** What is the depth of case hardening obtained when carburizing for 12 h at 900°C and hardening?

\[ DC = 0.41 \sqrt{12} = 1.42 \text{ mm} \]

### 6.3.2 Grades of steel

ISO recommendations for case-hardening steels are contained in ISO/R 685/X1-1970. Tables 6.16 and 6.17 list a selection of steels and their properties of those widely used both in Sweden and in the U.K. Figure 6.89 shows by means of a series of diagrams the guaranteed mechanical properties of the core of the alloy steels in dimensions up to 100 mm. The diagrams are...
<table>
<thead>
<tr>
<th>AISI</th>
<th>BS</th>
<th>En</th>
<th>S1S</th>
<th>Bofors</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>0.020</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
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<tr>
<td>1015</td>
<td>—</td>
<td>3</td>
<td>1730²</td>
<td>B 4 V</td>
<td>0.12</td>
<td>0.10</td>
<td>0.60</td>
<td>0.035</td>
<td>0.035</td>
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<tr>
<td>8620</td>
<td>803M20</td>
<td>362</td>
<td>2506¹</td>
<td>C R O 42</td>
<td>0.17</td>
<td>0.15</td>
<td>0.60</td>
<td>0.035</td>
<td>0.030</td>
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<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.18</td>
<td>0.040</td>
<td>0.090</td>
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<td></td>
<td></td>
<td>0.18</td>
<td>0.40</td>
<td>0.035</td>
<td>0.035</td>
<td>0.035</td>
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<td></td>
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<tr>
<td>A 3115</td>
<td>63M17²</td>
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<td>2511</td>
<td>DR 34</td>
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<td>0.15</td>
<td>0.70</td>
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<td>0.030</td>
<td>0.60</td>
<td>0.80</td>
<td>—</td>
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<td></td>
<td></td>
<td>0.18</td>
<td>0.40</td>
<td>0.035</td>
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<tr>
<td>A 3120</td>
<td>63M17²</td>
<td>352</td>
<td>2512</td>
<td>DR 44</td>
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<td>0.15</td>
<td>0.70</td>
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<td>0.030</td>
<td>0.40</td>
<td>0.80</td>
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<td></td>
<td></td>
<td>0.23</td>
<td>0.40</td>
<td>0.110</td>
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<td>0.035</td>
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<td>4720</td>
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<td>2523</td>
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<td>0.15</td>
<td>0.70</td>
<td>0.030</td>
<td>0.030</td>
<td>0.80</td>
<td>1.00</td>
<td>0.08</td>
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<tr>
<td></td>
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<td></td>
<td></td>
<td></td>
<td>0.23</td>
<td>0.40</td>
<td>0.110</td>
<td>0.035</td>
<td>0.035</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

1. C 0.14-0.20
2. S1S 170 corresponds to grade 3A to ESSR 405/32: 1970.
3. S1S 204 corresponds to grade 3B to ESSR 405/32: 1970.
6.3.3 Methods of carburizing

Regardless of what method is used, carburization always takes place via a gaseous phase. However, each method has its own intrinsic characteristics which produce differing case-hardening results.

Carburizing by means of solid substances (granular compounds, pack carburizing)

Carburization with charcoal alone (without activating agents or energizers) can take place thanks to the presence of the atmospheric oxygen enclosed in the carburizing box. The oxygen reacts with the charcoal and during the heating-up period a CO₂-rich mixture is produced which then continues to react with the charcoal as follows:

\[ \text{CO}_2 + \text{C} \rightarrow 2\text{CO} \]  
(6.1)

As the temperature rises the equilibrium is displaced to the right, i.e. the gas mixture becomes progressively richer in CO.

At the steel surface the CO breaks down as follows:

\[ 2\text{CO} \rightarrow \text{CO}_2 + \text{C} \]  
(6.2)

The atomic or nascent carbon thus liberated is readily dissolved by the austenite phase of the steel and diffuses into the body of the steel, even to the extent of forming carbides (cementite). The CO₂ produced reacts again with the charcoal according to (6.1) above and the cycle of reactions is repeated.

Since the amount of atmospheric oxygen in a charcoal packing can vary, and may be insufficient to produce the carburizing gas it is current practice to mix the charcoal with an energizer, usually barium carbonate, which reacts during the heating-up period as follows:

\[ \text{BaCO}_3 \rightarrow \text{BaO} + \text{CO}_2 \]  
(6.3)

\[ \text{CO}_2 + \text{C} \rightarrow 2\text{CO} \]  
(6.4)

The CO₂ originally formed then reacts with the carbon in the charcoal, producing the active CO.

Since all the reactions mentioned above can proceed in both directions, i.e. they are reversible, the correct way of representing reaction (6.4) is

\[ \text{CO}_2 + \text{C} \rightleftharpoons 2\text{CO} \]  
(6.5)

If the temperature is increased but the pressure kept constant the reaction proceeds from left to right, i.e. more CO is produced. On lowering the temperature the proportion of CO₂ increases at the expense of the CO. Figure 6.90 shows the equilibrium relationships existing between the constituents in the gaseous mixture expressed as percentages by volume, at a total pressure of one atmosphere. At 900 °C, which is a customary carburizing temperature, equilibrium is at about 96% CO and 4% CO₂.

This equilibrium curve is also included in Figure 6.91, which is part of the equilibrium diagram for the Fe–O–C system. Given that at 900 °C the mixed gases are in equilibrium according to Equation (6.5) the carbon concentration at the surface of the steel will adjust itself to some value between 0.70%C and 1.2%C which is the equilibrium carbon concentration at the phase-field boundary A₃ at this temperature. If the CO₂ content falls to 2%C, the
some Fe₃C may be precipitated; conversely, should for some reason the CO₂ content increase, owing to, say, the admittance of air to the carburizing compound, the carbon content of the steel will fall correspondingly. For example: at 900°C, 6% CO₂ results in about 0.70% C and 14% CO₂ in about 0.30% C.

When steel parts are being pack carburized in a carburizing compound continuously or regularly, boxes should be made from heat-resisting steel, type 25% Cr, 20% Ni being suitable for this purpose. A box for occasional work may be made from ordinary mild-steel sheet.

After being cleaned the parts are packed into the boxes on the bottom of which is first laid a layer of the compound, about 25 mm thick. The purpose of this layer is to keep the parts from coming into contact with the bottom itself of the box since they might otherwise become deformed during the carburizing process. When a layer of the parts has been placed in the box more compound is added to make a somewhat thinner layer, then more parts are packed and so on. When the top layer of the parts has been packed in the box there should be room at the top for a 50 mm layer of compound before the lid is placed in position.

It is important to make the lid airtight with clay or similar material as the carburizing process may be prevented by air ingress. During the heating-up period the clay dries and may crack and hence its sealing properties are not wholly reliable. A good way of preventing the ingress of air is to make a rimmed lid about 25 mm larger than the box. When the lid has been placed in position the box is turned upside-down and the space between it and the rim of the lid is sealed with refractory compound (see Figure 6.92). Round boxes made from heat-resisting steel sheet have adequate rigidity and there is generally no need to use any sealing material.

![Figure 6.92: Section through fully packed carburizing box](image)

Conventional carburizing compounds contain between 6 and 20% of energizer. This substance is usually evenly mixed with the charcoal and in principle, the carbon of the compound can be fully utilized. A batch of fresh carburizing compound will shrink a good deal in volume after its first use, a reduction by as much as a fifth of its original volume being not unusual. Hence, before starting another treatment fresh compound is added to make up the original volume. In this way a mixture is obtained which shrinks less and, more important, which shrinks by a known amount.

Carburizing compounds have relatively poor heat conductivity. Consequently when large carburizing boxes are used, workpieces lying in the middle of the box are carburized considerably less than those lying near the sides. The smaller the boxes used the more uniform is the carburizing effect likely to be. The boxes themselves must not be placed so close together that heat radiation or the free circulation of the furnace atmosphere is impeded. It is good practice to make the boxes cylindrical whereby more space between them is automatically created. The service life of cylindrical boxes is also longer and the risk of cracking smaller. From the point of view of both production economics and uniformity of carburization, optimum results are obtained if the diameter or width of the box is about 150 mm. Boxes having a diameter or width exceeding 300 mm should not be used unless the workpiece dictates this size.

The case depth is kept under control by inspecting test specimens or, if large series are concerned, by checking actual parts. The curves in Figure 6.88 can serve as a guide when estimating the time necessary to obtain a stipulated case depth. Since the curves refer to holding times at the temperatures concerned, the heating-up time must be assessed. The dimensions of the boxes, the size of the charge and the heating capacity of the furnace

![Figure 6.93: Experimental carburizing set-up in order to assess heating-up time at various distances from box walls. Box dimensions: 130 mm diameter × 320 mm](image)
are all important factors. A simple way of judging the temperature in the middle of the box is to insert a steel rod through a hole in the box and then pull it out when it is thought that the preset temperature has been reached. The colour of the red-hot bar will give an indication of the temperature inside the box. Alternatively, a tube closed at one end and protruding into the box may be welded to one side and at right angles to it. The temperature may then be determined by means of a thermocouple inserted into the tube through a hole in the box.

In order to assess heating time, tests have been carried out with boxes, 130 mm diameter x 320 mm. Specimens 10 mm diameter x 50 mm, were packed in the boxes at different distances from the wall. In one test they were placed immediately next to larger specimens as sketched in Figure 6.92. Thermocouples were attached to the specimen and were connected to a temperature recorder. The furnace measured $378 \times 500 \times 800$ mm and was rated at 24 kW.

During the heating-up period the temperature increased at the rate shown in Figure 6.94. Specimen No. 1 in box A reached the furnace temperature only slightly ahead of specimen No. 2. Specimen No. 3 required an additional 10 min to reach 900 °C. The specimens in box B required about twice as long to reach the pre-set temperature but they all reached it in about the same time. At first the temperature of specimen No. 6 increased more slowly than that of the others (see specimen No. 3). At the higher temperatures the rate of heating was faster compared with the other specimens. This was due to the radiation absorbed at the top of the box and also to the lower rate of heating of the larger specimens.

In the test just described it is estimated that the specimens in box A obtained about one hour's longer carburizing time than the specimens in box B. Assuming a total furnace time of, say, 5 h at 900 °C, this would imply 4 h and 3 h of actual carburizing time, respectively, which, in turn, would result in a carburized depth of 0.85 mm and 0.70 mm for the specimens in boxes A and B respectively (see Figure 6.98).

SALT-BATH CARBURIZING

The active carburizing agent in a salt bath is sodium cyanide (NaCN) or potassium cyanide (KCN). It is believed that the carburization proceeds via a gaseous phase according to the following reactions:

$$2 \text{NaCN} + \text{O}_2 \rightarrow 2 \text{NaCNO} \quad (6.6)$$

$$4 \text{NaCNO} \rightarrow 2 \text{NaCN} + \text{Na}_2\text{CO}_3 + 6 \text{CO} \quad (6.7)$$

$$3 \text{Fe} + 2 \text{CO} \rightarrow \text{Fe}_3\text{C} + \text{CO}_2 \quad (6.8)$$

The first reaction takes place at the interface between the salt bath and the atmosphere; the other two reactions take place at the interface between the salt bath and the steel. Some of the nitrogen liberated by reaction (6.7) is also taken up by the steel.

The conventional cyanide baths that give a case depth up to 0.8 mm have
a cyanide content (NaCN) of 40–50%. When a fresh cyanide bath is being prepared it is customary to start with pure NaCN which is melted down in the furnace. NaCN decomposes quite rapidly at the beginning of the process whereby the cyanide content assumes the value stated above.

A salt bath containing about 20% NaCN may be used to treat very thin parts requiring a case depth of only a few tenths of a millimetre. The parts are quenched direct into water after being carburized. Such a bath is prepared by melting together equal parts of sodium cyanide and anhydrous sodium carbonate.

To obtain case depths up to 1.5 mm, so-called activated baths are employed, the NaCN content of which is about 10%. These baths are also used in conjunction with baths for the martempering of steel at about 200°C, which contain sodium nitrite and potassium nitrate. When the parts are transferred from the cyanide bath to the martempering bath some cyanide salt will be carried over. If the content of the bath is more than 10% NaCN the reaction between the carbon in the cyanide salt being carried over and the oxygen in the martempering bath will become explosively violent.

The NaCN content of the cyanide bath should be checked daily and fresh salt added as required. Generally the NaCN content is satisfactorily maintained simply by replacing the volume of salt carried out with an equal amount of fresh salt. The surface of the bath should be kept covered with a layer of graphite fines which prevents radiation loss and gas evolution from the bath.

The amount of carbon and nitrogen picked up by the steel depends mainly on the cyanide content of the bath and on its temperature. Figure 6.95 shows that the carbon concentration increases whereas the nitrogen decreases as the cyanide content increases from 10 to 50%.

With rising temperature, the NaCN content remaining constant at about 50%, the concentration of carbon increases and that of nitrogen decreases as shown in Figure 6.96. Both diagrams have common coordinates at 950°C and 50% NaCN. The differences existing between the curves at this point are due to the somewhat different conditions under which the tests were carried out19. Obtained from the same source, Figure 6.97 gives the hardness profiles of a mild steel, specimens of which were carburized in a cyanide bath at different temperatures, oil quenched, heated to 780°C and water quenched. The diagram in Figure 6.88 can also be used to assess the case depth. Leaflets and brochures describing the various liquid bath carburizing salts generally give curves showing the total carburized depth but this information is usually of only minor interest.

Salt-bath carburizing is mainly used for small parts that in general require a case depth less than 0.5 mm. The smaller the case depth required, the greater is the economy achieved by using a salt bath since the rate of heating is very high compared with that in solid carburizing media. Before the parts are immersed in the salt bath they should be preheated to between 100°C and 400°C partly to remove any traces of moisture and partly to utilize more efficiently the capacity of the salt bath. For further information on carburizing salts the reader is recommended to contact the suppliers concerned.

It may happen that even when parts are being carburized in a cyanide bath the surface carbon concentration becomes too high, with the result that the quenched hardness may show somewhat low values on account of there being too much retained austenite present. The carbon content increases as the time of treatment increases and as the temperature rises. Therefore, when the parts are to be quenched direct from the cyanide bath a moderate carburizing temperature should be chosen. The result of a trial hardening treatment, using BS 655M13 (En 36 A) is shown in Figure 6.98. From these curves it can be seen how the thickness of the soft surface zone increases as the treatment time increases. The surface hardness of the specimens was also checked, using HRC, and from this it could be seen that in this instance the case depth should be about 0.4 mm if the hardness is to exceed 60 HRC.

By tempering the specimens at 180°C the surface hardness is reduced but
Figure 6.97: Curves showing depth of hardening in mild steel test bars, 14 mm diameter, carburized in a cyanide bath at various temperatures and oil quenched. Bars reheated to 780 °C and water quenched (after Waterfall).

Figure 6.98: Steel BS 555 M13, En 96 A (Before R R 11). Hardness as quenched after cyaniding at 870 °C for various times. (No tempering)

Figure 6.99: Steel BS 555 M13, En 96 A (Before R R 11). Hardness after cyanide hardening at 870 °C for various times. Tempering for 1 h at 180 °C.
the case-hardened depth remains largely the same (see Figures 6.98 and 6.99).

If a case depth exceeding 0.5 mm is required it is more economical to use higher carburizing temperatures, e.g. 900–925 °C. When alloy steels are quenched direct the surface hardness is rather low and consequently they are usually rehardened by a requenching treatment from some lower temperature. If such double quenching is resorted to, the first quench takes place in oil or in a salt bath. If the parts are allowed to cool in air oxidation of the surface may occur.

Bungardt, Brandis and Kroy\textsuperscript{11} have studied the carburizing process in a liquid salt bath at temperatures between 900 °C and 1000 °C in great detail and found that a normal NaCN salt bath will give a soft surface due to high content of retained austenite. There are, however, now available on the market salt baths that will build up a definite surface carbon concentration, say 0.80%, and a tolerable content of retained austenite.

N.B. Sodium cyanide is extremely poisonous if it gets into the human system. There are special instructions to be followed when work with cyanide salts is in progress and workers should make themselves familiar with these and the procedure to be adopted in case of accidents. Municipalities using bacteriological water purification, i.e. purifying sewage by means of bacteriological action, will require the installation of a detoxification unit since even very low cyanide concentrations will destroy the bacteria. Detoxification is usually carried out by 'neutralizing' the salt by means of sodium hypochlorite. The disposal of waste salt must be rigorously controlled in conformity with official regulations.

\textbf{GAS CARBURIZING}

During the last decade carburizing by means of gas has become the most popular method of case hardening. The furnaces are very dependable and relatively simple to operate. Since hydrocarbons enter into the process several reactions take place simultaneously. In addition to the reaction

$$2\text{CO} \rightarrow \text{C} + \text{CO}_2 \quad (6.9)$$

there is also the methane reaction

$$\text{CH}_4 \rightarrow \text{C} + 2\text{H}_2 \quad (6.10)$$

and the water–gas reaction

$$\text{CO} + \text{H}_2 \rightarrow \text{C} + \text{H}_2\text{O} \quad (6.11)$$

The composition of the furnace gas can be determined by analysis and hence the carbon potential may be estimated. The moisture content of the gas has a great influence on the carbon potential as indicated by reaction (6.11) and may be measured by simply determining the dew point of the gas. The dew point is that temperature at which water droplets (dew) are deposited from the gas (saturation). The relationship between moisture content and dew point is shown in Figure 6.100.

F. E. Harris\textsuperscript{12} has described a method of calculating the carbon potential from the composition of the gas. In the following are discussed only a few examples that show, by controlling the dew point, what carbon concentrations are obtainable at the surface by varying the temperature and gas composition. Figure 6.101 shows how the surface concentration of carbon varies with the dew point and the percentage of H\textsubscript{2} at a temperature of 925 °C when the CO content is kept constant at 20%. If a carbon concentration of, say, 0.80%, is aimed at, the dew point should be −3 °C with 60% H\textsubscript{2} or −17 °C with 20% H\textsubscript{2}.

\begin{figure}[h]
\centering
\includegraphics[width=0.5\textwidth]{dew_point_graph.png}
\caption{Relation between dew point and moisture content.}
\end{figure}

Figure 6.102 shows how the temperature and the dew point of a gas containing 40% H\textsubscript{2} and 20% CO influence the concentration of carbon at the surface. At 925 °C, 0.80% C is obtainable if the dew point is about −6 °C. If the temperature is only 815 °C the same carbon concentration would require a dew point of about +6 °C. The calculations are valid only when equilibrium has been established. In a retort furnace this occurs after a few hours' carburization. If a gas of known composition is used it is possible to follow the progress of the process from dew-point determinations.

There are a number of methods in existence for the continuous measurement and control of the carbon potential. One such method is based on infrared rays which are used to register the concentration of CO\textsubscript{2} in the gas. This value gives a measure of the CO concentration which, in turn, relates to the carbon potential. By means of a special measuring unit the resulting impulse can be converted direct to a numerical value of the carbon potential.

The control may also be carried out by measuring the resistance of a steel wire which is calibrated for different carbon concentrations. The resistance measured is proportional to the carbon concentration.

When gas carburizing is being employed, a primary requirement is the availability of furnaces sufficiently gas tight to prevent air ingress, since this would interfere with the process.

In principle there are four different ways of producing the gas, viz.:

1. Dripping carburizing liquid into the furnace. This liquid generally contains ethyl alcohol and white spirit. It drops on to a plate inside
furnace where it is dispersed and vaporized. By this method it is possible
to employ a liquid that produces a carrier gas having a relatively low
carbon potential. This potential can then be adjusted by dripping a more
active liquid into the furnace. In practice, however, only one liquid is
used and the carbon potential is adjusted by varying the volume of
liquid added.

2. Admission of gas from a gas-producing unit (Figure 6.103). The gas is
produced by endothermic combustion. The raw gas, usually propane,
is mixed with air in carefully balanced proportions and burnt in the
gas-producing unit to produce a carrier gas with a carbon potential of
0.13-0.50%. Before it enters the furnace the carrier gas is given an extra
addition of propane, thereby adjusting the carbon potential to the
required level which is usually around 0.8%.

3. Direct admission of gas and air into the furnace. The gas actually doing
the carburization is produced in the furnace chamber by the reaction
between the gas and air which are admitted in balanced amounts.

4. Carburizing in a vacuum furnace. Only hydrocarbons are admitted
into the furnace from which air is evacuated. Since no oxygen takes
part in the reaction there is no internal oxidation. The process can take
place at a temperature higher than that normally used, thus allowing
the carburization to proceed more quickly.
6.3.4 Influence of heat treatment and steel composition on case depth, surface hardness and core hardness

As a result of the carburizing treatment the steel has obtained a certain carbon content in the surface zone. On subjecting the steel to further heat treatment this zone can give it the required surface properties. Case hardening is, however, not merely a matter of surface hardness and depth of penetration. Stipulated requirements may call for a certain correlation between the hardness of the case and the core, which will be discussed in the further text.

DEPTO OF CARBON PENETRATION

In the first instance the depth of hardness penetration depends on the carbon content of the carburized layer. Provided that only martensite is formed in the case on quenching, the depth of case hardening would be equivalent to a depth of carbon penetration down to 0.40% C. This would agree well for small parts but as the section dimensions increase the rate of cooling decreases and hence the conditions necessary for the formation of martensite are changed. This implies that steel components of different sizes but which have the same depth of carbon penetration obtain a case-hardening depth that is dependent on the dimensions of the piece. Figure 6.104 illustrates this fact as applied to BS 637M17. The test specimens were all carburized together and were placed in such positions as would ensure their reaching the carburizing temperature approximately at the same time. A complete set of test specimens was case hardened at each of six different heat-treatment shops. In two instances the hardness was lower than 550 HV in the 145 mm diameter specimen. Results from practical experience also tend to verify that the hardenability of BS 637M17 may be insufficient when the dimensions of a solid component exceed 100 mm. Note that this limitation does not apply to gears since the rate of cooling of a gear tooth is greater than that of a solid shaft.

Obviously the depth of case hardening is also dependent on the quenching medium. Water quenching gives a greater depth of case hardening than oil quenching, particularly in heavy sections. The differences in case-hardening depths for different dimensions are evened out if the quenching takes place in water (see Figure 6.105).

![Figure 6.104. Case-hardening tests with steel BS 637M17. En 352 (Bofors D R 34). Curves show influence of specimen diameter on hardness profile of hardened case after pack carburizing at 900°C, followed by single quench from 820°C in oil. The depth of carbon penetration is about the same for all the specimens.](image1)

![Figure 6.105. Case-hardening tests with steel BS 637M17. En 352 (Bofors D R 34). Curves show influence of specimen diameter on hardness profile of hardened case after pack carburizing at 900°C, followed by single quench from 820°C in water. The depth of carbon penetration is about the same for all the specimens.](image2)

To a certain extent the depth of case hardening depends on the grade of steel. For light sections the differences in case depth are negligible but for heavy sections the depth of case hardening is influenced in some degree by the hardenability of the steel. Therefore, with components that contain heavy
sections and require a large depth of case hardening it may be expedient to use a steel more highly alloyed than is strictly necessary if only the strength requirements of the core were being considered. Owing to the enhanced hardenability of the more highly alloyed steel the carburizing time is shorter. As an example of this there are shown below the carburizing times required to give a 1.5 mm case depth to three different steels of D = 100 mm.

Steel grade  
805M20(En362)  637M17(En352)  815M17(En353)  
Carburizing time (h)  25  13  10

When the hardenability is influenced by the quenching temperature the latter is also a factor affecting the depth of case hardening. This may be seen in Figure 6.106 by comparing the depth of case hardening in the two specimens which were carburized at the same time but were quenched from different temperatures.

The course of the carburizing process may conveniently be followed by means of test bars of case-hardening steel which are taken out of the furnace and quenched in water. The bars are fractured and the fractured surface is inspected at a magnification of about ×10 to assess the depth of case hardening.

When the quenching is performed directly from the carburizing temperature the depth of case hardening is usually somewhat greater than if the rehardening takes place from a lower temperature. For practical reasons it is not always possible to use test specimens of the same dimensions as the parts being case hardened. Figures 6.104 and 6.105 may serve as a guide when a depth of case hardening of about 1 mm is required for both small and large pieces. For example, if the test specimen has a diameter of 10 mm and the workpiece, which is to be given a case-hardening depth of about 1 mm, has a diameter of 100 mm then, assuming that the conditions prevailing when Figure 6.104 was drawn also apply in the present instance, the depth of case hardening in the test specimen must be 1.35 mm.

**Surface Hardness**

As before, the carbon dissolved in the austenite is the decisive factor determining surface hardness. When the carbon concentration at the surface of conventional alloy case-hardening steels exceeds 0.70%, the Ms temperature falls steeply, the amount of retained austenite after quenching increases and the hardness decreases. If the surface layer of the steel contains the appropriate concentration of carbon for maximum hardness the quenching temperature is of minor importance to the hardness provided that the grain size is not altered. By varying the quenching temperature of an 'over-carburized' steel it is possible to control the amount of carbon going into solution and hence the amount of retained austenite which, in turn, affects the hardness.

During the course of an investigation with steel grades BS 655M13 (En 36 A), BS 637M17 (En 352) and BS 805M20 (En 362), specimens 25 mm in diameter × 200 mm were pack carburized for 9 h at 925 °C and quenched in oil. In one test series of specimens the hardness was determined, on a cross-section, from the surface towards the centre and in another the carbon concentration was estimated at 0.1 mm steps below the surface by machining off stock to the appropriate depth, the specimens having been previously annealed for 1 h at 630 °C. The results are shown in Figure 6.106.

The carbon concentration at the surface of the steels is about 1.40%. This
HEAT TREATMENT—SPECIAL

maximum hardness is reduced. For the Cr-Mn steel the maximum hardness occurs at 0.70% C and for the Cr-Mo steel (uppermost curve) at 1% C. H. U. Meyer has coined the term 'surface hardenability' to describe the capacity of different steels to attain a high surface hardness, particularly after direct quenching.

A low surface hardness implies a high content of retained austenite. The case of an over-carburized and hardened steel, the surface hardness of which is 500 HV, will have a microstructure like the one shown in Figure 6.108. This type of microstructure is definitely unsuitable for machine components and tools and makes them unserviceable for ordinary working conditions. It is therefore necessary to adopt measures that will give a maximum or near-maximum surface hardness.

When pack carburization is used the carbon concentration in the case can be controlled to a certain extent by adjusting the activity of the carburizing compound and the carburizing temperature. A low carbon activity in the compound and a low carburizing temperature result in a low carbon content in the surface. However, the carbon content increases with the carburizing time and this is illustrated by the curves in Figure 6.109. If the steel is quenched from a lower temperature, less carbon will be dissolved in the austenite and hardness gradients such as those shown in Figure 6.110 are obtained. Further lowering of the quenching temperature will result in still less carbon being dissolved in the austenite and still higher hardnesses. For light sections, quenching temperatures as low as 770 °C can be used whereby hardnesses as high as 67 HRC are obtained.

Figure 6.107. Influence of steel composition and surface carbon content on hardness obtained by direct quenching from 920 °C. quenching from the hot. (R. Thies)

A similar investigation has been carried out by E. Thies. The curves in Figure 6.107 show the dependence of the hardness on the carbon content after gas carburizing and direct quenching from 920 °C. As the Ni content of the steel increases the figure for the carbon concentration corresponding to...
When carburizing in gas it is possible to control the carbon content and keep it at the desired level, generally 0.7-0.8%. The quenching temperature is not so decisive a factor in this instance. For practical reasons the temperature in the furnace is lowered to about 830 °C before quenching. At this low temperature carbide is formed and hence the amount of carbon dissolved in the austenite is somewhat less. In addition the lower quenching temperature reduces distortion.

It is difficult to decide how much retained austenite will produce the best result in each individual instance. For some machine components, such as gears, it has been found, in practice, that retained austenite in larger amounts than previously considered advisable can have beneficial effects. The surface hardness is obviously influenced by the rate of cooling. When quenching into a martempering bath the hardness falls as the quenching bath temperature is raised. To ensure a hardness of at least 60 HRC the temperature of the bath...
should be about 180 °C. If, because of the presence of too much retained austenite, the hardness has inadvertently become too low, subzero treatment may be resorted to. There is no danger of cracks developing provided that the component concerned has a depth of case hardening that is normal for its size and shape.

It is often difficult to obtain a high surface hardness in alloy case-hardening steels, and a subzero treatment is thus commonly employed. Even unalloyed steels (which are water quenched) may contain appreciable amounts of retained austenite in the surface layers, particularly when the parts are quenched direct after a deep carburizing treatment. As is apparent from Figure 6.111, the surface hardness increases as the quenching temperature is lowered. When quenched from 770 °C the case has a hardness of about 900 HV and this hardness is not affected by a subzero treatment.

CORE HARDNESS

When a gear, for example, is being designed with regard to its flexural strength the mechanical properties of the material of the core form the basis for the calculations. Table 6.17 shows the dependence of core strength on the grade of steel and on the dimension. For example, a comparison between S15 2511 and S15 2512 in the dimension range of 6–11 mm shows that the former steel develops a core hardness of 300–405 HB and the latter 330–450 HB. As the dimensions of the sections increase the hardness differences between the two grades are reduced. The only difference between them in chemical composition is 0.05% carbon. However, this is quite large enough to give rise to substantial hardness differences in the core on quenching. Should a Jominy hardenability curve be available for the heat from which the part in question was made, it should be possible to calculate approximately the hardness to be expected on quenching.

The increase in hardness, particularly noticeable in light sections just below the carburized layer, may give rise to machining difficulties when the material of the core is being machined after the hardening treatment. In such cases the core hardness can be controlled by adjusting the temperature of quenching. Figure 6.112, which applies to S15 2511 (0.15% Cr), shows the effect produced by varying the quenching temperature. There is a measured difference of 60 HV between the specimen quenched from the highest and from the lowest temperature. After being quenched from 780 °C the specimen contained about 50% ferrite; after being quenched from 830 °C the ferrite content was judged to be 5%.

A similar investigation has been carried out on a plain carbon steel, S15 1730, the composition in this instance being 0.16% C, 0.32% Si and 0.84% Mn. Test specimens with diameters between 10 mm and 50 mm were carburized at 925 °C to a penetration depth of about 1 mm. One set of test specimens was quenched direct into water from the carburizing temperature. Another two sets were quenched in water; one from 830 °C and the other from 770 °C. The hardness transverse curves are shown in Figure 6.113. It can be seen that the hardness is most influenced by the sectional dimension.
HEAT TREATMENT—SPECIAL

After being quenched, the largest and the smallest specimens show at the centre of their sections a hardness difference of almost 200 HV.

6.3.5 Recommendations for case hardening

From practical experience gained in this field the following general heat-treatment recommendations are suggested for case-hardening steels. The treatment is designed to give a surface hardness of at least 60 HRC or 710 HV after quenching.

CARBURIZING IN SOLID COMPOUND (PACK CARBURIZING)

Temperature: 900–925°C

<table>
<thead>
<tr>
<th>Depth of case hardening mm</th>
<th>Method of hardening</th>
</tr>
</thead>
<tbody>
<tr>
<td>&lt; 0.40</td>
<td>Direct quench</td>
</tr>
<tr>
<td>0.40–1.25</td>
<td>Single quench from 800 to 820°C</td>
</tr>
<tr>
<td>&gt; 1.25</td>
<td>Double quench</td>
</tr>
</tbody>
</table>

When direct quenching is employed a carburizing temperature of 900°C is recommended. If a case-hardening depth of more than 1.25 mm is aimed at it might be good practice to perform the traditional practice of a double quench, i.e. a first quench from about 880°C and then a final quench. During the first heating for quenching, the carbon concentration is reduced by diffusion. In addition it breaks up and dissolves the carbide network that usually forms when pack carburizing is used to give a deep carbon penetration.

CARBURIZING IN A SALT BATH (LIQUID BATH CARBURIZING)

Temperature: 850–900°C

<table>
<thead>
<tr>
<th>Depth of case hardening mm</th>
<th>Method of hardening</th>
</tr>
</thead>
<tbody>
<tr>
<td>&lt; 0.50</td>
<td>Direct quench</td>
</tr>
<tr>
<td>0.50–0.75</td>
<td>Requench from 820°C</td>
</tr>
<tr>
<td>0.75–1.00</td>
<td>Requench from 800°C</td>
</tr>
<tr>
<td></td>
<td>(Double quench)</td>
</tr>
</tbody>
</table>

For direct quenching, a carburizing temperature of 850–870°C is recommended; for double quenching, 900°C.
CARBURIZING IN GAS (GAS CARBURIZING)

Temperature: 900–940 °C

The carburization is carried out to the required depth of case hardening and to a surface carbon concentration of 0.70–0.80%. When the process is completed the temperature in the furnace is lowered to 830 °C and the part is then quenched in a suitable medium.

TEMPERING

Case-hardened steels are tempered at temperatures generally around 160–220 °C. Temperatures below 160 °C should not be used, particularly if a grinding operation is to follow, since grinding cracks develop very easily. Tempering is not necessary after a cyaniding treatment that gives a case depth of only a few tenths of a millimetre.

The hardness falls quite rapidly when the steel is tempered between 160 °C and 200 °C. If a hardness of 60 HRC is required the tempering temperature should not be higher than 180 °C.

6.3.6 Case hardening of tool steels

The wear resistance of tools made from certain medium-carbon steels may be increased by case hardening. This treatment is very often applied to simple punches, for instance, made from S1. When such tools are being hardened they are heated in a cyanide bath at 900 °C, i.e. the normal hardening temperature for the steel, for some 10–30 min longer than the usual holding time and then quenched in the customary way in water or oil.

Hot-work tool steels are case hardened for certain applications. First, the tools are carburized at about 900 °C, generally in a solid compound, and then allowed to cool in the compound. If the tool is to be put to hot-work service it is quenched from 1000 °C. For cold-work applications it is better to use a lower quenching temperature, 980 °C for example. This gives the steel a somewhat higher impact strength.

Figure 6.114 shows the results obtained from a case-hardening test with H13. First, the specimens were pack carburized for 3 h at 900 °C. They were hardened by quenching from 1000 °C and then tempered at temperatures up to 650 °C.

It has been observed both from laboratory tests and in actual practice that H13 has a definite tendency to give irregular case-hardening results. This might possibly be due to the high Si content of the steel. On the other hand consistent results are always obtained when the steels Bofors QRO 45 and Bofors QRT 71 are case hardened. Figure 6.115 shows the hardness transverse curves in Bofors QRT 71 after it had been case hardened and then tempered at various temperatures.

The heat treatment just discussed, involving low quenching temperatures,
they have been quenched from the carburizing temperature they must be tempered. After machining the parts are hardened in the usual way.

2. Protective pastes of "NO CASE" type are applied to those surfaces that are to remain soft. These pastes are made up of a powder that contains mainly copper suspended in a varnish-type binder. The pastes are not suitable for use with gas or salt-bath carburizing. Specially compounded pastes are available for gas carburizing.

3. Electrolytic copper plating to a thickness of 30–50 μm. It is important to ensure that the parts are very carefully cleaned before being copper plated and that the copper bath itself is clean. An alkaline bath gives a more compact and even coating than an acidic one. Any pores in the copper deposit may be detected by means of a reagent consisting of 10 parts K₄Fe(CN)₆, 10 parts of NaCl and 20 parts of gelatine dissolved in distilled water. A filter paper is moistened with the solution and placed on the copper deposit. If there are any pores blue spots will appear on the paper in 3–5 min.

Those surfaces that are to be carburized can be laid bare in different ways:

(a) By machining off the copper deposit. In this case a machining allowance equal to the depth of cut must be planned beforehand. Great care must be exercised so that the copper coating is not damaged on those surfaces that are to be protected by it.

(b) By locally coating the surface with an insulating varnish or by sticking on a special tape before copper plating.

(c) By locally dissolving the copper deposit by chemical means. This method consists of painting the surfaces that are to remain soft with a special varnish (Bososol B 40) after the copper plating is completed. A red pigment is added to the varnish for detection purposes. When the varnish has dried the copper deposit is dissolved from the unprotected areas by immersing the parts in a solution containing 5 l water, 2.4 kg chromic acid and 125 ml concentrated sulphuric acid. Before commencing the carburizing treatment the protective varnish is dissolved by exposing it to trichlorethylene vapour for 2 min. Figure 6.116 shows a component that has been treated according to the above.

Figure 6.116. Selective carburized part shielded by copper plating

Electrolytic copper plating also serves to protect a surface against carburization by a cyanide bath of the 10% NaCN type. Higher concentrations of NaCN will cause the copper deposit to dissolve fairly quickly. Even a low concentration of cyanide will gradually dissolve copper. Regular carburization of copper-plated parts in a cyanide bath should be avoided since the copper that is dissolved in the salt will cause the bath gradually to lose its carburizing power.

6.3.8 Choice of case-hardening depth

When steel is case hardened it develops increased surface hardness and wear resistance, greater ability to withstand flexural and torsional forces and enhanced fatigue strength. The depth of case hardening is of decisive importance to several of these properties. If the surface of the steel is subjected to adhesive wear at low loads a case depth of some tenths of a millimetre is generally sufficient. As the load increases the case depth must also be increased.

The pressure applied to a case-hardened surface is propagated into the material. The most dangerous of the stresses thus created is considered to be the shear stress. In a section through a stressed surface the largest shear stresses act along a semicircle emanating from the edges of the surface to which the load is applied. If the same unit stress (N/mm²) is applied, the
The choice of case depth is often a matter of experience and is rarely the result of theoretical considerations. In actual practice, calculations are made complicated by the fact that the load is seldom evenly distributed across the surfaces of contact between two bodies. For case-hardened gears the design load is proportional to the gear module. The depth of case hardening is usually taken to be 0.15–0.18 times the module. Case depths of more than 2.0 mm are rarely encountered in gears.

Fatigue fractures that occur in unnotched highly polished test specimens do not start at the surface but originate instead in the boundary area between the core and the hardened case. This phenomenon may be explained by means of the schematic stress diagram shown in Figure 6.118 which represents a cross-section through the plane of the surface and the neutral axis. $H$ represents the strength of the material of the core and of the hardened case. If $OP$ represents the stress acting across the section the strength, $H$, of the steel is not exceeded. A larger stress, $O P'$, exceeds the strength of the material in the outer zone of the core (see the black area) and hence a crack appears here. To avoid this a higher core strength $H'$ or a larger case depth should be chosen.

![Figure 6.118. Schematic stress diagram for section between surface and neutral axis.](image_url)

The fatigue crack initiated in the boundary area is quickly propagated through the body of the core thus inducing a progressively increasing stress in the case. Cracks begin to appear here and ultimately the test specimen fractures suddenly. From this we see that the strength of the core has a decisive influence on the fatigue limit owing to the fact that the cracks are usually initiated in the core, the fatigue limit of which is lower than that of the carburized case.

There is, however, an optimum value of the core strength which, if it should be exceeded, will cause a reduction in the fatigue limit. In step with increasing contents of the alloying elements—carbon in particular—there is a successive lowering of the transformation temperature range of the austenite during its cooling from the hardening temperature of the steel. The strength of the core is hereby increased but the longer the time lapse before the transformation takes place in the core the smaller will be the compression stresses in the case. This implies a lower fatigue limit and a greater notch sensitivity of the latter. For maximum resistance to fatigue fracture the optimum tensile strength of the core after case hardening is usually in the region of 100–130 kp/mm² (980–1280 N/mm²).
6.4 NITRIDING

The carburizing treatments which were discussed in the previous section are austenitic thermochemical treatments, in that they involve the diffusional addition of the interstitial alloying elements into the austenite and rely on the subsequent transformation of the austenite to martensite to produce high surface hardness. The process of nitriding, however, is a ferritic thermochemical treatment and usually involves the introduction of atomic nitrogen into the ferrite phase in the temperature range 500–590 °C and consequently no phase transformation occurs on cooling to room temperature. The method was first used at the end of the 1920s and since then its application has continuously spread; due among other things to the fact that the process has been further developed and can now be applied to a much larger number of steels than was originally thought possible.

![Equilibrium diagram for iron-nitrogen](image)

*Figure 6.119. Equilibrium diagram for iron-nitrogen (M. Hansen)*

The properties imparted to steel by nitriding can be summarized as follows:

1. High surface hardness and wear strength, together with reduced risk of galling.
2. High resistance to tempering and high-temperature hardness.
3. High fatigue strength and low fatigue notch sensitivity.
4. Improved corrosion resistance for non-stainless steels.
5. High dimensional stability.

These properties will be dealt with later in this chapter; dimensional stability, however, will not be discussed until Chapter 7. The main factors which govern the depth of the nitride case are treatment time, temperature, nitrogen activity and steel composition.

When studying the nitriding process we can make use of the iron–nitrogen equilibrium diagram (see *Figure 6.119*). At the nitriding temperatures customarily used, the nitrogen will dissolve in iron, but only up to a concentration of 0.1%. When the nitrogen content exceeds this value, γ'-nitride is formed, the chemical formula of which is Fe₄N. If the nitrogen concentration exceeds about 0.6%, the γ'-nitride starts to change into ε-nitride. Below 500 °C ε'-nitride begins to form. The nitrogen content of this phase is about 11% and its chemical formula Fe₃N.

When making observations in the metallurgical microscope the γ' and ε-nitrides can be seen as a white surface layer, called the 'white layer'. *Figure 6.120*. Simultaneously with the increase in thickness of the white layer during nitriding, the nitrogen diffuses further into the steel. When the

*Figure 6.120. White layer and diffusion layer with precipitated acicular iron-nitride in steel containing 0.15% C after gas nitriding for 10 h at 500 °C, 500×*
solubility limit is exceeded, nitrides are precipitated at the grain boundaries and along certain crystallographic planes.

Among the alloying elements, Al, Cr and Mo are used as nitride formers. Nitrogen together with carbon, forms carbonitrides. The amount of carbon nitride can be increased by simultaneous pick-up of nitrogen and carbon by the steel.

Basically, all steels can be nitried\(^1\) and for further details regarding the different types of steel, see p. 407. The high surface hardness which is obtained after nitriding is due to the formation of finely dispersed nitrides and carbonitrides which distort the ferrite lattice.

### 6.4.1 Methods of nitriding

The three main methods of nitriding are gas nitriding, salt-bath nitriding and powder nitriding. There are variations within these main methods and the commonest variations will be described.

#### Gas Nitriding with Ammonia

In the original method of gas nitriding, ammonia is allowed to flow over the parts to be hardened, normally at about 510 °C. The ammonia dissociates in accordance with the equation:

\[
2\text{NH}_3 \rightarrow 2\text{N}_2 \text{+ 3H}_2
\]

At the instant of dissociation, nitrogen occurs in the atomic form, and as such it can be absorbed by the steel. Figure 6.121 gives a schematic representation of the dissociation reaction and nitrogen pick-up in the steel.

![Figure 6.121. Dissociation of ammonia and nitrogen pick-up in steel during gas nitriding](image)

Gas nitriding is used for parts that require a case depth between 0.2 and 0.7 mm. Nitriding is undertaken in an electric furnace equipped for precise temperature control. For large-scale nitriding pit furnaces with specially designed containers provided with tightly closing lids are used (see Figure 6.122). For small-scale nitriding operations a conventional muffle furnace with a separate nitriding box, as shown in Figure 6.123, can be used. Aluminium sheet or asbestos yarn is used as the caulking material between the flange and cover of the box.

The boxes used for nitriding should be made from a material which does not react with the gases. Nicked, inconel and similar alloys are ideal, but heat-resisting steels with 25% Cr and 20% Ni, particularly, have also proved quite suitable for this purpose. If only a few nitriding operations are to be carried out in an ad hoc installation, the nitriding boxes can be made from iron sheet. However, the iron will react with the gas and hence it is difficult to control the processes taking place during the nitriding operations in a newly made nitriding box as some of the gas is used up in nitriding the inner surfaces of the box.

The gas always contains a small quantity of water vapour, which must be removed before the gas is introduced into the nitriding box, otherwise the water vapour can cause oxidation of the nitrided parts. This removal is best achieved by passing it through a filter of unslaked lime, which should be...
replaced or regenerated by heating it at 1000°C before all the lime has become slaked.

To check that the ammonia gas in the nitriding box has decomposed (dissociated) according to the requirements of the process, the composition of the exit gas is determined by means of a dissociation pipette. The sketch in Figure 6.124 shows the general arrangement of one type of pipette customarily employed. During the course of the nitriding process, or preparatory to making a measurement, the gas passes through the taps C and D. During the measurement these taps are closed after which tap E is opened. The

![Figure 6.123. Standard-type muffle furnace used for nitriding](image)

![Figure 6.124. Dissociation pipette for nitriding](image)

water contained above at B is then sucked into the meter because the undisassociated ammonia dissolves in the water, which then takes up precisely the volume occupied previously by the ammonia. The remaining gas, i.e. the dissociated ammonia, which consists of nitrogen and hydrogen, does not dissolve in the water. Figure 6.125 shows some examples of height of the water column for different degrees of dissociation.

![Figure 6.125. Height of water column in meter for different degrees of dissociation](image)

The parts to be nitrided should be well cleaned and degreased. Under no circumstances should there be any traces of rust or mill scale. If certain areas of the part are to remain soft, they can be given an electrolytic coating, e.g. of copper, tin or nickel, which prevents the nitrogen from diffusing into the steel. "Stopping-off" agents containing tin are also available, and these are painted on to the surfaces to be protected. Since the layer of tin is molten at the nitriding temperatures, it is most important that the coating should be so thin (not exceeding 0.01 mm) that it does not spread to surfaces that are to be nitrided.

The parts are placed inside the nitriding boxes in such a way that all surfaces come into contact with the gas. After the box has been closed, it is purged with ammonia gas until all the air has been expelled, after which it is placed inside the furnace. The composition of the exit gas is checked regularly, using the dissociation pipette, and for the first 5–10 h it is kept at 15–20% residual gas. If determinations indicate a lower value than this, the gas velocity is too high and should be reduced. If the converse should prevail, the gas velocity must be increased. When, after adjustment, the system has arrived at a steady-state condition, it is enough to check the gas composition two or three times per day, a residual gas content of about 50% being regarded as a suitable value. Formerly a value of about 30% residual gas was regarded as satisfactory, but this can result in a more brittle nitrided layer. For further details on this point see p. 434.

When the nitriding is completed, the box is removed from the furnace without interrupting the flow of gas. After the charge has cooled to 200°C, the gas supply is cut off and the gases remaining in the box are most conveniently expelled with compressed air before the box is opened. The nitrided parts now normally exhibit a characteristic matt grey colour. If on occasions the colour has shades of yellow, blue or purple, it does not mean that the
nitriding procedure has been at fault. The various tints derive from the presence of oxygen in the system, which may have originated from incomplete drying of the gas, or from some leak in the box or in the supply tubing.

**GAS NITRIDING WITH AMMONIA AND NITROGEN OR HYDROGEN**

Although this method has very seldom been used it is included here for the sake of completeness. Minkevic and Sorokin recommend a mixture of 20% ammonia and 80% nitrogen. The low ammonia content gives a low nitrogen activity, and this results in a tougher case. However, similar results can be achieved by increasing the degree of dissociation when nitriding with ammonia only or by adding hydrogen.

**GAS NITRIDING WITH AMMONIA AND HYDROCARBONS**

When gas nitriding with ammonia, hydrocarbons can also be added, usually in the form of pure propane or endogas, produced from propane and air. Nitriding is performed at a somewhat higher temperature than normal, e.g.

![Isothermal section of the Fe-C-N phase diagram at 565°C (after Nauman and Langenscheid). C = cementite.](image)

Figure 6.126. Isothermal section of the Fe-C-N phase diagram at 565°C (after Nauman and Langenscheid). C = cementite.

570°C. The carbon diffuses into the steel at the same time as the nitrogen and together they form a carbonitride mainly of the ε-type which is illustrated in Figure 6.126. The composition and micro-hardness of the surface layer of pure iron which has been nitrided in an atmosphere of 50% ammonia and 50% propane at 580°C for 4 h is shown in Figure 6.127. The high carbon content in the surface layer gives improved wear resistance and less risk of scuffing.

**GLOW-DISCHARGE NITRIDING OR IONITRIDING**

The glow-discharge nitriding process was patented by Berghaus at the beginning of the 1930s. For many years this was not used extensively on an industrial scale, because relatively expensive equipment and diversified technology is essential if the process is to be carried out correctly. Thanks to development work during recent years, the process is gaining in popularity.

Nitriding is carried out in dilute ammonia gas or in a mixture of hydrogen and nitrogen. Hydrocarbons can be added to the gas mixture in the same way as during normal gas nitriding, and ε-carbonitride is formed in the white layer. The process can also be controlled in such a way that the surface layer is decarburized, which reduces the brittleness of the layer.
It is possible to obtain solely γ'-nitride or ε-nitride, or mixtures of both. The parts to be nitrided are connected as a cathode to an electric power source, whereby a glow discharge takes place at the surface of the steel. As a result of ionization, atomic nitrogen is liberated which can then diffuse into the steel in the usual way. Since the temperature required for the process is reached at the steel surface almost simultaneously with the commencement of discharge, glow-discharge nitriding takes place more rapidly in the initial stages of the process than does normal gas nitriding, in which nitrogen absorption starts only when the furnace and the charge have reached the desired temperature, which can take 1–5 h. Differences in the depth of nitriding are reduced by increasing the nitriding time and, after 40 h for example, it is difficult to observe any difference in case depth between material which has been glow-discharge nitrided or conventionally gas nitrided.

The method of glow-discharge nitriding must be regarded as the technically most advanced process in existence at present, since it enables the development and growth of the nitride layer to be kept under the closest control.

SALT-BATH NITRIDING

The nitriding of steel in a bath of molten salt is almost as old as the original gas nitriding method. The salt mixtures initially used contained 60–70% by weight NaCN and 30–40% by weight KCN. In addition there are a few per cent of carbonates (Na₂CO₃) and cyanates (Na₂CN₃O). By ageing the bath at 573 °C for 12 h (see Equation (6.6) on p. 350), the cyanate can be raised to the desired level, about 45%. Steel parts should not be immersed in the bath during the ageing stage. The normal working temperature for salt-bath nitriding is 550–570 °C and the holding time seldom exceeds 2 h. The salt bath gives off carbon and nitrogen in accordance with Equations (6.7) and (6.8) on p. 350.

Up to the middle of the 1950s, nitriding salt baths were used only on a very restricted scale and then mainly for short-period nitriding (10–30 min) of twist drills for example. If longer nitriding periods were used, pitting occurred on the steel surfaces and in addition the results of the nitriding were irregular. Both of these phenomena were due to variations in the cyanate content at different levels in the bath and also to the solution of iron from the crucible, resulting in oxidation of the surface of the steel parts comprising the charge. During the last decade the method has become more popular because, by injecting air into the bath, it has become possible to obtain better control of the cyanate content. A further improvement has been achieved by employing a titanium crucible, thus avoiding decomposition of the salt which took place previously as a result of contact with the iron crucibles hitherto used. The method which has been developed by DEGUSSA is called the Tufftride process (German: Tufiferbehandlung). Extensive literature has been published on the Tufftride process.

Another variation of salt-bath nitriding is 'Sulfinox treatment', in which the active constituent in the bath, apart from NaCN and Na₂CO₃, is sodium sulphide (Na₂S). During conventional salt-bath nitriding, nitrogen and carbon are absorbed by the surface of the steel. During the Sulfinox treatment, sulphur is present and this enhances still further the excellent anti-friction properties of the nitrided layer. Figure 6.128 shows the composition of the surface layer of a carbon steel containing 0.20% C after Sulfinox treatment at 570 °C for 2 h.

In some cases of Sulfinox treatment it is possible for light corrosion of the steel surface to take place, and this can amount to a loss of 0.005 mm from the surface. Furthermore a thin high-sulphur layer can be formed on the outside of the steel parts. This layer should not be ground off, because it has high anti-friction properties. Should it have to be removed for dimensional reasons, this is best done with steel wool. Normally, grinding should not be performed after salt-bath nitriding.

After Tufftride or Sulfinox treatment the best surface is obtained if the steel parts are quenched in warm water. This rapid cooling results in a supersaturated solid solution of nitrogen in α-iron, which adds an extra contribution to the increase in fatigue strength which is a characteristic feature of nitriding. However, the high internal stresses introduced at the same time reduce toughness. With regard to plain-carbon and low-alloy steels, if these are tempered at 200 °C for 1 h, nitride precipitation takes place which reduces the fatigue strength and increases toughness. During slow cooling, as with gas nitriding, this precipitation is obtained automatically (see Figure 6.120). It is essential to make a daily check on the cyanate content of the salt bath (also the sulphur content for the Sulfinox bath). The deposits which form in the bath and which sink to the bottom of the crucible should be collected and removed by means of a scoop. The entire bath should be renewed at regular intervals.
POWDER NITRIDING

In powder nitriding the parts to be treated are packed into boxes in roughly the same way as when pack carburizing. About 15% by weight of an energizer is first placed on the bottom of the box, after which is laid a layer of nitriding powder on which the parts are placed. (The parts must not come into contact with the energizer.) More nitriding powder is then added, one layer of powder alternating with one layer of parts. The boxes are then closed as tightly as possible and are placed in a muffle furnace at a temperature of 520–570°C. A nitriding period not exceeding 12 h is recommended.

Nitriding tests on steels BS905M39 (En 41 B), BS817M40 (En 24) and H13, gave roughly the same values for hardness and case depth as those obtained during gas nitriding for corresponding times and temperatures. After powder nitriding, the steel surfaces can exhibit some pitting and spalling.

6.4.2 Comparison between gas and salt-bath nitriding

In the preceding section we have described gas and salt-bath nitriding without making any direct comparisons between these methods. To some extent the differences between the two methods have been exaggerated and many statements which must now be regarded as outdated are still cited in literature references. The statements usually made when comparing the two methods are as follows:

1. Gas nitriding requires 12–120 h nitriding time whereas salt-bath nitriding requires 1–4 h.
2. Gas nitriding makes the steel brittle, but this is not the case with salt-bath nitriding.
3. The reason why steel becomes brittle during gas nitriding is the presence of the brittle tri-nitride which is not present after nitriding in the salt bath.

The usual reason for the statement that gas nitriding makes steel brittle is that gas-nitrided aluminium-alloyed steel is compared with plain-carbon or low-alloy steel that has been nitrided in a salt bath. It is difficult to find any test results in the literature relating to salt-bath nitrided aluminium-alloyed steel, and it is questionable whether any data on this subject have actually been published.

In other contexts the same steel has been employed for purposes of comparison, in such tests the case thickness of the gas-nitrided steel is considerably greater.

To permit a closer study of the circumstances prevailing during gas and salt-bath nitriding, a number of tests were made from quenched and tempered Cr-Ni-Mo-steel, BS817M40 (Bofors CRO 861), and then nitrided; one test was carried out in gas for 10–40 h at 510°C, with 50% degree of dissociation, and a second test in a Tufftride salt bath, for 2 and 4 h at 570°C.

SURFACE HARDNESS AND DEPTH OF HARDENING

After nitriding, taper sections were ground on a plane specimen surface and the hardness was measured in H V, load 1 kp.

After gas nitriding at 510°C, the surface hardness was found to be about 675 H V. Nitriding at 550°C gave a lower surface hardness of about 575 H V, but on the other hand the depth of nitriding was greater. Nitriding for 20 h at 550°C corresponds to 40 h at 510°C.

The specimens after the Tufftride treatment gave somewhat differing results, depending on where they were treated. In one test the resulting difference in case depth between the 2 and 4 h treatment was negligible, but in another instance there was a marked difference. In both instances the surface hardness was about 650 H V.

As shown in Figure 6.129, a comparison between gas-nitrided and Tufftrided specimens shows that 10 h of gas nitriding at 510°C corresponds very closely to 4 h nitriding in a salt bath at 570°C.
HEAT TREATMENT—SPECIAL 389

TESTING FOR BRITTLENESS

It is difficult to find a reliable method for assessing the brittleness of the white layer. In the present case the assessment was made by means of a metallographic study of a section through the white layer. In addition the hardness (H.B. 5750) of the specimens was measured and the appearance of the indentations was studied at 50 x magnification. From this study the following facts emerged:

During gas nitriding the thickness of the white layer increased with increasing time and temperature. However, the character of the layer remained homogeneous throughout. On performing the Brinell hardness test on a specimen after it had been nitrided for 10 h at 510°C, both radial and concentric cracks were observed. After nitriding for a longer period or at higher temperatures concentric cracks only were observed, the width of which decreased with increasing nitriding time. This is probably due to the enhanced load-bearing capacity of the nitrided layer.

The Tufftride treatment also caused the thickness of the white layer to increase with treatment time. After a treatment of 4 h the layer contained pores which could clearly be distinguished at 500 x magnification. According to Finnern the normal phenomenon that occurs even after a 90 min treatment. Brinell hardness testing gave rise to radial and concentric cracks after both a 2 h and 4 h Tufftride treatment.

Figure 6.13a shows the cross-section and test indentations in gas-nitried and Tufftrided specimens. On the basis of these test methods it is difficult to make any assertions as to which method of nitriding gives the toughest surface layer.

NITRIE FORMATION

The composition of the nitride layer in the nitrided specimens was established by means of X-ray diffraction (Cu-radiation). Figure 6.13a shows original curves for some of the gas-nitried specimens. After gas nitriding for 20 h and 40 h no reflection was obtained for a-Fe. As a result of the treatment time there was a reduction in the amount of γ'-FeN and an increase in the amount of ε-FeN. Gas nitriding at 550°C for 10 h and 20 h gave similar results.

After the 2 h Tufftride treatment at 570°C an indication of a-Fe could be detected. Both γ'-FeN and ε-FeN were also present. As a result of a 4 h treatment at 570°C, the a-Fe disappeared, the γ'-FeN disappeared almost completely, but the ε-FeN increased (see Figure 6.13b). The investigation was repeated using Cr-radiation, and basically the same results were obtained.

These investigations have shown clearly that γ'-FeN is present both after gas nitriding and salt-bath nitriding at the temperatures employed. Nevertheless, the amount of γ'-FeN is greater after gas nitriding than after salt-bath nitriding.

Summarizing, the investigations made have shown that none of the three

Figure 6.13a. Cross-section and test indentation marks on gas-nitried and tufftrided specimens of BS 817M40 (Before CRO 681). Gas-nitriding 10 h, 510°C. Tufftriding 4 h, 570°C.
Figure 9.131: X-ray diffraction diagram of gas-nitrided BS 817M40 (before CRO 801)
HEAT TREATMENT—SPECIAL

Statements cited by way of introduction, and which derive from the published literature, turns out to have any validity whatever. The opinion, often put forward, that low-alloy steel should be nitrided in a salt bath only and not gas-nitrided, can thus be disregarded.

ADVANTAGES AND DISADVANTAGES

The following additional points can be made for the two methods.

1. **Treatment time**
   In this respect, salt-bath nitriding has its obvious advantages when compared with conventional gas nitriding. This applies particularly if a nitriding depth of 0.1 mm (at 400 H v) is adequate, which is normally obtained after salt-bath nitriding for 2 h with the steel investigated.

2. **Cleanliness**
   With gas nitriding there is no soiling of the work premises or of the nitrided components. However, with salt-bath nitriding, it is quite easy to spill salt on the top of the furnace and on the surrounding floor area. Furthermore, parts treated in the salt bath must be cleaned by rinsing in hot water. The cyanide salt is toxic, but cases of poisoning are very rare provided that the correct procedure is followed. Since the rinsing water together with residues of salt can easily be flushed into the drainage system, we can expect that in future the Nature Conservancy Authorities will impose more stringent regulations. Even now, traces of cyanide must be detoxified in districts where the effluent water is subjected to bacteriological purification. The reason for this is that the cyanide destroys the bacteria that are used for purification purposes.

3. **Depth of nitriding**
   As a result of the formation of pores, which increases with increasing treatment time during salt-bath nitriding, the nitriding period is restricted to 4 h. No such restrictions are imposed on gas nitriding, but for practical reasons a nitriding time of 90 h should normally not be exceeded. Often a greater depth of nitriding is required than that obtainable with salt-bath nitriding, and this is when gas nitriding comes into its own.

4. **Dimensional stability**
   For the same depth of nitriding, the resulting distortion is the same for both methods, if by this we mean the change in dimensions occurring as a result of the compressive stresses induced in the steel during nitriding.

After gas nitriding, the box containing the nitrided parts is removed from the furnace and allowed to cool freely in air, whereas components which have been nitrided in a salt bath are quenched in water or oil after the treatment. As a result of the water quench thermal stresses are created which can result in changes of shape.

Figure 5.15b: X-ray diffraction diagram of nitrided 85 873440 (Bifor CRO 861)
5. Wear resistance

During salt-bath nitriding a higher content of e-carbonitride is formed than during conventional gas nitriding. In critical cases this higher content of e-carbonitride provides increased wear resistance and reduced risk of scuffing. By adding hydrocarbons during gas nitriding the content of e-carbonitride can be raised to the same level as that obtained during salt-bath nitriding.

6. Toughness

As a result of the rapid cooling employed during salt-bath nitriding, the toughness of the nitried layer is inferior to that obtained by gas nitriding, where cooling takes place slowly. The practical consequences of this have been observed, for instance, when comparisons were made between gas-nitrided and salt-bath nitrided moulds made from quenched and tempered steel. Several instances of spalling were observed after salt-bath nitriding.

What has been said above indicates that the shorter treatment time and the resultant lower costs represent the only definite advantage that salt-bath nitriding can offer vis-a-vis gas nitriding, which possesses many advantages within other spheres. Hence the heat treater as well as the user of nitrided material has to select which method is the most favourable one for his special or general case.

6.4.3 Nitridability

The concept of 'nitridability' includes on the one hand the ability of the steel to absorb nitrogen, and on the other the increase in hardness imparted by the nitrogen.

The effect of the alloying elements has been described in Chapter 3. For the sake of convenience the diagram shown previously is reproduced in Figure 6.132. When deciding on the composition of a nitriding steel for which a high surface hardness is required, the choice falls mainly on those elements that form nitrides, such as Al, Cr and Mo. The diagram shows how these elements along with a few others affect the surface hardness after nitriding. Al, closely followed by Ti, has the greatest effect, and then come Cr, Mo and V. In the Ni steels the same surface hardness has been obtained as in plain-carbon steel (about 400 HV). In steels containing several alloying elements higher hardness values are obtainable than if the alloying elements are used separately.

The depth of nitriding decreases with increasing content of alloying element, as shown in Figure 6.133. Al and Ti, which have the greatest effects in increasing the hardness, also have the strongest delaying influence on the diffusion of nitrogen into the steel as their contents increase. An optimum value for hardness and depth of nitriding is obtained at roughly 1.5% Al, and this is also the content normally used in Al-alloyed nitriding steels. The reason for the inhibiting effect of the alloying elements is that they bind the nitrogen as nitrides. Carbon, too, has a strong inhibiting effect on the diffusion of nitrogen.

The microstructure influences nitridability in the following two ways. A high content of free ferrite favours diffusion of nitrogen. A low carbide content in the structure favours both the diffusion of nitrogen and hardness. Usually alloy steels in the heat-treated state are employed for nitriding. i.e.

![Figure 6.132: Influence of alloying elements on hardness after nitriding. Base alloy 0.35%, C, 0.80% Si, 0.70% Mn](image)

they have been quenched and tempered at 500–650°C. Within this temperature range both precipitation and coagulation of carbides takes place. Since carbide precipitation starts preferentially at the grain boundaries, a strong barrier is formed at these points, where the diffusion-retarding effect is accentuated because diffusion—which normally takes place more rapidly at the grain boundaries than through the grains—is impeded by the nitrides and carbonitrides formed in the grain boundaries. The effect of the tempering temperature used during the heat treatment is illustrated in Figure 6.134, which applies to steel En 29 B (Bofors RO 7155).

Tests have shown that the nitridability of this steel after (a) induction hardening and tempering at 180°C, and (b) surface decarburization, is better (see Figure 6.135)—than after quenching and tempering (tough hardening).
Hodgson and Baron\(^2\) have made detailed tests on the nitriding of heat-treatable steels containing approximately 3\% Cr and 0.50\% Mo, with carbon contents 0.10-0.37\%, and subjected to varying tempering temperatures. The results are summarized in Figure 6.35, which, among other things, shows that the response of the steel to elevated tempering temperatures is dependent on the carbon content.

The grain size and carbide distribution in the steel influence the surface smoothness obtained after nitriding. With increasing grain size, the nitrides are precipitated in wide bands along the grain boundaries. If carbides have already been precipitated at the grain boundaries, the nitrogen forms carbonitrides with the carbides, which will hereby increase in size. As a result, the carbonitrides can displace individual grains in the surface layer, so that the white layer will become deformed or crack (spall). The photograph in Figure 6.37 gives an idea of the mechanism involved.

Measurements of surface smoothness on nitrided specimens of S1S 2940 (ISO steel 3) show that there is a definite correlation between grain size and surface smoothness, as has been shown by measurements made on material taken from the surface and centre of hardened and tempered rods, roughly 150 mm diameter. As shown in Figure 6.38, the surface smoothness of the nitrided specimens taken from the centre is inferior to the smoothness of those taken from the surface. The centre portions contain considerably
Heat treatment—special

more grain-boundary precipitates than the sections located near the surface, owing to the slower cooling of the centre during hardening. This investigation resulted in a modification of the composition of steel S18 2940 (ISO steel 3), so that its hardenability improved and the tendency to form grain-boundary precipitates in sections exceeding 100 mm was reduced (ISO steel 4). See Table 6.18 p. 408.

![Graph](image)

Figure 6.138. Surface smoothness as a function of grain size of nitrided specimen of ISO steel 3 taken from surface and centre.

Figure 6.139 shows the appearance of the nitrided surface of S18 2940, unmodified composition (ISO steel 3), with different grain sizes. If high surface smoothness is required in the central sections when nitriding large dimensions of ISO steel 3, it is advantageous to carry out the hardening and tempering after the rough machining.

![Images](image)

Figure 6.139. Appearance of surface after nitriding of steels with different grain size, ISO steel 3: (a) Grain size 7.5 ASTM. Surface smoothness 1.5 μm; (b) Grain size 5 ASTM. Surface smoothness 5 μm.

During hot working or heat treatment of steel S18 2940, the surface layer may become decarburized, and results in grain growth. On nitriding, this will have the consequences outlined above. If the machining allowance has been inadequate, spalling will occur during nitriding (Figure 6.140). If stress
relieving takes place at too high a temperature, recrystallization with the formation of columnar grains may occur in the surface layer. This type of grain growth, Figure 6.141, also causes rough surfaces and spalling.

The effect of nitriding time on the depth of nitriding can, for low-alloy steels, be derived from the simple formula for diffusion:

$$D_n = \sqrt{kt}$$

where $D_n$ = depth of nitriding

$k$ = a constant

$t$ = time in hours

The constant is a function of temperature and material.

When studying the nitrogen content at different depths below the surface after salt-bath nitriding of a carbon steel containing 0.15% C, the values shown in Figure 6.142 were obtained. These confirm the validity of the equation shown above quite well, if the calculations assume a constant nitrogen content. Basically, the effect of the nitriding temperature is such that the lower the temperature the higher is the surface hardness but, at the same time, the smaller is the depth of nitriding. This applies particularly to the alloy steels and is shown by way of example in Figures 6.143 and 6.144 which apply to steels En 41 B (Bofors ARO 75) and En 29 B (Bofors R O 7155).
Consequently the nitrogen activity is a function of the number of ammonia molecules dissociated at the steel surface per unit of time. At constant pressure and temperature the degree of dissociation is reduced as the flow rate increases but the product $a \cdot v$ increases. Therefore in actual practice the nitrogen content can be checked by means of the degree of dissociation and controlled by the flow rate.

A high nitrogen activity is thus obtained with high flow rates, which give a low degree of dissociation. A low nitrogen activity is obtained with low flow rates, which give a high degree of dissociation. If the flow rate is too low, the turbulence in the nitriding box is inadequate unless the box is provided with a fan. This can result in uneven nitriding at different locations in the box. The degree of dissociation increases as the temperature increases so that by raising the temperature a constant degree of dissociation can be ensured with a high gas flow rate.

At the commencement of the process it is customary to employ a low degree of dissociation (high gas flow rate), the object being to expel the air present rapidly and to build up a high nitrogen content in the surface layer. The thickness of the white layer with a 30% degree of dissociation increases in the manner shown in Figures 6.145 and 6.146. Layer thicknesses around 0.02 mm or more may be unwelcome because of the risk of spalling, and hence it is essential to control the layer thickness so that it assumes a suitable value.

In 1943 C. F. Floo took out a patent for a method of controlling the thickness of the white layer by means of the degree of dissociation. However,
this method has not been used on the scale anticipated. A modified method however has been developed by Bofors and has been used by them for some years with good results. During preliminary investigations performed using En 41 B (Bofors ARO 75) before the method was taken into use, the degree of dissociation and nitriding temperature were varied. Some of the results of the tests are shown in Figure 6.147. An almost ideal build-up of the nitriding layer is obtained if, after 5 h, the degree of dissociation is increased to 85%. At still higher values the steel becomes de-nitrided.

In actual practice it has proved feasible to operate with a degree of dissociation of roughly 50% even in furnaces without fans. If higher degrees of dissociation are required, it is essential to have furnaces with fans, or to increase the nitriding temperature. An increase in the degree of dissociation from 30 to 50% at 510°C implies a reduction in the amount of ammonia by ca 25%. If the degree of dissociation is increased to 50–60%, considerable increases are achievable in the service life of gas-nitrided components.

6.4.4 Determination of depth of nitriding

The depth of nitriding, designated Dₙ, is determined in accordance with SIS 117009 by making HV measurements on a cross-section through a nitrided surface. The limiting hardness is 400 HV. This value is applicable only to constructional steels of the type included in Table 6.18 on p. 408. At the time of writing there were no ISO standards for the determination of
the depth of nitriding, one reason being the difficulty in deciding on suitable limiting values for carbon steel for which 400 HV is too high, or for tool steels, the basic hardness of which exceeds 400 HV. The depth of nitriding can be assessed with the aid of a fractured surface in the same way as that employed for case-hardening steels. Figure 6.148 shows the agreement obtained between the values assessed and determined at 400 HV for steel BS 905M39 (En 41 B).

6.4.5 Nitriding different types of steel

Some of the types of steel that are not commonly used for nitriding have been discussed in the preceding section, but we will now give details of a few more. ISO R 683/X-1970 includes the following steels, Tables 6.18, 6.19 and 6.20, which are primarily intended for nitriding.

Chromium-molybdenum-alloyed nitriding steels containing approximately 3% Cr and 0.5% Mo are made with carbon contents between 0.20 and 0.40%. One of the more widely used steels is En 29 B (ISO steel 1) which has a hardness profile as shown in Figure 6.149. This steel finds its main use for large machine components such as shafts and gear wheels (see Figure 6.150).

![Hardness vs. Depth Below Surface](image)

Figure 6.149: Influence of tempering temperature on hardness and depth of nitriding of steel En 29 B (ISO 7125). Nitrided for 66 h at 510°C.

Aluminium-alloyed nitriding steels are manufactured with varying compositions, depending on hardenability requirements. One of the rather commonly occurring steels is BS 905M39 (En 41 B), which on nitriding achieves a surface hardness of about 1100 HV and has the hardness profile shown in Figure 6.151. A nitriding period shorter than 5 h is not recommended if the very highest surface hardness is required.

Alloy heat-treatable steels of the type BS 708A37 (En 14 B) and BS 817M40 (En 24) exhibit a surface hardness of around 650 HV after gas nitriding at 510°C, if their initial state was hardened and tempered. If the nitriding temperature is raised to 550°C, the hardness drops to around 550 HV. Basically the graph in Figure 6.152 applies also to BS 768A37.

<table>
<thead>
<tr>
<th>Type of Steel</th>
<th>C %</th>
<th>Si %</th>
<th>Mn %</th>
<th>Cr %</th>
<th>Mo %</th>
<th>Al %</th>
<th>S %</th>
<th>P %</th>
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<td>0.15-0.40</td>
<td>0.40-0.70</td>
<td>0.00</td>
<td>0.035</td>
<td>0.00</td>
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</tr>
<tr>
<td>2</td>
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<td>0.40-0.70</td>
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<td>0.00</td>
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</tr>
<tr>
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<td>0.55-0.80</td>
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</tr>
<tr>
<td>4</td>
<td>0.34-0.45</td>
<td>0.25-0.50</td>
<td>0.65-0.90</td>
<td>0.00</td>
<td>0.035</td>
<td>0.05</td>
<td>0.05</td>
<td>0.05</td>
</tr>
</tbody>
</table>
Alloy case-hardening steels, e.g. BS 637M17 (Bofors D R 34) can be nitrided in the annealed state, whereby the hardness and depth of nitriding obtained are somewhat higher than that achieved with heat-treatable steels (see Figure 6.153).

Unalloyed case-hardening steels and plain low-carbon steels show, after gas nitriding, the hardness profiles given in Figure 6.154. After salt-bath

### Table 6.20 Mechanical Properties of Nitriding Steels in Accordance with ISO

<table>
<thead>
<tr>
<th>Type of Steel</th>
<th>Diameter</th>
<th>$R_y$ min</th>
<th>$R_m$ min</th>
<th>$A$ min</th>
<th>$KCU$ min</th>
<th>Hardness of nitrided surface</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$mm$</td>
<td>$kgf/mm^2$</td>
<td>$kgf/mm^2$</td>
<td>$%$</td>
<td>$kgf/cm^2$</td>
<td>HV min</td>
</tr>
<tr>
<td>1</td>
<td>≤16</td>
<td>90</td>
<td>110-130</td>
<td>10</td>
<td>5</td>
<td>800</td>
</tr>
<tr>
<td></td>
<td>&gt; 16 ≤ 40</td>
<td>85</td>
<td>105-125</td>
<td>10</td>
<td>6</td>
<td>800</td>
</tr>
<tr>
<td>2</td>
<td>&gt;40 ≤ 100</td>
<td>80</td>
<td>100-120</td>
<td>11</td>
<td>6</td>
<td>800</td>
</tr>
<tr>
<td></td>
<td>&gt;100 ≤ 160</td>
<td>75</td>
<td>95-115</td>
<td>12</td>
<td>6</td>
<td>800</td>
</tr>
<tr>
<td></td>
<td>&gt;160 ≤ 250</td>
<td>70</td>
<td>90-110</td>
<td>12</td>
<td>6</td>
<td>800</td>
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<tr>
<td>3</td>
<td>≤70</td>
<td>110</td>
<td>130-150</td>
<td>8</td>
<td>3</td>
<td>800</td>
</tr>
<tr>
<td></td>
<td>&gt;70 ≤ 100</td>
<td>60</td>
<td>80-95</td>
<td>14</td>
<td>5</td>
<td>950</td>
</tr>
<tr>
<td></td>
<td>&gt;100 ≤ 160</td>
<td>75</td>
<td>95-115</td>
<td>12</td>
<td>4</td>
<td>950</td>
</tr>
</tbody>
</table>

* $R_y$ = yield stress (0.2% proof stress)

$R_m$ = tensile strength

$A$ = percentage elongation after fracture ($L_0 = 5d$)

$KCU$ = impact strength with 1" notch

HR 15N = Rockwell superhard (X-axis) hardness with 15kgf load

---

**Figure 6.150. Nitrided gears made from En 29 B (Bofors RO 7155)**

**Figure 6.151. Hardness profile through surface layer of BS 905M39 (Bofors ARO 75) after gas nitriding for 6-80 h at 510°C**
nitriding or gas nitriding with added hydrocarbons, the hardness distribution obtained is roughly the same. However, in the outermost layer the surface hardness is somewhat greater.

*Low-alloy tool steels* type S15 2550 (Bofors HRO 1243) can be nitried either in the annealed or hardened and tempered state. The surface hardness becomes roughly 500 HV.

High-alloy tool steels containing 5–12% Cr of type A 2, D 2 and D 6 follow a common pattern, which suggests that the highest hardness is obtained if, after hardening, the steel is tempered at only 200 °C. The surface hardness is 1100–1200 HV (see Figures 6.155 and 6.156).

If, for example, the hardening temperature is raised to 1050 °C, secondary hardening takes place during nitriding. The hardness of the core obviously drops with increased nitriding time but it is considerably higher than if a normal hardening temperature had been used (see Figure 6.157). A tempering treatment at 510 °C prior to nitriding has not affected the depth of nitriding or the surface hardness. Steel D 6, after hardening from 1050 °C, behaves in precisely the same way during nitriding as steel D 2.

*Hot-work steels* of the types normally employed exhibit surface hardness values after nitriding that lie between 1100 and 1300 HV. Owing to the high
alloy content, the depth of nitriding for these steels is relatively small as is also the case for the high-alloy chromium steels. Since the steels are normally tempered at temperatures above the nitriding temperature, the hardness of the core is not reduced during nitriding. As shown by Figures 6.158 and 6.159 the temperature employed during tempering has a marked effect on the depth of nitriding for steel H 13. A high initial hardness tends to have a favourable effect on the depth of nitriding. Nitriding can be carried out between 510 and 525°C without noticeably affecting the surface hardness.

Nitriding at 550°C gives a surface hardness of about 1 100 H V and a somewhat greater depth of hardening than that obtained at lower temperatures. Tools made from this steel are nitrided on quite a large scale for both cold- and hot-work service. Figure 6.160 illustrates nitrided pistons made from H 13 which are intended for production of plastics under high pressure.
High-speed steel tools are also normally salt-bath nitrided for 10–60 min. With gas nitriding the period should not exceed 10 h. If periods longer than this are employed, the outermost layer becomes soft (see Figure 6.161). Hardness values above 1300 HV are obtainable which may be too high for many purposes. In such cases the tools are tempered at 550–600 °C after nitriding, thereby increasing the toughness.

Stainless steels cannot be nitrided without special pre-treatment, because the film of chromium oxide which forms on the surface of the steel after contact with atmospheric oxygen prevents nitrogen absorption. For martensitic stainless steels the oxide film can be removed or thinned down by a light sand blasting immediately before nitriding.

The same also applies to a certain extent to the 18/8 austenitic steels and similar steels. Better results are obtained with these steels if chemical pre-treatment is carried out in accordance with any of the following methods:

1. Pickling in 50% by volume hydrochloric acid at 70 °C, with subsequent thorough rinsing in water.
2. Pickling in 1% by volume orthophosphoric acid containing 0-7% primary zinc phosphate at 70 °C without subsequent washing.

In both cases nitriding must be carried out directly after the pre-treatment.

If, for practical reasons, this should not be possible, the parts can be dealt with in accordance with any of the following methods:

3. Pickling in 10% hot sulphuric acid, rinsing in water, and copper-plating in a copper cyanide bath to give a coating of thickness 0.3 μm.
4. According to Coppola, 12% Cr stainless steel can be nitrided after lapping the steel surface followed by passivation in hot nitric acid, or heating in air at 510°C. Heating in air at 260–370°C prior to nitriding gives unsatisfactory results.

Stainless steels are nitrided at higher temperatures than those used for other steels. For martensitic stainless steels, 520–550°C is used, and for the austenitic steels, 570–620°C. It is important that the air should be purged from the furnace before heating starts.

Figure 6.162 shows results obtained from nitriding tests on creep-resisting 13% Cr steel at temperatures between 520 and 600°C for 60 h. This graph can also be used for other martensitic steels of the type A13I 430. Figure 6.162

![Graph showing nitriding results](image)

**Figure 6.162** Gas nitriding of 13% Cr steel (Before ROP 46) for 60 h at 520–600°C

As a result of nitriding, there is a considerable reduction in resistance to rusting, and hence it is doubtful whether a stainless steel chosen for its corrosion-resistant properties should be subjected to nitriding. Nitriding of steel A 286 which is employed as a creep-resisting steel can, on the other hand, be quite justifiable. Also, as a result of nitriding, austenitic steels can become slightly magnetic in the nitrided layer, partly because the iron nitride Fe₃N is ferro-magnetic, and partly because of partial transformation of austenite to ferrite in the nitrided layer.

6.4.6 Properties of nitrided steels

**Wear Resistance**

The wear resistance of steels increases with increasing hardness. Since the concept of 'wear resistance' is difficult to define, a warning should be given against any generalizations. If the hardness of a cutting edge should exceed a certain value, small fragments of the edge break away; as machining continues, the edge becomes blunt and a pattern reminiscent of wear arises. The nitrided layer is very liable to chip in this way. In the case of abrasive wear of a scratching or eroding nature, it is doubtful whether nitriding will provide any long-term improvement in wear resistance. With gouging or grinding abrasion, nitriding is not to be recommended. Figure 6.164 shows the results obtained from shot-blasting tests with steel grit on steels of different hardness, the least abrasion being reported for the nitrided steel.

![Graph showing abrasion loss](image)

**Figure 6.164** Abrasion loss occurring during blasting with steel grit on unnitrided, case-hardened and nitrided test specimens respectively (after Lambert)
It is for adhesive wear especially that the main advantages are achieved with nitriding, because the coefficient of friction is reduced, as is also the risk of scuffing. A further improvement is achieved by the Sulfinux treatment, a fact confirmed by practical tests and laboratory investigations. During wear tests in an Amsler machine, in which two discs made from a 0.15\% carbon steel rotated against each other at a speed of 440 and 400 rpm respectively, the values shown in Figure 6.165 were obtained. The specimen given the

![Figure 6.165. Amsler test on discs made from carbon steel with 0.15\% C subjected to different treatments. Loading 20 kN (after Waterfall)\(^{(1)}\)](image)

Sulfinux treatment is superior to both the case-hardened and the nitrided specimens. Scuffing has started at the point where the curves noticeably change their direction. In the initial stages the Sulfinux-treated specimen suffered somewhat more abrasion than the case-hardened or the nitrided specimens because of the wearing away of the black surface coating.

Under certain conditions nitrided steels can function without lubrication, this being particularly marked with Sulfinux-treated steels. This characteristic has been tested in a Faville-Levally machine, in which a test bar rotates at a speed of 300 rpm between two hardened V-jaws. The pressure is gradually raised until scuffing occurs. When tests were made with oil-lubricated specimens made from 0.15\% carbon steel, the untreated bar scuffed at a loading of 380 kN. The Sulfinux-treated bar exhibited no sign of scuffing whatever at the maximum loading of the machine, i.e. 1140 kN (see Figure 6.166). Corresponding tests were made on unlubricated specimens, and the untreated specimen scuffed at a loading of 250 kN. The test with the Sulfinux-treated bar had to be discontinued at a loading of 630 kN because the specimen lengthened as a result of the frictional heat (see Figure 6.167).

![Figure 6.166. Test bars and jaws after Faville-Levally test in oil SAE 30. Untreated (left): Scuffing at pressure of 380 kN. Sulfinux treated specimen (right): No scuffing at pressure of 1140 kN (after Waterfall)\(^{(2)}\)](image)

![Figure 6.167. Test bars and jaws after Faville-Levally test without oil. Untreated (left): Scuffing at pressure of 250 kN. Sulfinux treated (right): No scuffing at pressure of 630 kN, but test bar lengthened as a result of plastic deformation (after Waterfall)\(^{(3)}\)](image)

Practical tests were performed involving the Sulfinux treatment of case-hardened steel parts that exhibited a tendency to scuffing. As an example we can mention a bushing made from case-hardened Cr-Ni steel which started scuffing against the shaft after 7 min at a speed of 4000 rpm. After Sulfinux treatment at 550°C for 2 h the bushing could be run for 2 h without scuffing. Because of the heat of friction generated the machine then started to operate sluggishly. After being allowed to cool to room temperature the running test could be resumed.

**HIGH RESISTANCE TO TEMPERING AND HIGH HARDNESS AT ELEVATED TEMPERATURES**

These are valuable properties of nitrided steels. After nitriding, the steel can be heated up to the same temperature as that at which it was nitrided, and in some cases to an even higher temperature, without loss of hardness at
Nitrided steels are characterized by high fatigue strength and low notch sensitivity. The increase in fatigue strength is highest for plain-carbon and low-alloy steels, and this generally increases as the nitriding time increases. The standard nitriding steels heat treated to an ultimate tensile strength of 90 kp/mm² have a fatigue strength of roughly 45 kp/mm², as determined during rotary bending fatigue tests on highly finished ground test bars.

Room temperature (see Figure 6.168). Figure 6.169 shows that the hardness at elevated temperatures is also relatively high. The excellent wear resistance and elevated-temperature hardness combined ensure that nitrided steels can be used with good results for hot-work dies. However, the choice of steel and nitriding time must be adapted to suit working conditions.
HEAT TREATMENT—SPECIAL

10 mm diameter. By gas nitriding at 510°C for 20 h the fatigue strength increases to roughly 56 kpm²/mm². Nitriding for 60 h gives a fatigue strength of around 62 kpm²/mm² at 10⁷ load cycles.

Tests involving gas nitriding and salt-bath nitriding of smooth test bars made from steel DIN C 35 and 34 Cr 4 show that as regards fatigue strength in bending gas nitriding is somewhat more advantageous for the first, and salt-bath nitriding is rather better for the second steel (see Figure 6.170)²⁸. However, the differences are so slight that it does not really matter which method of nitriding is employed.

The notch effect on fatigue strength is extremely marked. The effect of this unfavourable factor is lessened by the compressive stresses which are introduced into the steel during nitriding. Hence nitriding is a very useful process if the aim is to increase the fatigue strength of machine components which, because of their ultimate application, have to be provided with notches, e.g. bolts. Figure 6.171 shows that the fatigue strength drops by half as a result of the notch effect in hardened and tempered test bars made from DIN 14 CrMoV 69. By undertaking a Tufttride treatment for 90 min at 570°C the fatigue strength increases both for the unnotched and for the notched bars, the nitried notched bar reaching the same value as that achieved by the non-nitried, unnotched bars²⁹.

The depth of nitriding which gives the maximum fatigue strength is dependent on material thickness and shape. This is illustrated in Figure 6.171.

6.172 which shows the torsional fatigue strength as a function of nitriding depth after the gas nitriding of test bars 6:5 mm diameter. One test bar has a transverse hole 1 mm diameter, which reduces the fatigue strength. With this bar the maximum fatigue strength is achieved with a nitriding depth of 0.3 mm. In the case of the undrilled bar the maximum fatigue strength is obtained with a nitriding depth of 0.4 mm. With thin-walled parts, salt-bath nitriding is preferable in such cases as compared with gas nitriding because it is easier to control the depth of nitriding during bath nitriding when such shallow depths are required.

CORROSION RESISTANCE

As a result of nitriding the corrosion resistance of non-stainless steels is increased. From the corrosion viewpoint the white layer is equivalent to a 13% martensitic chromium steel. In stainless steels, corrosion resistance is reduced because the chromium is bound as nitrides. The temperature rise during nitriding can also occasion inter-crystalline corrosion in non-stabilized 18/8 steels. Corrosion resistance is greatest for nitried and polished surfaces, however by grinding off, say 0.03 mm, the white layer is removed and corrosion resistance is reduced. After grinding 0.03–0.06 mm steel BS 905M39 has better corrosion resistance than the conventional heat-treatable steels and BS 42037. When tested in salt-water spray, the two last mentioned groups exhibited roughly the same corrosion resistance in the nitried layer (see Figure 6.173). From various nitriding tests on 18/8 steels, contradictory
of 15 mm diameter × 50 mm. They were tested in the 'normal state'; after a Sulfinuz treatment for 1 h at 570°C; after gas nitriding for 30 h at 510°C; and after gas nitriding and then having approximately 0.05 mm on the diameter ground off. Figure 6.174 illustrates the appearance of the specimens after sixteen days in the tropical chamber. The matt surface on the specimens that were nitrided only is due to the film of oxide that forms during nitriding and that tends to reduce somewhat the corrosion resistance. The best results are obtained if, as mentioned above, polishing is undertaken subsequent to...
nitriding. After grinding, the specimen made from BS 905M39 has withstood the corrosion attack better than have the other grades of steel. The poor results obtained after Sulfinox treatment were due in part to the fact that the specimens were not polished prior to corrosion testing.

6.5 CARBONITRIDING

6.5.1 Definition

During carbonitriding, C and N are absorbed simultaneously in the steel, the N increasing the hardness of the carburized layer. This process can be carried out in a salt bath or gas. The treatment temperature is normally 800–900°C, but both lower or higher temperatures can be employed. Basically, carbonitriding in a salt bath is the same as cyanide-bath hardening. In the section on case hardening we have already discussed N pick-up during carburizing in the cyanide bath. However, unless anything to the contrary is stated, by carbonitriding is meant carburizing in gas with simultaneous N pick-up. Carburizing takes place in the same way as described previously. Nitrogen, in the form of gaseous ammonia, is supplied direct into the furnace or in the same feed pipe as is used by the carburizing gases. The ammonia content is usually kept between 3 and 8%. The normal value for the N content of the surface layer is about 0.4%. After treatment the steel is hardened in the same way as after conventional gas carburizing.

In some cases ammonia is supplied only during the last 30–60 min of the process. The object of this procedure is to avoid the soft skin which can occur as a result of internal oxidation in the surface layer. As a result of such internal oxidation, the alloying elements are combined with oxygen and thus do not contribute to hardenability.

The description 'carbonitriding' is somewhat misleading, because we are not dealing with nitriding of the lower temperature type described earlier. The main emphasis is on carburizing and more correctly the process should be called nitro-carburizing. The name 'carbonitriding' is however so widely used now that it probably cannot be changed. Since 'correct' carbonitriding, i.e. nitriding with a supply of carbonaceous gas to the ammonia gas, has begun to be used on an increasing scale (see p. 383), we should perhaps speak of carbonitriding above or below A1.

6.5.2 Theoretical background

At the treatment temperatures normally applied the solubility of N in α-iron is approximately 0.1%. Its rate of diffusion is roughly the same as that of C, as shown in Table 6.21. The diffusion of N in α-iron proceeds ~ 50 times more quickly than it does in γ-iron. The corresponding figure for C is 40. However, the equilibrium diagram for iron-nitrogen resembles the iron-carbon diagram (see Figure 6.119) and at the temperatures used for carbonitriding, nitriding can be taken in principle as equivalent to carburizing.

<table>
<thead>
<tr>
<th>Temperature °C</th>
<th>Diffusion coefficients cm² s⁻¹ in α-iron</th>
</tr>
</thead>
<tbody>
<tr>
<td>600</td>
<td>5.3 × 10⁻⁸</td>
</tr>
<tr>
<td>700</td>
<td>3.6 × 10⁻⁸</td>
</tr>
<tr>
<td>800</td>
<td>2.3 × 10⁻⁸</td>
</tr>
</tbody>
</table>

Chatterjee-Fischer and Schaaber studied the relationship between N content on the one hand and treatment temperature, the amount of ammonia supplied and the C content on the other. Figure 6.175 shows the effect of treatment temperature and the amount of ammonia supplied. The N content increases with increasing quantity of ammonia and decreases with increasing temperature. Figure 6.176a and b show that the C content has very little influence on the N content with ammonia addition up to 5%.

The investigations were made on iron foils 0.05 mm in thickness. Parallel tests were also made using cylindrical test specimens made from plain-carbon and alloy steels, from which thin surface layers were machined off to enable the N content to be determined. In practically all cases the same N contents were found in the removed 0.1 mm outer layer, as were found in the foils. Figure 6.177 shows a typical distribution of C and N in the surface layer of a carbonitrided milled carbon steel.

Préosini made detailed studies of the effect of N on the hardenability of steel and established that contents of around 0.2% N have a favourable influence. This is illustrated in Figure 6.178 which is a TTT diagram for a
430 **HEAT TREATMENT—SPECIAL**

Steel with 0.8% C and 0.22% N. For comparison purposes the transformation curves have been drawn for the same steel without N. Apart from a direct increase in hardenability, N results in a lowering of both $A_s$ and $M_s$. Contents higher than 0.20% increase hardenability to a moderate extent, but lower $A_s$ and $M_s$ considerably. The values in Table 6.22 were obtained on the same steel as above, with higher N contents. Since N lowers $M_s$ substantially the amount of retained austenite in alloy steels can become high when carbo-nitriding.

![Graph showing isothermal transformation of austenite with 0.80% C and 0.22% N after carbo-nitriding at 660°C (after Prénovot)].

![Graph showing C and N content](image)

**Figure 6.176(a) and (b).** Nitrogen content as a function of carbon content in foils which were carbo-nitrided with different ammonia additions at 850 and 930°C (after Chattey-Buecher and Schade).
Tests have also been made using Jominy test bars which were carburized or carbonitrided using different N activities. The test bars were then austenitized at 820 °C for 30 min and quenched in the conventional manner. The hardness was measured after the test bars had been ground down to depths between 0.1 and 0.5 mm below the surface. The contents of C and N in these layers were analysed. The results obtained are shown in Figure 6.179. With N contents below 0.20%, there is hardly any tendency towards increased hardenability, but this tendency increases noticeably with higher N contents. The increase in hardenability can be studied in the diagrams at Jominy distances around 10 mm.

6.5.3 Conclusions

1. Carbonitriding gives increased hardenability and wear resistance compared with ordinary case hardening.
2. Carbonitriding gives a hard and uniform layer or case to such plain-carbon steels as tend to have soft spots when quenched in water.
3. Plain-carbon and low-alloy steels which normally have to be quenched in water may, when carbonitrided, sometimes be quenched in oil.

6.6 INDUCTION HARDENING

6.6.1 Fundamental principles

When electric current passes through a conductor, an electro-magnetic field is created round it. If the conductor consists of a coil, a magnetic field

---

**Table 6.22**

<table>
<thead>
<tr>
<th>N %</th>
<th>A1, °C</th>
<th>M1, °C</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.0</td>
<td>720</td>
<td>205</td>
</tr>
<tr>
<td>0.19</td>
<td>682</td>
<td>154</td>
</tr>
<tr>
<td>0.65</td>
<td>670</td>
<td>108</td>
</tr>
</tbody>
</table>

---

*Figure 6.179: Jominy diagrams. After either carburizing or carbonitriding the Jominy test specimen is austenitized at 820°C for 30 min. These material is 0.80%C, 0.97%N, 0.40%Mn. The test in the diagrams gives the carbon and nitrogen contents at different depths below the surface (after Pratelloni): (a) carburized Jominy test specimen; (b) carburized Jominy test specimen (0.08-0.37% N); (c) carbonitrided Jominy test specimen (0.08-0.16% N).*

*Figure 6.180: Path of magnetic flux through a metal bar inserted in a coil through which an electric current is flowing.*
flows through the coil. This field persists even if a metal bar is inserted into
the coil, as shown in Figure 6.180. If the magnetic flux is created by a high-
frequency alternating current, it gives rise to eddy currents in the surface of
the metal bar, which consequently becomes heated (see Figure 6.181). In
iron, hysteresis losses also contribute to some extent to the temperature rise
up to the Curie point (768 °C), above which iron is non-magnetic.

![Figure 6.181](image)

The depth of penetration of the heat is governed mainly by the power
and frequency employed. The normal power input is 0.1–2 kW/cm² of the
heated surface. The relationship between depth of penetration and frequency
can be calculated approximately by using the following simplified expressions,
which are valid for the temperature rise in steel up to the hardening tem-
perature:

\[
d_l = \frac{20}{\sqrt{f}} \quad \text{cold state (20 °C)}
\]

\[
d_s = \frac{500}{\sqrt{f}} \quad \text{hot state (800 °C)}
\]

where \(d\) = depth of penetration in mm

\(f\) = frequency in cycles per s (Hz)

Owing to heat conduction in the material during heating, the overall depth
of penetration is larger. It is possible to calculate the additional penetration
due to heat conduction from the following expression.

\[d_h = 0.2 \sqrt{t}\]

where \(d_h\) = depth of penetration in mm

\(t\) = time in s

The total depth of penetration is obviously \(d + d_h\). It should be stressed that
these expressions give only a rough estimate of the depth of penetration and
they have been included here only to show the fundamental effects of fre-
quency and time.

<table>
<thead>
<tr>
<th>Type of steel</th>
<th>C %</th>
<th>Si %</th>
<th>S % max</th>
<th>P % max</th>
<th>Ni %</th>
<th>Cr %</th>
</tr>
</thead>
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<td>0.31–0.39</td>
<td>0.5–0.80</td>
<td>0.03</td>
<td>0.03</td>
<td>0.30</td>
<td>0.70</td>
</tr>
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<td>2</td>
<td>0.38–0.44</td>
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<td>0.03</td>
<td>0.03</td>
<td>0.30</td>
<td>0.70</td>
</tr>
<tr>
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<td>0.5–0.80</td>
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<td>0.03</td>
<td>0.30</td>
<td>0.70</td>
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<tr>
<td>4</td>
<td>0.48–0.55</td>
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<td>0.03</td>
<td>0.30</td>
<td>0.70</td>
</tr>
<tr>
<td>5</td>
<td>0.55–0.60</td>
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<td>0.03</td>
<td>0.30</td>
<td>0.70</td>
</tr>
<tr>
<td>6</td>
<td>0.25–0.30</td>
<td>0.5–0.80</td>
<td>0.03</td>
<td>0.03</td>
<td>0.30</td>
<td>0.70</td>
</tr>
<tr>
<td>7</td>
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<td>0.5–0.80</td>
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<td>0.03</td>
<td>0.30</td>
<td>0.70</td>
</tr>
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<td>8</td>
<td>0.25–0.30</td>
<td>0.5–0.80</td>
<td>0.03</td>
<td>0.03</td>
<td>0.30</td>
<td>0.70</td>
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<td>9</td>
<td>0.25–0.30</td>
<td>0.5–0.80</td>
<td>0.03</td>
<td>0.03</td>
<td>0.30</td>
<td>0.70</td>
</tr>
<tr>
<td>10</td>
<td>0.25–0.30</td>
<td>0.5–0.80</td>
<td>0.03</td>
<td>0.03</td>
<td>0.30</td>
<td>0.70</td>
</tr>
<tr>
<td>11</td>
<td>0.25–0.30</td>
<td>0.5–0.80</td>
<td>0.03</td>
<td>0.03</td>
<td>0.30</td>
<td>0.70</td>
</tr>
</tbody>
</table>
6.6.2 Steel grades for induction hardening

By induction hardening we wish to achieve a considerable increase in surface hardness, and steels are used having C contents of 0.30–0.50%, which give hardness values of 50–60 HRC. The upper limit is governed by the C content that may risk hardening cracks. From practical tests it has been found that this C content is around 0.50%. If the heat treatment is well controlled, higher C contents are permissible, e.g. for rolls, that are made from steels having 0.80% C, 1.8% Cr and optionally 0.25% Mo. In accordance with ISO 683/XII–1972, the steels in Table 6.23 are recommended for flame and induction hardening. Steel 3, with increased Mn content (ca 1.30%) has also proved suitable for this purpose. Steel 9 corresponds to BS 1084/42 (En 19 C). Cr–Mo steels with lower carbon contents are also widely used (see Table 6.6). The steels are normally quenched in water. In certain cases the alloy steels can be cooled by means of an oil emulsion. Steels 9–11 can also be oil quenched.

The steel, prior to induction hardening, should be in the hardened and tempered or normalized state, and the hardness should be such as is acceptable for the unhardened sections. Fully annealed steel is not so suitable, because normally the time needed for dissolution of the carbides is longer than the normal heating-up time. Consequently induction hardening of fully annealed steel can result in inadequate and irregular values unless the heating time is matched to suit the fully annealed state.

6.6.3 Equipment for induction hardening

As the frequency governs the depth of penetration (depth of hardening), equipment should be selected mainly on the basis of the required depth of hardening. Since the latter generally increases with the size of the hardened part, the consequence is that, for large parts, equipment operating at low frequencies and low power input per unit of surface is selected whereas for small parts the best results are obtained using equipment operating at high frequencies and with high power input per unit of surface (Table 6.24).

<table>
<thead>
<tr>
<th>Table 6.24 A GUIDE TO FREQUENCY SELECTION</th>
</tr>
</thead>
<tbody>
<tr>
<td>Frequency range</td>
</tr>
<tr>
<td>Intermediate frequency</td>
</tr>
<tr>
<td>High frequency</td>
</tr>
</tbody>
</table>

As an example it can be stated that for the hardening of large rolls for cold rolling, Bofors use three rotary converters operating at 300 Hz with a total power output of 900 kW and a power input per unit surface of 0.1 kW/cm².

436 HEAT TREATMENT—SPECIAL

by means of which a hardness depth of up to 20 mm can be obtained. For the hardening of small components, valve generators operating at frequencies around 500,000 Hz and power outputs up to 100 kW are employed.

6.6.4 Working coils and fixtures

The coils, also known as inductors, also become heated by the electric current and by thermal radiation from the heated steel. The coils are often made from copper tubing and are cooled during operation by internal water flow. Figures 6.182a–c show three basic types of working coils designed for external and internal heating, and for heating of a flat surface.

The internal diameter of the copper tube is matched to suit generator

Figure 6.182(a)–(c). Types of coils used for induction heating

Figure 6.183. Coil for spin-hardening of rolls
power output. For the power output normally required, around 50 kW, tubing with an internal diameter of around 5 mm is used. The coils are usually wound with a distance of 2–5 mm between turns and with roughly the same spacing between the coil and the workpiece. By varying this distance it is possible to influence the rate of heating to a very large extent. Figure 6.183 illustrates a coil that is fixed in a hard plastic mounting. The individual turns of the coil in Figure 6.184 have been positioned using glass fibre spacers.

![Coil for hardening of tapered shafts](image)

Figure 6.184. Coil for hardening of tapered shafts

The first coil is designed for heating rolls for cold rolling, and the second for dealing with shafts having tapered sections.

Parts that are to be hardened must be held in a fixture of some sort so that they are correctly located in the inductor, and small components must be firmly held so that they are not displaced by the magnetic field. A metallic fixture that comes near the coil may become heated by the induced current. To avoid this the fixture and centre washers are normally made from an insulating and heat-resistant material such as 'sindango' or a heat-resistant plastic.

### 6.6.5 Procedure during induction hardening

According to the grade employed, the steel, after it has been heated in the coil, is cooled with water or oil or, in certain cases in air. Heating and cooling can be carried out in accordance with two main methods, these being single-shot hardening and progressive hardening.

During single-shot hardening the component is positioned and heated, after which it can be quenched in various ways as shown in Figure 6.185.

(a) The component can either be immersed by hand in the quenching liquid or it can be released from the fixture so that it drops into the liquid.

(b) In hardening machines the heated steel parts can be fed automatically into a quenching spray when heating has terminated.

(c) The coil can also be provided with a quenching spray which starts to operate when the high-frequency current is switched off.

Figure 6.186 shows a gear that has been heated to the hardening temperature in the inductor located above and which is just going to be transferred into the quenching spray.

The parts may also rotate during heating and, if possible, during cooling.
This method is employed for parts that have rotational symmetry. This gives a more uniform depth of hardness penetration, because it compensates for any irregularities in the coil.

**Progressive Hardening**

Progressive hardening is used if the power output is not adequate to permit single-shot hardening, or if a particularly shallow depth of hardness is required. The method is employed for very long parts, and normally requires a cross-sectional area that is uniform along the entire length of the hardened surface. During hardening, the component should be held in position, e.g. in a rotating chuck or between a pair of jaws. It is fed forward at a certain rate through the coil, so that it is heated and quenched progressively during its passage through the coil and spray. A separate spray as shown in Figure 6.187a can be employed; alternatively, the working coil can be combined with the spray into one unit, so that the spray ring itself also functions as the inductor, Figure 6.187b.

![Figure 6.187: Progressive hardening of shafts using (a) separate spray and (b) integrated working and spray coil](image)

![Figure 6.188: Inductor with spray quench ring brazed to it](image)

Figure 6.188 illustrates a unit where the spray coil is brazed to the underlying inductor. During hardening, the spray coil is located underneath the inductor. Figure 6.189 shows the progressive hardening of a roll. Progressive hardening can also be carried out on parts that are not axially symmetrical, e.g. as shown in Figure 6.190. Here cooling is carried out by means of air.

**6.6.6 The influence of various factors on hardness and depth of hardening**

As mentioned previously, the C content has a decisive influence on hardness after hardening. As the C content increases up to about 0.80%, the hardness also increases, particularly when in combination with other alloying elements. Both the hardening temperature and holding time are decisive factors that govern the hardness and depth of hardening. For induction hardening with conditions otherwise remaining the same, the depth of hardening is governed by the frequency of the inductive current. Since the hardening
temperature, heating time and holding time can be controlled relatively easily during induction hardening, it is possible with this method to determine in advance the required hardness depth with a relatively high degree of certainty by using calculated and empirical values.

The hardening temperatures employed for induction hardening are normally some 50°C higher than those used for conventional hardening. The holding times are extremely short. In many cases there is no holding time, i.e. cooling starts directly when the desired temperature has been reached. Temperatures are normally controlled by means of optical pyrometers.

Figures 6.191–194 show examples of hardness and depth of hardening of steels having a base composition of 0.7% Mn, 1.1% Cr, 0.25% Mo and varying C contents. Induction hardening was carried out at 10 000 Hz. In this case the power output employed was the same throughout. Hence the factor governing the hardening temperature was the heating-up time. The test specimens, 50 mm in diameter, were made from hardened and tempered steel with a hardness of ca 300 H.B. Quenching took place in water directly when the specimens had reached the desired temperature.

Figures 6.195 and 6.196 show the results from tests using a frequency of 400 000 Hz. In this case the power output was regulated so that the heating-up times to the desired hardening temperature were different. Here too the specimens had a diameter of 50 mm and they were quenched in water directly as the current was cut off. Tests at temperatures of 875 and 900°C gave almost
identical results. Use of the high frequency resulted in the depth of hardening being just about half the value obtained at 10000 Hz; compare Figure 6.196 with Figure 6.192.

All the experiments described above were made on steels that had been normalized or alternatively hardened and tempered. It was not possible to observe whether the mode of pretreatment had made any difference. When hardening rolls, relatively deep hardening is required. This can be achieved by employing a low frequency, e.g. 500 Hz, and by pre-heating the rolls.

Figure 6.197 shows the temperature as a function of the time during the heating-up and cooling of a pre-heated roll with a diameter of 300 mm. Figure 6.198 shows the hardness distribution and the microstructure in the surface layer.

After being induction hardened, the steel is usually tempered at 150–200 °C, or higher if a lower hardness is desired. Parts that are to be ground should be tempered at at least 160 °C. If the parts are tempered at lower
Figure 6.197. Temperature as a function of time during heating and cooling of a roll for cold rolling. 200 mm diameter, preheated to 300°C (after Zetterlund and Björnander).

Figure 6.199. SIS 2244, AISI 4140, BS7984A2, En 19 B

Figure 6.200. SIS 2244, AISI 4140, BS7984A2, En 19 C

Figures 6.199-200. Curves showing hardening for SIS 2244 (Bofors RO 752) and SIS 2244 (Bofors RO 912) after induction heating to 900°C, heating time 25 s, and subsequent tempering at different temperatures. Frequency 10 kHz.
temperatures there is considerable risk of cracking during grinding. Figures 6.199 and 6.200 illustrate how tempering influences hardness.

Tempering can also be performed by induction heating, this being particularly suitable for localized tempering. Figure 6.201 illustrates a coil design employed for the tempering of flanges on car axles. During tempering, two axles are placed in a fixture with the flanges facing each other. The axles rotate during heating.

<table>
<thead>
<tr>
<th>Type of steel</th>
<th>Hardness HRC</th>
</tr>
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<tbody>
<tr>
<td>1</td>
<td>50–55</td>
</tr>
<tr>
<td>2</td>
<td>53–57</td>
</tr>
<tr>
<td>3</td>
<td>56–61</td>
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<tr>
<td>4</td>
<td>57–62</td>
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<td>58–63</td>
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<td>52–57</td>
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<td>56–61</td>
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<td>9</td>
<td>55–60</td>
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<tr>
<td>10</td>
<td>55–60</td>
</tr>
<tr>
<td>11</td>
<td>55–60</td>
</tr>
</tbody>
</table>

After tempering at 180 °C the hardness values given in Table 6.25 may be expected for the steels included in ISO 683/XII–1972.

### 6.6.7 Examples of induction-hardened machine components

Machine components which are induction-hardened are mainly bodies having rotational symmetry, such as gear wheels and shafts. Gears can be hardened in accordance with any of the methods described in Figure 6.185.

but the gear should preferably rotate. The power input and the temperature employed should be so matched that the central portions of the gear are not heated to temperatures beyond 200 °C, which can be checked from the tempering colours. This is particularly important in the case of gears with finish-machined internal splines, which may be deformed during heating. Rotating the part during induction hardening, also called spin hardening, is generally suitable for gears with modules up to 5.

Acceptable contour hardening of a gear by means of spin hardening can be obtained if a high energy concentration, ca 2 kW/cm², and high frequency are employed. There is, however, a risk of overheating the tips of the gears and inadequate hardening at the tooth roots. When using intermediate frequencies the teeth first become hot at the roots, after which the heat diffuses towards the tips. In this case the teeth become through-hardened. More uniform contour hardening is achieved if heating is first started at an intermediate frequency until the tooth roots show a slight red heat, after which the
tooth tips are heated by means of a high-frequency current for a short period.
When hardening large gears, machines are employed which harden one tooth at a time by progressive heating.
In the following sections summarise methods of induction hardening for gears. Figures 6.20a–d show the hardness zones for the differing methods.

(a) SINGLE-SHOT HARDENING OF TOOTH TIPS

The hardening is quite simple. The wear resistance of the teeth increases, but the strength remains unchanged. This method can be employed with high-frequency hardening up to module 3, and with intermediate-frequency hardening up to module 5.

(b) SINGLE-SHOT HARDENING OF COMPLETE TEETH

Method (b) requires somewhat more experience than method (a). When using method (b) not only the wear resistance but also the bending strength of the teeth is increased. This method is recommended for modules up to 5. Figure 6.188 shows an example of this method of hardening.

(c) FLANK HARDENING (PROGRESSIVE HARDENING)

This method is used for modules ≥ 2 using high-frequency hardening and for modules ≥ 5 using intermediate frequency hardening. In principle the properties obtained are the same as by method (a). The hardening process is relatively simple.

(d) TOOTH GAP HARDENING (PROGRESSIVE HARDENING)

The sphere of application is the same as for method (b). A considerable increase in bending and fatigue strength is achieved. This 'contour hardening' is actually the ideal method for gears, but it requires considerable experience. Trial hardening heat treatments and metallographical investigations are necessary before good results can be expected.

Induction hardening is very advantageous for long shafts on which only the bearing sections need hardening. Thanks to induction hardening we avoid deformation and distortion that could occur if the entire shaft were hardened. Figure 6.203 shows a grinding-machine spindle that has been induction hardened on two bearing surfaces and on the flange. Shafts fitted with splines made from hardened and tempered steels often crack at the splines owing to overloading and fatigue. This susceptibility to cracking can easily be overcome by induction hardening. The depth of hardening below the base of the spline should at least be equal to the height of the spline ribs. In the

Figure 6.203: Grinding-machine spindle which has been induction-hardened at the cylindrical and tapered bearing surfaces and at the flange at the middle of the spindle. Steel BS 708.4.42 (Before RO 952)

Figure 6.204: Machining component which has been induction-hardened at the splines. Steel BS 708.4.42 (Before RO 952)

case of the part shown in Figure 6.204 the optimal strength was obtained when the depth of hardening was 7 mm.

6.6.8 Advantages and disadvantages of induction hardening

From the preceding section it is easy to understand that induction hardening can offer considerable advantages when compared with conventional methods of hardening. The major advantages are:
1. Restricted localized hardening
2. Short heating-up periods.
4. Only slight deformation.
5. Increased fatigue strength.
6. Any straightening required can be carried out on the unhardened surfaces and—to a certain extent—on the hardened surfaces.
7. The process can be incorporated in a production line.
8. Low operating costs.

The following are among the disadvantages:

1. The high capital costs necessitate high degree of equipment utilization.
2. The method is restricted to components having a shape suitable for induction hardening.
3. Only a limited number of grades of steel can be induction hardened.

Induction heating can be used for heat-treatment processes other than those mentioned above, e.g. for continuous annealing or normalizing of steel tubing or for brazing and welding.

6.7 FLAME HARDENING

Flame hardening and induction hardening are two methods which, in many respects, give equivalent results. The advantages of induction hardening enumerated in the preceding section also apply, by and large, to flame hardening. The capital investment costs for flame hardening are lower than for induction hardening, but the operating costs are higher. Flame hardening in its simplest form consists of heating the steel to the hardening temperature by a flame from a welding torch and then quenching it in water or oil.

The grades of steel which were recommended for induction hardening are also suitable for flame hardening, which, moreover, can be used for the partial hardening of tools from alloy tool steels. If the steels exhibit sufficiently high hardenability, the requisite hardness can be achieved by cooling in air. In certain cases flame hardening is preferred to induction hardening, because it is easier to direct the heat to selected surface areas.

6.7.1 Methods of hardening

The gas used for heating is a mixture of oxygen and a combustible gas, usually acetylene, natural gas or propane. It is mixed in a burner unit which has a shape to suit the part to be hardened. The appearance of the part also governs the method of hardening employed. We can distinguish between the following three main types.

1 MANUAL HARDENING

The designation indicates that the steel part is heated by hand with a welding torch or some other burner to the hardening temperature and is then cooled

in water, oil or air. This method is suitable if a small surface, e.g. the tip of a screw, is to be hardened. Larger surfaces can also be hardened by hand, but this requires larger burners. By making sweeping movements with the burner, the operator ensures that the temperature is uniform over the entire surface to be heated.

When hardening a large number of parts it is customary to keep the burner stationary and to hold the parts by hand or in a fixture. Hardening can also be automated so that the parts are held in a fixture which travels past the burner, where they are heated for some seconds and then dropped into the quenching liquid.

2 SPIN HARDENING

This method is employed for bodies having rotational symmetry, the bodies being placed on a rotating table or in a chuck. The speed of rotation is relatively low, normally about one revolution per second. The number of torches used depends on the size and shape of the part. The method is also applicable to parts that are not absolutely symmetrical. Since the heating

operation may take several minutes, the temperature tends to be equalized by heat conduction into the steel. Gears with modules up to 5 can be spin hardened. After they have been heated to hardening temperature the parts are either cooled by spray or by immersion into a quench tank. Figure 6.205 illustrates the spin hardening of a trolley wheel.

3 PROGRESSIVE HARDENING

This can be carried out on a flat surface or in combination with spin hardening. In the first instance a burner combined with a cooling spray is passed over the surface which is to be heated (see Figure 6.206). The rate of movement
is relatively low, 50–200 mm/min, and naturally this has to be adapted to suit the size of burner and the required depth of hardening. The tip of the inner cone of the flame is kept only a few mm away from the workpiece.

If progressive hardening is combined with spin hardening (see Figure 6.207), the speed of rotation and feed are of considerable importance to the uniformity of the hardened layer. As a result of the rotary movement the cooling water is thrown off tangentially. At higher rotational speeds this can interfere with the heating-up process and can also become a nuisance owing to water splashing into the workshop. At low rotation speeds, with a high feed and only a few torches, it is possible for ‘spirals’ to appear around the workpiece which contain alternate soft and hard bands (see Figure 6.208). In such a case ring burners should be used.

### 6.7.2 Hardness and depth of hardening

As is the case with conventional hardening, the hardness is governed by the composition of the steel (mainly the C content), by the hardening temperature and by the rate of cooling. The hardness values given for induction hardening apply also to flame hardening. The depth of hardening, also, is affected by the factors mentioned above. For a given steel cooled at a certain rate, the heating-up time and intensity of heating are the decisive factors. It is very difficult to illustrate the effect of these factors by means of a table or diagram. General knowledge of heat treatment as well as good equipment and sound judgement are required if the desired results are to be achieved.

Irrespective of the method of hardening, the hardening temperature is reached more rapidly with acetylene–oxygen than with propane–oxygen. Hence, with acetylene–oxygen gas a much more sharply defined depth of hardening can be achieved.

With flame hardening there is a risk of oxidation of the steel surface, particularly if the steel should be overheated. At temperatures of around 1100 °C, which can be reached quite easily, and under oxidizing conditions, the steel can become so oxidized that copper—which is always present in steel and which does not oxidize to the same extent as iron—becomes concentrated and melts. If the steel surface is subjected to tensile stresses copper diffuses into the steel along the grain boundaries and causes surface cracking.

The Swedish Association Metal Working Industries (Sveriges Mekanikbund) has undertaken a research project on flame hardening to study the extent of hardening and depth of hardening during progressive flame hardening treatments. For the flame-hardening trials the modified flame-cutting machine shown in Figure 6.210 was used, which incorporated the burner units illustrated in Figure 6.206. The temperature distribution in the steel surface was determined on test bars size 25 × 50 × 100 mm, the 25 × 100 mm surface being subjected to the hardening trials. The temperature distribution with water cooling and air cooling respectively is shown in Figures 6.210 and 6.211. By burner distance is meant the distance between the burner orifice and the steel surface being hardened.

Figure 6.212 shows the hardness distribution in test bars made from carbon
steel C 50 after trial hardening when using a feed rate of 50 mm/min and different burner distances. It is also possible for tool steels of type O 1 and D 2 to be flame hardened without cracking. With these steels it is sufficient to cool them in air in order to obtain surface hardness values of around 800 H V. Figure 6.213 shows

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**Figure 6.210.** Temperature distribution at different distances below the surface during flame hardening followed by water quenching.

**Figure 6.211.** Temperature distribution at different distances below the surface during flame hardening followed by cooling in air.

**Figure 6.212.** Relationship between hardness profile and burner distance during flame hardening of a carbon steel with 0.5% C. Test specimen dimensions: 25 x 75 x 150 mm, feed 50 mm/min. Water spray quenching. Temperature measured 10 mm below the surface.
results from a trial with steel O 1. During a hardening run, with the minimum burner distance, the steel became coarse grained in the surface layer.

Steel D 2 is unaffected by overheating as regards grain size, but it is liable to contain retained austenite and be prone to melting. During the tests it was clear that this effect was due, among other things, to the rather low heat conductivity of the steel, which resulted in a steep temperature gradient. Figure 6.214 also shows that the temperatures measured 10 mm below the surface are considerably higher than in the preceding case. When hardening the edges only of the test bars, and with only a slight change in the feed rate, both normal and overheated structures were obtained, as shown in Figure 6.215.

6.7.3 Examples of flame-hardened machine components and tools

Gears can be flame-hardened by several methods. Up to module 5, spin hardening is used. If only the tooth tips are to be hardened, gears of larger modules can be spin hardened. With gears of module 4 or more, one tooth at a time can be progressively hardened. This can be undertaken in such a way that either the flanks on one tooth are hardened simultaneously (flank hardening) or the flanks on two teeth are hardened simultaneously (tooth-gap hardening). From modules 6 and up it is also possible to harden the space between the teeth (tooth bases), so that the fatigue strength is increased. The principles involved resemble very closely those described for the induction heating of the example shown in Figure 6.202.

Progressive hardening can be carried out with a quenching spray directly after heating, this being essential for gears made from carbon steel. In the
case of alloy steel gears similar to the one shown in Figure 6.216, each tooth is heated progressively, after which the gear wheel is rotated 15-60° and the tooth allowed to cool in air. The other parts of the gear are kept cool by means of cooling water.

A similar method is employed for spur gears and sprocket wheels which are placed vertically, with roughly one quarter of the wheel immersed in water. When the tooth located at the highest position has been heated to the hardening temperature, the wheel is rotated through 180°, so that the heated tooth is quenched. For single-tooth hardening pre-indexed machines are used which give symmetrical stresses in the gear wheel, and thereby only slight deformation is created.

**LONG OBJECTS**

Long objects, such as lathe guides, are hardened progressively, whereby a flame-cutting machine can be used for the feed. Figure 6.217 shows the way in which hardening progresses. The depth of hardening is regulated by the amount of heat supplied (gas pressure) and the feed rate. Figure 6.218 illustrates lathe guides that have been hardened by using the same gas-pressure setting but different feed rates.

One special case is illustrated in Figure 6.219, where a cam shaft for a printing press is hardened simultaneously along four sections, length about 100 mm each. The larger part of the cam shaft is immersed in water, which provides sufficient cooling to ensure that satisfactory hardening is achieved.
A similar procedure can be used when hardening long edges, for example. The object is placed in water with the edge just level with the surface. As the flame sweeps towards the edge, the water is blown away and the edge is exposed and heated. As the burner travels onwards, the water flows back and quenches the edge.

During the flame hardening of long and relatively thin objects it is difficult to avoid deformation. When one side of a straight edge is being hardened it might be presumed that this will become convex, since the martensite formed has a greater volume than the other structural constituents. This does in fact occur quite frequently and is advantageous when straightening is called for. The straight edge may also become concave, which is due to the fact that the upsetting occurring during heating exceeds the increase in volume that has taken place during the martensite formation. The degree of contraction can be regulated by means of the feed rate.

**PLATES**

Plates can be successfully flame hardened with the aid of wide burners. Large plates can be hardened in several passes, but a soft zone between the passes must be allowed for. Figure 6.220 illustrates a cross-section of a flame-hardened plate made from Cr-Ni-Mo steel.

**SIMPLE BLANKING TOOLS**

Simple blanking tools (punching tools), as mentioned previously, can be given a wear-resistant cutting edge by means of flame hardening. This method is used mainly for tools employed in blanking large plates. Such a tool can either be built up of shear plates which are bolted firmly to a parent tool, or it can be manufactured from plate in which the desired contour has been cut to shape by flame cutting and then machined. In the former instance the shear plates can be dismantled and flame hardened separately if so required. In the latter case hardening must be carried out using a torch which is guided around the entire cutting edge. When the torch has returned to the point which was hardened first, a narrow, softer zone forms, the location of which should be at a point where minimum wear is anticipated.

Carbon steel containing 0.50%, C, as well as steel W 1, can be used for simple tools. Quenching is carried out by water spraying. It may also suffice to keep the tools immersed in water, with the cutting edge just above the surface of the water.

**MOULDS FOR PLASTICS**

Moulds for plastics are largely made from steel P 20 hardened and tempered. If the plastic is very abrasive or if the pressure at the mating surfaces is extremely high, flame hardening of these mating surfaces can impart a considerable improvement in wear resistance. After flame hardening and cooling in air the hardness of P 20 (Bofors 1968) is 50–53 HRC.

The following recommendations are given for flame hardening the edges of moulds made from steel P 20 when using a conventional welding torch or burner which is passed over the edges of the mould as shown in Figure 6.221.
1. The area to be hardened should be clean and free from scale.
2. An oxy-acetylene flame is preferable, initially with slightly more acetylene than oxygen—a 'blue or green cone' will be noticed in the flame.
3. Oxygen should be increased until the 'blue or green cone' disappears.
4. A slight pre-heat by movement of the flame over the beginning of the area to be hardened should be given.
5. The tip of the cone flame should then be held about 8 mm (1/2 in) back from the edge of the material, approximately 8 mm (1/2 in) above the material and at an angle of 60° to the horizontal plane and 15°-30° to the vertical.
6. Once a dull-red hot spot becomes visible, the flame must be moved slowly along, so that the red spot follows the flame at a constant speed and remains a constant size. Speed is approximately 50 mm (2 in) per minute.
7. The flame should not be allowed to dwell or the process to discontinue and then be restarted. If the flame dwells, sparks may appear, which indicates burning of the metal. If the process is discontinued it is virtually impossible to start where the flame left off and the localized uneven heating gives rise to stressed or soft areas.
8. Tempering at 180–200 °C is recommended, but not necessary, after flame hardening. Local heating with the flame is used.

A suitable nozzle for flame hardening is AGA Welding nozzle No. 4. Alternatively BOC nozzle No. 18. Safe 3 range.

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7

Dimensional changes during hardening and tempering

One of the most difficult problems facing the heat treater is the dimensional changes that occur in the steel during hardening and tempering. By dimensional changes we mean changes in both shape and size.

7.1 DIMENSIONAL CHANGES DURING HARDENING

One of the main causes of dimensional changes is the stresses which occur as a consequence of the contraction of the material during cooling, i.e. thermal stresses. The other main cause is the transformation stresses which occur as a result of the martensite formation.

7.1.1 Thermal stresses

When a body cools, the outer layer cools more quickly and contracts. The inner, softer parts try, during this process, to assume a spherical shape, this being the shape to which they offer the least resistance during deformation. Hence the main rule is that all bodies with shape deviating from the spherical one attempt to assume this shape during rapid cooling.

The effect of thermal stresses can best be studied in a low-carbon steel, in which we can disregard the likelihood of martensite formation. Austenitic steel is also a suitable object for study. Frether and Lowitzer performed a series of investigations which show the effect of cooling rate on slabs size 200 x 200 x 20 mm made from mild steel. Slab a in Figure 7.1 is solid whereas slab b has an inner square hole size 100 x 100 mm. To illustrate the dimensional changes more clearly, these have been drawn to a larger scale. Figure 7.1 shows clearly that the more drastic the quench, the greater are the changes. Figure 7.2 shows that the greater the temperature drop during cooling, the greater is the deformation.
The high-temperature strength of the steel also has some effect. As shown by Figure 7.3, the steel with the greatest high-temperature strength (18/8 steel) exhibits the highest dimensional stability.

### 7.1.2 Transformation stresses

During heating and cooling steels pass through a series of structural transformations. The various structural constituents possess different densities and hence differing values for specific volume, as shown by Table 7.1.

The amount of carbon dissolved in the austenite or martensite has a relatively high effect on the specific volume. When calculating the changes in volume which take place during the transformations in conjunction with hardening, due regard must be paid to the carbon content, as is also shown by Table 7.2.

The changes in length can be studied by using a dilatometer, in which
a bar of steel is placed between two quartz bars, the changes in length being measured during heating and cooling. If cooling takes place at such a high rate that martensite is formed, we obtain a diagram, as shown in Figure 7.4, where the continuous curve is particularly significant. During heating a continuous increase in length occurs up to \( A_1 \), where the steel shrinks as it transforms to austenite. After the austenite formation is completed, the length increases again. However, the coefficient of longitudinal expansion is not the same for austenite as for ferrite.

On cooling, thermal contraction takes place and during martensite formation the length of the steel increases. After cooling to room temperature all martensitic steels contain some retained austenite, the amount increasing with the amount of alloying elements dissolved during austenitization. The larger the quantity of retained austenite contained in the steel after hardening the smaller is the increase in length or increase in volume. If the retained austenite content is sufficiently high, we generally obtain a reduction in volume, see curve \( T_3 \) in Figure 7.4. However, the volume of the austenite increases with the quantity of dissolved carbon, and this has to be taken into account when calculating the changes in volume.

Taking as a basis the proportions of martensite and austenite, together with the amount of carbon dissolved therein, it is possible with the aid of the data in Table 7.2 to calculate the changes in volume occurring during hardening. If the steel contains undissolved cementite this volume has to be deducted during the calculations. Consequently the equation for use in the theoretical calculations takes the following form:

\[
\frac{\Delta V}{V} = \left(100 - V_a - V_m\right) \left(1 - 0.68 \times C\right) + \frac{V_a}{100} - \frac{V_m}{100} - 4.64 + 2.21 \times C
\]

where \( \Delta V/V \) = change in volume in %

\( V_a = \frac{\%}{\%} \) by volume undissolved cementite

\( V_a = \frac{\%}{\%} \) by volume austenite

\( 100 - V_a - V_m = \frac{\%}{\%} \) by volume martensite

\( C = \frac{\%}{\%} \) by weight of carbon dissolved in austenite and martensite respectively.

In the case of a carbon steel containing 1% C it should be possible in theory, to avoid changes in volume during hardening if, for example, the steel contains 10% undissolved cementite and 13% retained austenite. In this case the carbon content in the martensite is ca 0.38%. However, with such a low carbon content we do not achieve a retained austenite content as high as 13% (see Figure 1.22). If all the cementite were to be dissolved, we would require a retained austenite content of ca 40% to avoid any changes in volume occurring in this steel. This situation too, is impossible with this carbon steel. Hence we must always expect an increase in volume when hardening a 1% carbon steel, assuming that full hardening is achieved. Since such a steel becomes fully hardened only in dimensions less than approximately 10 mm, the increase in volume in the larger dimensions is quite moderate, because in such cases the hardened layer comprises only a few per cent of the total volume.

Assuming that the contents of martensite and retained austenite remain the same as in the carbon steel, we might expect steels with higher harden-
abilities to show larger volume increases. However, since the amount of retained austenite is greater in the higher-alloy steels this compensates for the increase in volume which would result from the increased hardness.

In medium-high and high-alloy tool steels the quantity of retained austenite and hence the increase in volume can be controlled by means of the hardening temperature, as shown schematically in Figure 7.4. In this case the hardening temperature \( T_2 \) does not result in any increase in volume after cooling to room temperature. If there is a volume diminution because a high hardening temperature \( (T_f) \) is used, the original volume can be restored by means of subzero cooling.

In a material with the same properties in all directions (isotropic material) the relative changes in length equals a third of that of volume. Engineering steels are, however, anisotropic which means that the linear change occurring during hardening will not be the same in the direction of rolling as in the directions at right angles to it.

In, for example, high alloy content chromium steels of type D-2 and D-6 the carbides are elongated in the direction of rolling in the manner shown in Figure 7.5. During hardening these steels expand more in the direction of rolling than in the other directions. According to Frehse\textsuperscript{1} this is because during the heating-up period to the austenitizing temperature the carbides do not expand so much as the matrix. During cooling the carbides impede the thermal contraction of the matrix in the longitudinal direction. No strict scientific proof has been put forward in support of this theory, but the model is simple and easy to grasp and provides a good means of remembering the phenomenon. Other grades of steel also exhibit marked anisotropy.

### 7.2 Dimensional Changes during Tempering

Before this section is read, it is recommended that Section 1.4 be read through once more. During tempering structural transformations occur which change the volume of the steel and its state of stress. There is a certain correlation between the tempering temperature, volume and state of stress. From the educational viewpoint it might be advisable here to distinguish between changes in volume and changes in stress state.

#### 7.2.1 Changes in Volume

During tempering the martensite decomposes to form ferrite and cementite which implies that there is a continuous decrease in volume. As a result of tempering at high temperatures, the volume reverts by and large to its original value prior to hardening, provided that we can disregard plastic deformation (see Figure 7.6). Since we cannot count on an initial value of 100% martensite prior to tempering, the continuous curve in Figure 7.6 is not representative of the changes in volume occurring in the grades of steel which can be employed for engineering purposes.

The dashed curves plotted in Figure 7.6, which represent increases in volume during different tempering stages, are only schematic, as is the entire diagram which is representative of many different types of steel.

The retained austenite which in plain-carbon and low-alloy steels is transformed to bainite in the second stage of tempering at about 300°C results in an increase in volume. Because of the relationship between bainite and retained austenite, taken in conjunction with the tempering temperature employed, either an increase or a decrease in volume may take place within this temperature range.

When tempering high-alloy tool steels in the temperature range 500-
600 °C, very finely distributed carbides are precipitated. This gives rise to a stress condition which results in increased hardness and greater volume. Simultaneously with the precipitation of carbides the alloy content of the matrix is reduced which implies that the $M_i$ of the retained austenite will be raised. During cooling down from the tempering temperature the retained austenite will consequently be transformed which also results in an increase in volume.

### 7.2.2 Changes in stress conditions

The stress condition which prevails in the steel after hardening is, as mentioned previously, governed by the thermal stresses and transformation stresses which have occurred during hardening. The continuous decomposition of martensite during tempering causes at the same time a continuous reduction in the state of stress which also facilitates the transformation from austenite to martensite. With the aid of the split-ring test, Brown and Cohen studied the stress changes occurring in the hardened ball-bearing steel of the type AISI 52100. After hardening the rings were stressed to 42.2 kp/mm² (415 N/mm²) and 19.7 kp/mm² (194 N/mm²), and then tempered for 1 h at different temperatures.

After relaxation the rings were measured and the changes in stress conditions were calculated. Even after tempering at 260 °C (500 °F) 85% stress relaxation was measured, see Figure 7.7.

With the aid of X-ray diffraction, stress measurements were made by Bofors on an induction-hardened cold roll after tempering at different temperatures. The results of the measurements are shown in Figure 7.8. The very high initial stresses were due to the extra stress additions which are created during surface hardening, since the annular hardened zone cannot expand freely during martensite formation. After tempering at 260 °C the stress reduction was about 60%. This figure may be considerably less in asymmetrical bodies that have been drastically quenched. In such cases appreciable stress relieving is not obtained before a tempering temperature of around 600 °C is attained (see Figure 5.10).

### 7.3 Examples of dimensional changes during the hardening and tempering of tool steels

Theoretical calculations which are based on the specific volumes of the different structural phases can never provide values which are of practical application, because these presuppose a knowledge of the amount of each individual constituent present in the steel after hardening. Furthermore the calculations do not allow for the dimensional changes which occur as a result of thermal stresses, or allow for the anisotropy of the steel. Hence we nearly always have to rely on empirical values. During the last 15-year period comprehensive experiments have been made by Bofors aimed at determining
the dimensional changes which occur during the hardening and tempering of the most commonly used types of tool and constructional steels. With the theoretical background as outlined above, and using the practical experience obtained both during the tests described above and from close co-operation with heat-treatment shops it has proved possible to assess the dimensional changes occurring even for bodies whose shape and size differ from those employed during the experiments. Each grade of steel was found to have its own characteristics which are not applicable to other grades. Nevertheless, it is possible to assign the steels to groups in the manner shown below. The dimensional changes as a percentage, mentioned in the graphs, represent the average values of several tests and we must expect a scatter of up to \( \pm 0.5\% \). In the following, by longitudinal direction we mean the direction of rolling.

### 7.3.1 Plain-carbon steels

The low hardenability of these steels makes the total increase in volume highly dependent on the material thickness of the tool. The steels obtain a high martensite content, and sections completely hardened through, i.e. material thicknesses not exceeding 10 mm, exhibit the largest possible increase in volume on hardening. In dies of considerable material thickness on the other hand the proportion of martensite by volume is so small that it is possible almost to disregard the increase in volume.

**Example 1.** Figure 7.9 shows the results obtained from experiments using embossing rolls made of steel W 1 which could not be ground on the engraved outer diameter after heat treatment. In the roll having the largest material thickness the increase in diameter remained relatively small after hardening. After tempering at 180 °C the dimensions shrank below the original values. In the roll made of thinner material the depth of hardness was greater, so that there was a greater increase in volume. Since the roll hardened through at the ends, the largest increase in diameter occurred there.

### 7.3.2 Low-alloy steels

The majority of the low-alloy tool steels are employed in dimensions such that they are completely or almost completely fully hardened. In the through-hardening section sizes the dimensional changes occurring are, viewed as a percentage, relatively uniform, even in the case of different material thicknesses. In the event of different degrees of partial through-hardening, however, the dimensional changes occurring are irregular. In those sizes where the steels are surface-hardening, the volume increases least.

**Example 2.** Steel O 1 is a grade normally used for dies for blanking tools, the thicknesses employed for this purpose ranging usually up to 30 mm. **Figure 7.10** shows how through-hardened plates of this grade change in dimensions on quenching in oil or in a martempering bath. Quenching in oil results in an increase of roughly 0.1%, in dimensions in all directions. During martempering the longitudinal dimension increases to a greater extent, mainly at the expense of thickness. Undoubtedly the reason for this is the less drastic thermal shock occurring during martempering which does not result in such a considerable contraction in the longitudinal direction as does the more severe shock imposed during oil quenching. This experiment was undertaken on steels from different heats and at different heat-treatment shops and unequivocal results were obtained. If the considerable risk of cracking involved in oil quenching can be disregarded, this method of quenching, as applied to plates described here as well as to similar plates, is to be preferred to martempering when dimensional stability in longitudinal direction is required. During this experiment the dimensional deviations occurring as regards the planeness of the sides were relatively slight and these have not been included in the diagrams.

**Example 3.** During one series of experiments plates made from steel O 1.1 were hardened from three different hardening temperatures, involving two different holding times. These variables had only negligible influence, as can also be deduced from **Figure 7.11** which includes the highest and lowest hardening temperatures. On tempering up to 200 °C a slight contraction occurs in all directions of the plate. At higher temperatures there is an increase in dimensions, with a maximum at 300 °C, after which the dimensions once again decrease. The increased volume at 300 °C is occasioned by the transformation of retained austenite. At 700 °C the dimensions have reverted almost completely to their original values (this point has not been included in the diagram).

**Example 4.** When hardening a die made from steel O 1 having dimen-
sions as shown in Figure 7.12, the manufacturer had relied on there being no change in the dimension of 37550 mm between the holes. Owing to the shape of the die, it was hardened by quenching in a martempering bath. After tempering for 2 h at 200 °C the length measurement was found to be 375.35 mm, this being an increase of 0.09%. Further heat treatments were performed with the following results:

Tempering 4 h 220 °C 375.75 mm
Tempering 3 h 400 °C 375.30 mm
Annealing 10 h 700 °C 374.90 mm

Tempering at 220 °C for 4 h corresponds approximately to tempering at 300 °C for 1 h, which gives the maximum increase in length. By carrying out a tempering at 400 °C, the length dimension was reduced to somewhat below the value obtained after tempering at 200 °C. After annealing, the die reverted to what was probably the original dimension.

The die was re-hardened by quenching in oil and tempered as follows:

Tempering 4 h 180 °C 374.90 mm
Tempering 6 h 180 °C 374.99 mm
Tempering 6 h 180 °C 375.01 mm

It may seem odd that no increase in length occurred after quenching in
oil. Regarded from the viewpoint of cooling intensity, the dimension 30 × 165 mm is equivalent to 50 mm diameter. In this dimension the steel is not fully through-hardening so that the increase in volume is less. Tempering for 4 h at 180 °C corresponds approximately to the lowest value on the length variation curve. After tempering for 6 h at 180 °C the dimension has increased. The long tempering time at 180 °C (4 + 6 h) is equivalent to tempering for 1 h at 205 °C, and at this temperature the transformation of the retained austenite starts, which results in increased volume.

Example 5. By subzero cooling, e.g. at −80 °C, the dimensions increase, provided that tempering has not been undertaken at excessively high temperatures. In one case a hardened and tempered bushing of steel O 1 was to be shrunk into a cast iron stand. Since it was not possible to heat the stand; the bushing was subzero cooled but as a result it became so large that it could not be shrunk into the stand. In such cases it is advisable to carry out a subzero treatment and then grind the workpiece before cooling for shrinkage is carried out.

Example 6. S1S 2092 (Bofors SR 1855) differs only slightly from S1S 2140 (O 1) as regards change in shape. Hardening experiments were undertaken with S1S 2092 on test bars as shown in Figure 7.10. Hardening temperatures of 850, 870 and 890 °C gave very similar results. To illustrate the difference which is encountered as regards warping during hardening in oil
DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

and martenpering. Figure 7.13 shows the largest and smallest dimensional deviations for two test bars.

Example 7. When hardening thin-walled rings made from S1S 2092 and O 1, we must expect an increase in both the outer and inner diameters. During the subsequent tempering, the rings follow in principle the same course of dimensional variations as illustrated previously. The rings in Figure 7.14 after quenching were tempered at 180 °C only, so that the inner diameter shrunk until it reached almost the original value.

Figure 7.14. Change in diameter of rings made from low-alloy tool steel after quenching in oil and tempering at 180 °C. (a) Ring dimensions: diameter 14/18 x 31 mm. (b) = dimensions after tempering, A = dimensions after tempering. (a). Steel S1S 2092 (German SR 1853). Hardening temperature 810 °C. Tempering temperature 180 °C. (b). Steel S1S 2092 (German SR 1853). Hardening temperature 860 °C. Tempering temperature 180 °C.

Example 8. As the wall thickness of rings made from these types of steel increases, the picture changes. When the inner diameter is reduced so that it is roughly half the outside diameter, the inner diameter will shrink during hardening (see Figure 7.15).

Figure 7.15. Change in diameters of rings made from S1S 2092 (German SR 1853) after quenching in oil from 870 °C.

Example 9. When the thickness of rings made from S1S 2092 and O 1 increases to cross-sectional areas which are equivalent to a bar ca 50 mm diameter, the depth of hardening is reduced and the increase in volume is negligible. As the wall thickness increases still further, the depth of hardening is reduced further and both inner and outer diameters shrink, which is because the thermal stresses dominate.

During measurements made on blanking die rings made from S1S 2092 with cross-sectional areas within the critical range, the following variations in outer and inner diameters were observed after quenching in oil and subsequently tempering at 200 °C Table 7.5.

Table 7.5

<table>
<thead>
<tr>
<th>OD</th>
<th>ID</th>
<th>H</th>
<th>Cross-section</th>
<th>ΔOD</th>
<th>ΔID</th>
</tr>
</thead>
<tbody>
<tr>
<td>283</td>
<td>193</td>
<td>65</td>
<td>65 x 45</td>
<td>-0.2</td>
<td>-0.2</td>
</tr>
<tr>
<td>376</td>
<td>270</td>
<td>65</td>
<td>65 x 53</td>
<td>-0.5</td>
<td>-0.8</td>
</tr>
<tr>
<td>439</td>
<td>281</td>
<td>65</td>
<td>65 x 79</td>
<td>-1.1</td>
<td>-1.4</td>
</tr>
</tbody>
</table>

OD = outer diameter mm
ID = inner diameter mm
H = height mm

Example 10. With bar diameters exceeding about 80 mm, or equivalent rectangular cross-sections (see Figure 4.39), the depth of hardening for steel S1S 2092 is often inadequate. In such cases it is possible, during hardening, first to quench the steel in water for several minutes and then in oil.

For such a test, rings having dimensions 495/335 diameter x 175 mm were used. The cross-sectional area 80 x 175 mm is equivalent to a diameter of 115 mm. Quenching took place first in strongly agitated oil only, and then in water for 1 min after which the quenching was finished in oil. The following diameter changes were observed (Table 7.4).

Table 7.4

<table>
<thead>
<tr>
<th>Method of hardening</th>
<th>ΔOD</th>
<th>ΔID</th>
<th>ΔH</th>
<th>Depth of hardening mm</th>
</tr>
</thead>
<tbody>
<tr>
<td>Quenching in oil</td>
<td>-0.46 ± 0.14</td>
<td>-0.81 ± 0.06</td>
<td>0.09 ± 0.05</td>
<td>5</td>
</tr>
<tr>
<td>Quenching in water/oil</td>
<td>-1.12 ± 0.42</td>
<td>-1.86 ± 0.22</td>
<td>±0.00 ± 0.05</td>
<td>10</td>
</tr>
</tbody>
</table>

In spite of the fact that the depth of hardening was greater for the rings which were quenched in water/oil, the reduction in diameter was greater, this being due to the fact that the thermal stresses exceed the transformation stresses.

Example 11. As mentioned previously, and as shown in Figure 7.13, the steel often reverts to its original dimensions after a full anneal. If the heat treatment which has been performed is considered incorrect, as regards dimensional variations, there is consequently the possibility, that by per-
FORMING A FULL ANNEAL, THE ORIGINAL DIMENSIONS WILL BE OBTAINED. DETAILS WILL BE GIVEN HERE OF SUCH A CASE WHICH CONCERNED A DIE MADE FROM SIS 2092.

The length of the die, Figure 7.16, was the factor governing the serviceability of the tool. According to information received the length should have been 123.00 mm. After hardening this had increased to 123.38 mm, which is 0.3% Since this increase in length appeared to be unreasonable, the plate was thoroughly annealed, whereupon it shrank to 123.17 mm, which is 0.17% and which can be regarded as a more reasonable value. The plate was subsequently rehardened in exactly the same manner as employed on the first occasion, whereupon the length increased by 0.17%. By tempering it at 200°C the length was reduced somewhat. It is highly probable that the original length dimension was greater than that stated. Details are given of the results of the measurements and of the complete heat treatment process in Table 7.5.

![Figure 7.16. Blanking die made from SIS 2092 (Bofors SR 1855) showing dimensional changes after different heat treatments.](image)

<table>
<thead>
<tr>
<th>Heat treatment</th>
<th>A</th>
<th>B</th>
<th>C</th>
<th>D</th>
<th>E</th>
<th>F</th>
<th>1</th>
<th>2</th>
<th>3</th>
<th>4</th>
<th>5</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>123.38</td>
<td>123.38</td>
<td>123.39</td>
<td>94.85</td>
<td>94.81</td>
<td>94.81</td>
<td>9.98</td>
<td>9.98</td>
<td>9.97</td>
<td>9.98</td>
<td>9.97</td>
</tr>
<tr>
<td>2</td>
<td>123.17</td>
<td>123.18</td>
<td>123.16</td>
<td>94.70</td>
<td>94.66</td>
<td>94.67</td>
<td>9.97</td>
<td>9.97</td>
<td>9.95</td>
<td>9.97</td>
<td>9.95</td>
</tr>
<tr>
<td>3</td>
<td>123.38</td>
<td>123.36</td>
<td>123.37</td>
<td>94.88</td>
<td>94.81</td>
<td>94.84</td>
<td>9.98</td>
<td>9.98</td>
<td>9.97</td>
<td>9.97</td>
<td>9.96</td>
</tr>
<tr>
<td>4</td>
<td>123.31</td>
<td>123.31</td>
<td>123.30</td>
<td>94.85</td>
<td>94.76</td>
<td>94.80</td>
<td>9.97</td>
<td>9.97</td>
<td>9.96</td>
<td>9.96</td>
<td>9.96</td>
</tr>
</tbody>
</table>

The conventional ball-bearing steel AISI 52100 has approximately the same characteristics as the two steels mentioned above. Consequently it has been possible in several cases to apply experience gained with these steels to the ball-bearing steel. Example 12. A hardness of 50–55 HRC had been requested for a ring-shaped die diameter 300/210 × 30 mm made from AISI 52100. After hardening in oil the die was tempered at 300°C. Its hardness was found to be 55 HRC. Both the external and internal dimensions had increased by approximately 0.35 mm and the die was unserviceable, because the inner diameter was the ruling dimension. It was then tempered at 400°C, so that the hardness became 52 HRC. The inner diameter was reduced by 0.28 mm, which was then sufficient to permit the die to be employed.

SIS 2580 (Bofors HRO 1243) varies dimensionally in the same way as steels O 1 and SIS 2092, which means that the dimensions increase roughly by 0.1% in all directions on quenching in oil. When quenching in a martempering bath the length and width increase somewhat more at the expense of...
DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

The thickness which decreases. On tempering, the largest dimension is encountered after heating for 1 h at 250°C, i.e. in the temperature range where the retained austenite transforms to bainite.

Example 13. Figure 7.17 shows the values measured on plates size 25 x 70 x 100 mm after different heat treatment processes. The lowest diagram in the figure which shows the dimensional changes occurring after cooling in air indicates that the percentage increase in thickness is somewhat greater than for the other dimensions. Since the absolute dimensional changes as regards thickness are very slight it is probable that the dimensional increase was caused by the thin film of oxide which formed during cooling in air.

As S18 2550 exhibits very good hardenability we can expect that the dimensional changes outlined above should be representative at least for material thicknesses up to roughly 200 mm. The percentage changes in dimension should then decrease as the dimensions increase.

Example 14. In Figure 7.18 a photograph of the inner part of a mould for pressing plastic transportation boxes weighing 9 kg is shown. The dimensions are 300 x 400 x 800 mm. The part is made in two halves. Hardening was performed from 840°C in compressed air to 240°C and then in still air after which the die was tempered at 400°C twice to give a hardness of 360–390 HR. The length which in this particular case was the vital dimension increased by 0.4 mm, which is 0.05%. Compare this with the lower portion of Figure 7.17 where the increase in length after hardening in air and tempering at 400°C is 0.04%.

Figure 7.19. Tool made from steel of the following composition: 0.55% C, 1.5% Cr, 3% Ni (Before HR 119)

Tempering took place at 180°C. The following dimensional changes were measured:

- Length +0.30 mm = 0.05%
- Width +0.65 mm = 0.11%
- Thickness +0.10 mm = 0.04%

There is a surprisingly good agreement with the values obtained after corresponding heat treatment, as shown by Figure 7.17. However, it is not
DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

It is advisable to calculate and design to be close, dimensionally, because numerous factors can affect dimensional changes.

Steel A2 occupies an intermediate position between the low-alloy and the high-alloy chromium steels. The hardening temperature used is moderate but is sufficiently high to give rise to changes in shape as governed by the cooling rate. During the hardening of this steel, quenching in oil can give rise to much greater deformation than quenching in a martempering bath or cooling in air. This change in shape can, for example, result in a plate becoming thicker in the centre than at the sides, so that there is a change in parallelism between the sides. Hence, this steel should not be quenched in oil. For the sake of completeness we should, however, include the following description of values obtained from hardening tests made using all the above-mentioned cooling media.

Example 16. Figures 7.20, 7.21 and 7.22 show the average values for the dimensional changes occurring in plates, size 25 x 70 x 100 mm after hardening in oil, martempering bath and air respectively. On quenching in oil there is a reduction in the length and width but the thickness increases. A sketch has been included in Figure 7.20 which shows the relatively large deviations involved as regards parallelism between the sides. On cooling in the martempering bath and in air, the length and width increase, but the thickness decreases. As regards the various cooling media, this steel consequently behaves in fundamentally the same way as the grades mentioned previously. The dimensional changes occurring after martempering are considerably less for this steel, than for the other tool steels.

The effect of the hardening temperature is quite obvious. The volume of the specimen increases with increasing hardening temperature, this being due to the increased amount of retained austenite.

During tempering, steel A2 passes through all the transformation stages which were sketched in Figure 7.6. Together with the continuous shrinkage.
which is occasioned by the decomposition of the martensite, there is first 
at roughly 300 °C an increase in volume due to the transformation of the 
retained austenite to bainite. After tempering at 450 °C or higher carbidie 
precipitation takes place and during cooling the retained austenite is trans-
fomed to martensite. The increase in volume occurring during tempering 

![Diagram](image1)

![Diagram](image2)

Figure 7.22. Dimensional changes after hardening and tempering of plates made from steel A 2 
(Before R G P 21). Specimen dimensions: 25 × 70 × 100 mm. Holding time at austenitizing tem-
perature 25 min. Air cooling from 925, 930 and 975 °C. Holding time during tempering 1 h. 
Longitudinal direction = direction of rolling

is greatest for those specimens which are hardened from the highest 
temperature. Note also the TTT diagrams in Figure 1.15 which illustrate 
the transformation of the retained austenite during the tempering of this steel. 
In the lower temperature range the transformation takes place isothermally 
to bainite, and in the upper temperature range martensite is formed when the 
steel cools from the tempering temperature.

![Diagram](image3)

Figure 7.23(a)–(b). Dimensional changes of plates made from steel D 6 (Before R T 66) after 
quenching in oil and austempering respectively. Specimen dimensions: 25 × 70 × 100 mm. Holding 
time at hardening temperature 25 min.
planeness was extremely good. As in the case of the steels mentioned previously, the length increased to a greater extent after martenising than after hardening in oil. Similarly, the volume decreased with increasing hardening temperature.

During tempering all dimensions decreased with increasing temperature (see Figure 7.24). At about 300°C the retained austenite was transformed to bainite which tended to impede the reduction in dimensions, which is at its maximum at 400°C. At higher temperatures the volume increased because of the formation of new martensite and because of carbide precipitation. After tempering at temperatures above 600°C the steel contracted.

When assessing the dimensional variations of ring-shaped dies made from D6, as regards the change in diameter we can assume as an average value the mean of the width and thickness variations.

Example 18. A blanking die made from D6 having dimensions and appearance as shown in Figure 7.25 shrank approximately 1 mm on the specified dimensions during hardening. The die was unserviceable because the outer dimensions were the decisive factor. As the hardness was only

Figure 7.24. Dimensional changes during tempering after hardening of steel D6 (Böhler R.T.69), Specimen dimensions: 15 x 75 x 100 mm. Holding time at tempering temperature 1 h. Longitudinal direction = direction of rolling.
DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING 493

ca 55 HRC, a high retained austenite content was suspected due to an excessively high hardening temperature. After annealing the die measurements indicated the dimensions as shown in the diagram. Hardening then took place from 950°C whereupon the outer diameter increased by approximately 0.1 mm. Tempering was undertaken at the lowest practicable temperature (160°C), after which the die was found to be exactly to size and could be used in spite of the fact that no grinding tolerance had been included. In this particular case it would undoubtedly have been possible to achieve an increase in size had the tool been tempered at 550°C or if it had been subzero cooled after the first hardening operation. Subzero treatment is recommended only in exceptional cases, because of the risk of cracking.

Example 19. Twelve feeder rollers of three different sizes made from D 6 were to be hardened to a minimum of 58 HRC. After hardening and tempering there should still be an allowance left for grinding in the hole, i.e. the internal diameter should not increase. Hardening took place from 980°C in a martempering bath at a temperature of 220°C. The initial tempering was done at 170°C for 1 h, after which the dimensions of the rollers were measured. Tempering was then carried out at 490°C for 1 h followed by measurement. Figure 7.26 shows a schematic diagram of the rollers and Figure 7.27 shows the dimensions and dimensional changes occurring after hardening and tempering. With only one exception, all dimensions reduced after hardening and tempering at 170°C. Nearly all dimensions were increased as a result of tempering at 490°C. The final result was that there was a small increase in the inner diameter (0.02 mm) of only one of the twelve rollers.

In this case the tempering temperature of 490°C must be regarded as the maximum value, because a higher temperature would have resulted in an increase in the inner diameter, see Figure 7.24. This investigation also gives a good idea of the scatter range to be expected when hardening several tools of the same size.

Figure 7.25. Blanking die ring made from D 6 (Bofors R T 60) which contracted during the first hardening operation by ca 3 mm along the dimensions shown in the sketch.

Figure 7.26. Sketch of feed rolls made from steel D 6 (Bofors R T 60).

Figure 7.27(a)–(c). Dimensional changes after hardening and tempering of feed rolls made from steel D 6 (Bofors R T 60). See p. 496 for (c).
Example 20. Tools having an elongated shape made from D 2 which were to operate at a somewhat elevated temperature were hardened from 1030 °C and then tempered at 560 °C. The length increased by ca 0.1%, which is 1.5 mm for the length concerned, i.e. 1500 mm. The reason for this relatively considerable increase in length was that the tools had been hardened from the highest temperature within the specified range for steel D 2. Since this excess length has to be ground off, the tool was thenceforth made 1.5 mm shorter in length which—with the heat treatment remaining the same—resulted in a considerable reduction in grinding costs.

7.3.4 Hot-work steels

Steels BH 13 and BH 10A (Bofors Q R O 45) differ considerably from other steels as regards dimensional changes because during hardening they always contract in the longitudinal direction (direction of rolling). Normally the width varies least, but the thickness can increase or decrease depending upon the method of quenching and material dimensions. When quenching flat workpieces in oil the deformation can be relatively large, and this is reflected in the thickening of the plates in the centre. Round tools, e.g. dies, change only negligibly in diameter. On cooling in oil it is possible for relatively severe swelling to occur on the plane surfaces.
Example 21. The hardening of straight-edge dies made from H13 size 25 x 50 x 150 mm was undertaken by quenching in oil, martempering bath and air respectively. There was a relatively large amount of deformation after quenching in oil as shown in Figure 7.28 but there was only a negligible change on quenching in the martempering bath (see Figure 7.28b). Figure 7.29 shows the dimensional changes occurring during hardening and tempering, along with the scatter band.

Example 22. Large dies made from H13 for the forging of steel are usually hardened and tempered to about 400 HB before machining and the change in dimensions of the rough-machined blocks is of minor importance. Hardening in oil is not recommended for finish-machined workpieces where stringent requirements are imposed regarding dimensional stability. Apart from the unavoidable shrinkage in the longitudinal direction, some deformation of the sides also occurs during hardening in air. As shown in Figure 7.30 those can become slightly concave, but convex deformation is more normally encountered.

Example 23. During hardening of the die-casting die shown in Figure 7.30, dimensional changes after hardening and tempering of a die block made from steel H13 (Before R.O.P. 19). Air-hardening from 1050 °C. Tempering at 400 °C twice for 2 h. Hardness 415 H B. Size 150 x 250 x 550 mm.
DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING 499

5.36 measurements were made before and after heat treatment. The hardening temperature was 1030°C and quenching was carried out in a martempering bath at 300°C. The die was tempered at 590°C for 4 h then at 580°C for 4 h. The changes in dimensions are shown in Figure 7.31. The length decreased by 0.08% which corresponds well with Figure 7.29c. At the lower end the width increased by ca 0.48%, which also agrees with Figure 7.29c. On the other hand, at the upper end of the die the width decreased. However, this may have been due to a permanent set caused by the weight of the die while it was hanging from its upper holes.

Example 24. The swelling which occurs with circular dies is illustrated in Figure 7.32 for specimens made from H 13, 75 mm diameter × 50 mm. Heating was carried out at 1050°C in a salt bath. Cooling in oil resulted in greater deformation than did cooling in a martempering bath. During tempering of the specimen the values shown in Figure 7.33 were obtained.

Example 25. Four rings, diameter 201/217 × 120 mm and two rings diameter 254/217 × 120 mm made from H 13 and with 0.4%, V (H 11 mod.) were quenched from 1025°C in a martempering bath at 400°C. Tempering was carried out at 560°C, twice for 2 h. The variation in all dimensions was within the scatter range of ±0.05 mm.

Example 26. Six rings made from H 11 mod. with dimensions as shown in Figure 7.34 were hardened in oil from 1025°C. They were tempered at about 560°C, twice for 2 h. Figure 7.34 shows the changes in dimension as a function of diameter. The majority of the dimensional changes are within the range −0.10 ± 0.03%, in the case of the outer diameter, and within −0.20 ± 0.03%, for the inner diameter. Increases as well as reductions in height were observed.

Example 27. If a martempering bath is not available and air cooling is inadvisable, the steel may be hardened by first cooling it in oil for some seconds and then letting it cool in air. This produces about the same dimensional stability as that achieved with martempering. When using the latter treatment, however, it is possible for the temperature of the bath to influence the change in dimensions as will be shown in the example.

In this case plates made from BH 10A (Bofors QRO 45) were used of size
502 DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING
50 × 75 × 100 mm, and were hardened from 1050°C. Regardless of the method
of cooling employed, the length decreased, whereas the width varied around
the 0-value. After hardening, the plates were in all cases thicker at the centre
than at the edges. In Table 7.6 the variations in thickness after tempering at

Table 7.6 EFFECT OF THE COOLANT ON CHANGE IN THICKNESS DURING HARDENING
OF BOVORS (R.O. 45)
Tempering temperature 550°C (t = 50 mm)

<table>
<thead>
<tr>
<th>Cooling medium</th>
<th>Variation limits</th>
<th>Range of variation</th>
</tr>
</thead>
<tbody>
<tr>
<td>Martempering bath 350°C</td>
<td>-0.01</td>
<td>0.01</td>
</tr>
<tr>
<td>Air</td>
<td>±0.00</td>
<td>±0.07</td>
</tr>
<tr>
<td>Martempering bath 250°C</td>
<td>±0.00</td>
<td>±0.14</td>
</tr>
<tr>
<td>Oil 30s, followed by air</td>
<td>±0.01</td>
<td>±0.14</td>
</tr>
<tr>
<td>Oil 60s, followed by air</td>
<td>±0.06</td>
<td>±0.12</td>
</tr>
<tr>
<td>Martempering bath 200°C</td>
<td>±0.01</td>
<td>±0.20</td>
</tr>
<tr>
<td>Oil</td>
<td>±0.04</td>
<td>±0.33</td>
</tr>
</tbody>
</table>

550°C provide a measure of the suitability of the coolant. Cooling in a
martempering bath at 350°C gives the smallest change in dimensions.

Figure 7.35 shows the values measured after tempering at different
temperatures for the specimen which was quenched in a martempering bath
at 350°C.

7.3.5 High-speed steels

As regards deformation during hardening and tempering, high-speed steels
exhibit some similarity to the high-alloy chromium steels. Because of the
high solution temperatures involved, high-speed steels have a relatively high
retained austenite content after hardening. This means that we must expect
shrinkage in all dimensions, except possibly in the length dimension, when
hardening from normal temperatures. When tempering at around 600°C,
all dimensions increase quite considerably.

Example 28. Test plates made from M2 size 25 × 70 × 100 mm were
pre-heated at 850°C and austenitized at 1210°C for 6 min, then quenched
in oil, in martempering bath at 400°C, and cooled in air, respectively. The
specimens were tempered for 1 h at ten different temperatures up to 650°C.
As a check, one series was tempered at 550°C only. This tempering operation
was carried out twice with intervening checks on dimensions. The average
values for the changes in dimensions are shown for all specimens in Figure
7.36. During hardening a reduction in volume occurred, due to the high
retained austenite content, and the thickness—taken as a percentage—
decreased the most, the width coming next. On quenching in oil there was
also a reduction in the length, whereas on martempering the length remained
almost unaltered. On cooling in air there was a very slight increase in length.
On tempering up to 500°C there was hardly any change in dimensions at
all, but at higher temperatures there was a sudden increase in all dimensions.
The specimens which were tempered only at 550 °C increased in size during the first tempering operation by roughly the same amount as the specimens which were tempered continuously up to 550 °C. During the second tempering of the first-mentioned specimens, the dimensions increased to a value which corresponds roughly to tempering at 570 °C for 1 h. Tempering for 2 h at 550 °C is equivalent to 1 h at 570 °C (see Figure 6.65). After hardening in oil the specimen had become visibly deformed, which was not the case after martempering or air hardening (see Figure 7.37).

Example 29. A series of tests, exactly similar to the one described in Example 28 has been carried out with steel M 42. Both steels behave in roughly the same way, but the variations between the dimensional changes in the different measured directions are considerably greater in the case of M 42 (see Figure 7.38). The increases in size after tempering in the range around 550 °C are also considerably greater for this steel than for M 2. Figure 7.39 provides sketches in which the scale shows the dimensional changes.

It is extremely important to be familiar with the large increases in size occurring in hardened high-speed steels after tempering at about 550 °C because the majority of high-speed steels are tempered precisely in this range. Concerning the changes in diameter of round dies, we can work on
7.4 DIMENSIONAL CHANGES DURING CASE HARDENING

During recent decades considerable interest has been devoted particularly by gear manufacturers to coping with the deformations which occur during case hardening. One of the reasons for this work is that it is desirable to avoid the expensive post-hardening grinding operations. The dimensional changes which occur during case hardening are governed by a wide variety of factors, and of these by way of introduction we can mention:

1. **The hardenability of the steel.** As the hardenability increases and the thickness of the material decreases, the volume increase becomes larger.

2. **Type of steel.** Cr–Ni steels, Cr–Ni–Mo steels and to some extent Cr–Mn steels behave in a similar manner. On the other hand Cr–Mo steels can exhibit some variations in behaviour as regards change in shape.

3. **Depth of case hardening.** This factor is extremely difficult to assess, because a considerable influence on the structure, properties and thickness of the case-hardened layer is exercised both by the carbon content in the surface layer and by the hardening temperature. However, it is clearly obvious that the depth of case hardening influences the dimensional changes.

4. **Method of hardening.** In principle direct hardening causes the least dimensional variations and double hardening causes the greatest changes. The deformation increases with the number of heating and cooling operations.

5. **Material dimensions.** A clear tendency towards shrinkage in the diameter of gears has been observed in connexion with small or moderate dimensions or material thicknesses. As the dimensions of gears and rings increase, the shrinkage decreases, and at a certain dimension there is an increase in the diameter.

**Example 30.** Rings with outer diameter 100 mm and inner diameter 50 and 75 mm, and height 50 mm were fabricated from steel BS 655M13 (3 Cr–Ni steel), BS 637M17 (En 352) and from BS 805M20 (En 362). The rings were gas carburized at 940°C to a depth of about 1 mm. Prior to quenching which was carried out in oil and martempering bath, respectively, the temperature was lowered to 830°C. After tempering at 170°C the dimensions of the rings were checked. It was noted that there had been a reduction in the outer and inner diameters and an increase in the height of all the rings. No significant difference was observed in the changes in dimensions between the various grades. Prior to the finish machining, the rings had been stress-relieved or normalized, respectively. Untreated rings were also included in the test. It was not possible to detect whether any influence was exerted by the different types of pre-treatment. Quenching in oil and in martempering bath also gave quite similar results.
DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

The dimensional changes which are shown for steel En 352 in Figure 7.40 and which comprise the average value for 18 rings of each dimension, are also representative for the other grades. Throughout, the dimensional changes were greater in the case of the thin-walled rings the volume of which had also increased more than that of the thick-walled rings. The increase in volume is shown in the bar diagram in Figure 7.41. It is probable that the smaller increase in volume on the part of the thick-walled rings was due to

![Figure 7.40. Changes in dimensions after case hardening of rings made from BS 637M17, En 352 (Bifor DR 34). Case-hardening depth 10 mm](image)

![Figure 7.41. Changes in volume after case hardening of rings made from different case-hardening steels](image)

508 DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

unsatisfactory through-hardening. The somewhat reduced hardenability of steel En 362 is also reflected in the results from the martempering.

Example 31. Rings made from En 352 with dimensions larger than those employed in Example 30, were case hardened in the same way as the latter, by martempering and tempered at 170°C and 300°C. The dimensional changes occurring are shown in Figure 7.42. Examples 30 and 31 show clearly that the reduction in diameter decreases as the thickness of the rings increases and that there is a critical dimension above which the diameter increases. This situation has also been confirmed by making measurements on gears having dimensions similar to those employed in Example 31 (see the Example 36). By tempering at temperatures exceeding 170°C the dimensions can be affected in the manner illustrated in Figure 7.42.

Example 32. The influence of the depth of case hardening was examined using rings made from En 352 size 100x50x50 mm. The rings were packed carburized at 900°C to a depth of 10 and 0.5 mm respectively. The rings were subsequently hardened in oil from 820°C and tempered at 170°C. One ring was hardened without being carburized. The dimensional changes are shown in Figure 7.43. As the depth of case hardening increases, a greater increase in volume is encountered during hardening, due to the formation of a larger quantity of martensite. Undoubtedly, the change in diameter is governed by the depth of case hardening.

Example 33. Bushings made from En 352 were to be produced with tolerances for which turning was sufficient. It was desirable that the bushings should be case hardened without the necessity of subsequent grinding to adjust the dimensions. A test batch of 10 bushings with dimensions as shown in Figure 7.44 was produced and check measurements were made. The bushings were gas carburized at 940°C to a depth of 1 mm. After the temperature had dropped to 830°C, they were martempered at 180°C and then tempered at 180°C.

The dimensional variations between the various points (A–C and D–F respectively) were equal (±0.01 mm) and the different bushings showed similar scatter. The values in Table 7.7 were obtained.

### Table 7.7: Changes in Dimensions Occurring during Case Hardening of Bushings

<table>
<thead>
<tr>
<th>Measurement</th>
<th>Prior to Hardening mm</th>
<th>After Hardening mm</th>
<th>Change(^{\circ})</th>
</tr>
</thead>
<tbody>
<tr>
<td>Outer diameter</td>
<td>55.29</td>
<td>55.52</td>
<td>−0.13</td>
</tr>
<tr>
<td>Inner diameter</td>
<td>44.08</td>
<td>43.93</td>
<td>−0.34</td>
</tr>
<tr>
<td>Length</td>
<td>103.55</td>
<td>103.49</td>
<td>−0.06</td>
</tr>
</tbody>
</table>

All dimensions decreased but the relatively large reduction in the inner diameters resulted in an increase of 0.43\(^{\circ}\) in volume. This increase in volume is more than twice the amount occurring with the thin-walled rings mentioned in Example 30 and this can be ascribed to the efficient through-hardening and to the comparatively large case depth of these thin-walled bushings.
To avoid shrinkage, hardening on mandrels is carried out, whereby the mandrel is made from a tool steel which is hardened and tempered at 400°C at least. When the quenching of gears or sleeves with inner splines is being carried out, the mandrels are made cylindrical so that the cooling medium
DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

511 can pass between the mandrel and the component which is to be hardened. For components with cylindrical holes it is advisable to provide the mandrel with grooves. To facilitate extraction of the mandrel, it can be Sulfirax treated prior to hardening.

Example 34. A connecting sleeve with dimensions as shown in Figure 7.45 was made from BS 655M13 and was pack-carburized to a depth of 0.7 mm. Hardening took place from 820 °C. Directly before the oil quenching, the sleeve was placed on a mandrel, the diameter of which lay just in the centre of the tolerance range (d = 106.55 mm). During a trial hardening of one sleeve, the inner diameter increased but only negligibly. Since the inner diameter of the sleeves was at the lower end of the tolerance zone, the remaining sleeves were hardened on a mandrel having a diameter of 106.67 mm. After hardening, 3 of the sleeves were wholly within the tolerance range, whereas 3 were on the plus side. A suitable mandrel diameter would have been 106.60 mm.

Example 35. Prior to the hardening of the splined sleeve as shown in Figure 7.46 it was found that the outer diameter was on the low side and in part below the tolerance range. Hardening was carried out in the same way as in Example 34. The diameter of the mandrel was the same as the inner diameter of the sleeve. The dimensions of the sleeves changed as shown in Figure 7.46. The amount of shrinkage was greater than desired but the sleeves...
DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

514

DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

shrink, but where shrinkage is prevented by the larger portion of the gear, the hole in which increases in size during hardening.

In each gear the diameter of the hole increased so much that the wheel could not be used. In order to rescue the unhardened gears, 8 throughgoing holes were drilled in the larger wheels and corresponding counterboring was made in the smallest gear, as shown in Figure 7.48. During subsequent

case hardening, we obtained the hole dimensions as shown in the diagram which meant that most of the gears could be used. This investigation, undertaken by Atlas Copco at Avosverken, shows clearly the influence of steel bulk on the dimensional changes occurring during case hardening.

7.5 DIMENSIONAL CHANGES DURING NITRIDING

Nitriding is the heat treatment which results in the least dimensional changes and gives the highest hardness values. Since the treatment is carried out at a

Example 36. During the hardening of gears, as shown in Figure 7.47, carburization took place in gas at 930 °C. After the temperature had dropped
to 830 °C, the gears were quenched in oil and tempered at 180 °C. The depth of case hardening was about 1 mm. The cross section of the steel in these gears was in all cases greater than that in the largest ring described in Example 31. There is very considerable similarity between the largest ring and the smallest gear. This also applies to the deformation of the hole. As the thickness of material in the gears increased, the hole diameter increased during hardening. At that part of the intermediate gear, where the hub is broader, the increase in diameter was negligible. In this case we can regard the hub as a separate ring which—considering its dimensions—should
DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

Temperature around only 500°C, no phase transformations take place in the base material. As a result of nitrogen pick-up in the surface layer an increase in volume does occur which occasions a swelling of the surface layer, besides setting up a stress condition between the surface layer and the core.

The amount of swelling ranges from 0.005 to 0.02 mm, depending on the length of the nitriding treatment. If the nitriding takes less than 60 h we can expect that the swelling will be 0.01-0.015 mm per side. With components having fairly heavy sections, e.g. more than 50 mm, the dimensional changes are restricted mainly to the swelling as mentioned above. When the section size, compared with the thickness of the nitried layer, is reduced there occurs a dimensional change, the magnitude of which is determined by the ratio of the cross-sectional areas of the nitried layer and the core.

The following approximate equation can be employed to calculate the increase in length:

\[ \Delta l = k \frac{N}{K} \]

where \( \Delta l \) = increase in length, %
\( k \) = constant, the magnitude of which depends on the section and the grade of material
\( N \) = cross-sectional area of the nitried layer
\( K \) = cross-sectional area of the core

Cross-sectional area of the nitried layer + cross-sectional area of the core = total cross-sectional area. This equation applies mainly to \( N/K \) values less than 0.7.

The following values (Table 7.8) for the constant \( k \) were obtained by experiment.

<table>
<thead>
<tr>
<th>Section</th>
<th>BS 905M39 (En 41B) (Before ARO 75)</th>
<th>BS 817M40 (En 24) and En 29B (Before CRO 061 and RO 7155)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Circular</td>
<td>0.3</td>
<td>0.15</td>
</tr>
<tr>
<td>Rectangular</td>
<td>0.4</td>
<td>0.2</td>
</tr>
</tbody>
</table>

To provide a guide for calculating the dimensional variations occurring during nitriding, some general examples will be given first, followed by details of several cases where both calculations and measurements were made.

Example 37. A straight-edge, size 10 x 20 x 1000 mm made from BS 905M39 (En 41B) is to be nitrided for 60 h. The depth of nitriding is estimated at 0.52 mm and the swelling of the surface layer at 0.01 mm.

Area of nitried layer \( N = 30 \text{ mm}^2 \)

Area of core \( K = 170 \text{ mm}^2 \)

516 DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

\( \Delta l = 0.4 \times \frac{30}{170} = 0.071\% \)

The change in length is \( \frac{0.071 \times 10}{100} + 2 \times 0.01 = 0.73 \text{ mm} \).

Example 38. A ring of En 29B size 150 x 134 x 10 mm is to be nitrided for 30 h. Depth of nitriding is 0.25 mm. Swelling of the surface layer is 0.005 mm. When calculating the areas we take as a basis the longitudinal cross-sectional area of the ring which is 10 x 8 mm:

Area of nitried layer \( N = 12 \text{ mm}^2 \)

Area of core \( K = 68 \text{ mm}^2 \)

\( \Delta d = \frac{0.2 \times 12}{68} = 0.035\% \)

The change in diameter is calculated on the basis of the mean diameter 142 mm.

\( \frac{0.035 \times 142}{100} = 0.05 \text{ mm} \)

The outer diameter increases by 0.05 + 2 x 0.005 = 0.06 mm.

The inner diameter increases by 0.05 - 2 x 0.005 = 0.04 mm.

Example 39. A number of rings size outer diameter = 368 mm, inner diameter = 348 mm and height = 181 mm which were to be nitrided could not be ground after nitriding. The rings were made from BS 817M40 (En 24). Prior to final machining the rings were stress relieved and then nitrided for 30 h. With the aid of the equation presented previously, the increase in the diameters was found to be as follows:

\( OD = 0.094 \text{ mm} \)

\( ID = 0.074 \text{ mm} \)

These values were taken as a basis for dimensioning the rings prior to nitriding.

The increase in diameter of 8 rings was measured after nitriding, the following average values being obtained:

\( OD = 0.087 \text{ mm} \)

\( ID = 0.067 \text{ mm} \)

In addition to this there was some ovality which when calculated for all rings amounted to \( \pm 0.03 \text{ mm} \). The rings could be used for the intended purpose without grinding.

Example 40. When a certain definite relationship exists between the outer and inner diameters, the inner diameter will shrink. To gain a better understanding of the effect of this, calculations were made for rings made from En 41 B with a width of 50 mm, outer diameter 70 mm and with
518 DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

varying sizes for the inner diameter. A depth of nitriding of 0.50 mm was
stipulated causing a surface swelling of 0.01 mm. The results are shown in
Figure 7.49. As indicated by the graph, the inner diameter shrinks in rings
having a wall thickness greater than 15 mm.

Example 41. Practical tests were made by Hubbard and Robinson
(Figure 7.50), using the same steels as in Example 38. The nitriding period
during these tests was 72 h and the nitriding temperature was 525°C which
gives both a greater depth of nitriding and more swelling than that shown
in Example 40. Bearing this in mind, the agreement between the two diagrams
is extremely close.

During the nitriding of unsymmetrical bodies it is possible for irregular
deformation to occur which may be due on the one hand to the fact that on
one side a body which is otherwise symmetrical has a larger surface area,
e.g. it incorporates a U-groove. The face having the largest surface area will
expand more during nitriding. This circumstance is well illustrated by the
following example:

Example 42. During the test the specimen shown in Figure 7.51 was
used. In all cases the surfaces of the slot were insulated so that no swelling
could take place there. The other plane surfaces were also insulated. Nitriding
was performed for 30 h and 60 h. The material comprising the test bodies
was BS 905M39 (En 41 B). Results are shown in Table 7.9.

**Table 7.9** DIMENSIONAL CHANGES DURING THE NITRIDING OF US NAVY

<table>
<thead>
<tr>
<th>Test Specimens</th>
</tr>
</thead>
<tbody>
<tr>
<td>Specimen number</td>
</tr>
<tr>
<td>1</td>
</tr>
<tr>
<td>2</td>
</tr>
<tr>
<td>3</td>
</tr>
<tr>
<td>4</td>
</tr>
</tbody>
</table>

![Figure 7.51. US Navy distortion-test specimen](image)

Example 43. The cylinders shown in Figure 7.52 were made from a
Cr-Mn case-hardening steel. Previously, during case hardening, the deforma-
tion produced was relatively large and grinding became expensive. Hence
519 tests were made employing gas nitriding for 30 h at 510°C. The surface hardness was ca. 625 HV and at a depth of 0.4 mm the hardness of the nitrided case was 400 HV. Measurements made at 20 points on 3 cylinders showed that all dimensions had increased during nitriding. The maximum increase, viz. 0.11 mm, was observed on the outer diameter (see the arrow in Figure 7.52). The inner diameter followed next with an increase of 0.08 mm.

Figure 7.52. Nitrided cylinder for compressed-air engine

The diameters in 90° angle increased in size by up to 0.04 mm. The largest measured diameter increases were due to a larger dimensional increase taking place in the intervening sections where the wall thickness is least on the one side and where there are milled recesses on the other. Both of these factors co-operate in the sense of making the linear increase the largest in this section, with the result that the diameter measurement in the 90° angle increases the most.

Example 44. When making sealing rings from BS 905M39 (En 41 B) it was found after check measurements that the rings, after lapping of the plane surfaces, were concave on one side and convex on the other. This change in shape had occurred because lapping was carried out down to different depths on the plane surfaces. This mistake was rectified by fine-grinding the rings prior to nitriding, after which only very slight lapping was required.

520 Dimensional Changes During Hardening and Tempering 7.6 Ageing

By ageing of tool steels we mean in the present context the change that a hardened and usually tempered steel undergoes at room temperature, or in certain cases at somewhat higher temperatures, changes that can result in altered dimensions. The principles underlying these changes have been discussed in Sections 1.4 and 7.2. The ageing which takes place within the temperature range concerned here involves decomposition of the martensite to ε-carbide whereby the volume is reduced, and also the transformation of retained austenite to bainite which is accompanied by an increase in volume. Some form of stress relief also takes place during ageing and this can result in warping. The grade of steel, and its heat treatment and the temperature at which ageing takes place, all play a part in governing which of the two effects mentioned above will predominate during ageing.

In carbon steel which has aged at room temperature the decomposition of martensite is the dominant factor if the steel has not been tempered prior to ageing. As the tempering temperature increases prior to ageing, this effect is reduced. As shown in Figure 7.53 after tempering at 200°C relatively high stability has been achieved.

![Figure 7.53. Steel W1. Change in length as a function of ageing time at 20°C after specimens were hardened and tempered at different temperatures (after Aerbuch)](image)

In certain cases conditions are more complicated in the alloyed steels. To gain some insight into the ageing process, this phenomenon will be discussed below for steel W 1 and O 1. The data have been derived from American sources. For the sake of comparison we have also included the values for steel L 3 having the composition 1.0% C, 1.5% Cr and 0.20% V.

After the test specimens had been hardened they were tempered, usually...
for 1 h at 150 °C, they were subzero cooled or subjected to a combination of both treatments. The test specimens were then kept at room temperature and were measured after certain intervals of time. The results obtained with steel W 1 are shown in Table 7.10.

**Table 7.10** W 1. Change in length occurring as a result of ageing at 20 °C

<table>
<thead>
<tr>
<th>Treatment temperatures °C</th>
<th>Hardness HRC</th>
<th>Change in length L x 10^-6 after ageing at 20 °C</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>1 week 1 month 3 months 1 year 3 years</td>
</tr>
<tr>
<td>150</td>
<td>65</td>
<td>+90    -175  -265  -405  -8</td>
</tr>
<tr>
<td>-196</td>
<td>65</td>
<td>-100   +40  +205  +310  +480</td>
</tr>
<tr>
<td>150, -196</td>
<td>65</td>
<td>-4     -8    +10   -14   -17</td>
</tr>
<tr>
<td>150, -196, 150</td>
<td>65</td>
<td>-5     -8    +10   -14   -17</td>
</tr>
</tbody>
</table>

Regardless of the treatment employed after hardening, a reduction in length occurs during ageing. After subzero-cooling alone the contraction is greater than the larger quantity of martensite which decomposes during ageing, coupled with the fact that the effect of any retained austenite transformation is almost completely absent. This effect is enhanced, the lower the subzero-cooling temperature employed. The smallest change observed in the tests took place after tempering at 150 °C. According to Figure 7.53 the changes are even less after tempering at 200 and 260 °C. The conclusion to be drawn from these tests is that tempering at 180-200 °C is adequate to stabilize steel W 1. Table 7.11 shows the values obtained during the corresponding investigations made on steel O 1.

**Table 7.11** O 1. Change in length as a result of ageing at 20 °C

<table>
<thead>
<tr>
<th>Treatment temperatures °C</th>
<th>Hardness HRC</th>
<th>Change in length L x 10^-6 after ageing at 20 °C</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>1 week 1 month 3 months 1 year 3 years</td>
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<tr>
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<td>63</td>
<td>-14    +30   -48   -87   -1</td>
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<tr>
<td>150, -196</td>
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<td>-3     -5    -7    -8    -8</td>
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<td>150, -196, 150</td>
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<td>-5     -8    -10   -13   -13</td>
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The results are fundamentally similar to those obtained with steel W 1. Tests made on steel L 3 also exhibit largely similar results and the values obtained with this steel after hardening and tempering at different temperatures, and after hardening, subzero-cooling and tempering respectively, are therefore likely to be applicable to steel O 1 (see Figures 7.54 and 7.55).

After hardening and tempering at 150-200 °C the effects of ageing can be ignored, according to these experiments. If tempering is undertaken after
DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

523

Subzero-cooling to \(-196\) \(^\circ\)C, it must on the other hand be performed at 260 \(^\circ\)C in order to prevent any reduction in volume during ageing.

Experiments involving martempering of steel O 1 showed that an increase in volume occurs during ageing if tempering is undertaken at 120 \(^\circ\)C and that the changes in dimensions can be ignored after tempering at 150-

200 \(^\circ\)C.

'Artificial ageing', i.e. heating to 100-150 \(^\circ\)C, e.g. for 24 h, does not result in any improvement as regards dimensional stability at room temperature for these steels if they have already been tempered in the normal temperature range, i.e. 180-200 \(^\circ\)C. This fact has been confirmed by an extensive investigation on steel S 18 2092 (Bofors SR 1855), 30 mm diameter \times 100 mm. After hardening and tempering at 170 \(^\circ\)C for 1 h the specimens were aged at temperatures between 130-170 \(^\circ\)C for 6-96 h. After this ageing process quite large dimensional changes were observed. Subsequently both the aged and the un-aged test bars were kept at room temperature for 42 months, during which time two measurements were made at definite intervals. Except for some few values all the dimensional changes measured during this 'natural ageing' were less than \(\pm 0.003\) mm. With regard to the majority of measuring gauges such dimensional changes are of very little practical importance.

In all the cases mentioned above only one tempering operation was performed, but since this took place after the steel had (after hardening) cooled to room temperature, double tempering was not required. In those cases where stringent requirements are imposed regarding dimensional stability, to be on the safe side a double tempering should always be carried out. It is then also important that the tool should be allowed to cool slowly to room temperature so that no thermal stresses are created.

Since measuring appliances can be subjected to forces which may involve the transformation of retained austenite to martensite it is most advisable to undertake subzero-cooling between the tempering operations.

With high-alloy steels such as D 6 and M 2 the conditions prevailing during ageing are much more complicated. If extremely high dimensional stability is required attention must be devoted mainly to ensuring a well-defined heat treatment process, which is modified on the basis of tests made to establish susceptibility to ageing.

7.7 DESIGNING FOR HEAT TREATMENT

A very considerable responsibility rests on the heat treaters, particularly during the hardening of tools. He has to undertake a treatment which costs some pence per kilogram, whereas the kilogram price of the tool can be 10-50 times this. The risks of cracking are great even if the heat-treatment personnel are experts at their job. Their work will be greatly facilitated if the shape of the tool is such as to facilitate heat treatment. There are two simple rules which the designer should bear in mind when he designs a die or machine component which is to be heated treated, and these rules can be expressed as follows:

1. Large radii (fillets) and smooth changes of section in the material.
2. As uniform a distribution of the material as possible.

524 DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

The main causes of quenching cracks are stresses which are set up as a result of temperature differences in different parts of the material, also the transformation stresses during formation of martensite.

It is extremely difficult to avoid harmful stresses in steel during hardening. At sharp re-entrant corners, stress concentrations prevail whereby the stresses can exceed the tensile strength of the steel and result in cracking. Figures 7.56a-d show some examples of tools which cracked because of their unsuitable shape.

Figure 7.56. Examples of tools which have fractured during hardening because of unsuitable design

a. The main reason is that there is no fillet radius. The tool was made from the through-hardening steel D 2. Tensile stresses were set up in the surface layer during hardening and this resulted in cracking.

b. The plate was made from a shallow-hardening steel. The material cooled most rapidly at the corners and martensite formed first in these areas. When martensite formation then started in the sides, the compressive stresses around the holes were so great that fracture occurred. If there had been more material around the holes martensite formation would have occurred there later. Furthermore the stresses would not have been so great across a larger cross-sectional area. By insulating these holes it is possible to reduce the cooling rate and the degree of through hardening.

c. The die was made from steel O 1 and hardened in oil. It cracked at the change of section between the cylindrical parts because of dissimilar material thickness and the small fillet radius. During quenching the thin plate became martensitic first and expanded, so that the material in the upper part was subjected to tensile stresses. Cooling was not so efficient in the fillet and the dimensions of the upper part were too
large to admit of martensite formation. The upper part shrank during the cooling and the tensile stresses increased at the fillet where the structure was inhomogeneous and could not accommodate the stresses. In this case not only was the design incorrect; the choice of material was also dubious. It might have been possible to rescue the component by quenching it in a martempering bath with high cooling intensity. The chances of a successful outcome would have improved had the die been made from SIS 2550 (Bofors HR O 1243).

d. This component was also made from O 1. It was hardened in oil and cracked at the sudden change of section. Martensite formation occurred first in the smaller, cylindrical part, which caused the surrounding material to become deformed. During the martensite formation in the larger cylindrical section, the latter increased in volume, which created tensile stresses in the central parts. Furthermore the tensile stresses increased as a result of thermal shrinkage of the smaller, cylindrical part. This die would probably not have fractured, if the hole had gone right through or if the fillet radius had been at least 5 mm. Martempering is recommended for this die.

Figure 7.57 shows a mandrel made from steel W 1 which cracked in the keyway which has no bottom radius. Figure 7.58 shows a part of a die made from SIS 2550 (Bofors HR O 1243) which cracked because of sudden changes of section.

During the oil-hardening of through-hardening steels, tensile stresses are created in the surface. Very often hardness cracks are formed in such dies as

Figure 7.57. Mandrel made from steel W 1 (Bofors B 20 V) which cracked through the keyway owing to absence of fillet radius at bottom of groove

Figure 7.58. Tool made from steel SIS 2550 (Bofors HR O 1243) which cracked during hardening owing to excessively non-uniform sections

Figure 7.59. Cracked tool made from steel O 1 (Bofors RT 1733). The die has been quenched in oil. 1×
DIMENSIONAL CHANGES DURING HARDENING AND TEMPERING

8

Tables

8.1 SOME STEEL STANDARD SPECIFICATIONS AND COMPARABLE BOFORS AND UDDEHOLM (UHB) STEELS

<table>
<thead>
<tr>
<th>Table 8.1</th>
<th>AMERICAN STEEL STANDARD SPECIFICATIONS</th>
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<tr>
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<td>ROP 57</td>
</tr>
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<td>RT 60</td>
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<td>ROP 19</td>
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<td>O 1</td>
<td>RT 1733</td>
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<td>RTO 912</td>
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Stainless steels

| 302  | RIM 29 | SS 3 M |
| 303  | RIM 29 | SS 3 M |
| 304  | RIM 29 | SS 3 M |
| 304 L | RLM 290 | SS 3 L |
| 316  | RLM 215 | SS 24 |
| 316 L | RLM 217 | SS 24 L |
| 321  | RLM 294 | SS 53 |
| 339  | RJ 214 | SS 44 |
| 347  | RLM 295 | SS 63 |
| 416  | RLM 3 | SS 41 |
| 420  | 2 R 77 | SS 6 |
| 430 F | 2 R 77 | SS 72 |
| 431  | 2 R 3 | SS 22 |
| 440 C | 2 RO 189 | — |

528

REFERENCES

2. LEMEN, B. S., 'Distortion in Tool Steel', ASM (1959)
3. FREHSE, J., 'Anisotropic Dimensional Changes During the Heat Treatment of Ledgecarbide Chrome Tool Steels', Archiv für das Eisenhüttenwesen, 24, No. 11/12, 483–495 (1953) (in German)
5. AVERBACH, B. L., COHEN, M., and FLETCHER, S. G., 'The Dimensional Stability of Steel, Part III. Decomposition of Martensite and Austenite at Room Temperature', Trans. ASM, 40, 728 (1948)
6. LEMEN, B. S., and AVERBACH, B. L., Summary Report I. Measurement and Control of the Dimensional Behavior of Metals, Massachusetts Institute of Technology (September 1955)
7. LEMEN, B. S., AVERBACH, B. L., and COHEN, M., 'The Dimensional Stability of Steel, Part IV. Tool Steel', Trans. ASM, 41, 1061 (1949)

a result of sharp punch marks. The die in Figure 7.59 is a typical example. It was made from steel O 1 and hardened by oil quenching, which was resorted to because this method of hardening would give rise to a smaller increase in length than if quenching in a martempering bath were used. It was possible to solve the problem in either of the following two ways:

1. Keep the sharp punch marks, quench in a martempering bath and grind to the required dimensions afterwards.
2. Engrave by means of an engraving machine, which gives smooth transitions, and quench in oil. The client selected this solution.
### Table 8.2 British Steel Standard Specifications

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<td>En 58 J</td>
<td>R1M 215</td>
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| Constructional steels |
| 708A37 | En 19 B | RO 752 |
| 708A42 | En 19 C | RO 952 |
| 817M40 | En 24 | CRO 861 |
| — | En 29 B | RO 755 |
| — | En 80 A | KR 75 |
| 905M39 | En 32 C | B 4 V |
| 616M40 | En 41 B | AR0 75 |
| 617M17 | En 110 | CRO 861 |
| 617M17 | En 352 | DR 54 |
| 615M17 | En 353 | CRO 53 |
| 805M40 | En 362 | CRO 42 |

| Tool steels |
| BW 1 B | B 20 V | 20 |
| BO 1 | RT 733 | ARNE |
| BS 1 | RTO 912 | REGIN 3 |
| BA 2 | ROP 21 | RIGOR |
| BD 2 | ROP 57 | SVÆRKE 21 |
| BH 13 | ROP 19 | ORVAR 2 |
| BH 10A | QRO 45 | QRO 45 |

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### Table 8.3 German Steel Standard Specifications

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Stainless steels |
| 4021 | X 20 Cr 13 | 2 R 47 | SS 31 |
| 4034 | X 40 Cr 13 | 2 R 77 | SS 26 |
| 4057 | X 22 CrNi 17 | RE 39 | SS 22 |
| 4084 | X 12 CrMo 17 | 2 R 3 | SS 72 |
| 4112 | X 90 CrMoV 18 | (2 RO 189) | — |
| 4300 | X 12 CrNi 18 3 | R1M 29 | SS 3 |
| 4301 | X 5 CrNi 8 9 | R1M 291 | SS 3 MM |
| 4305 | X 12 CrNi 18 8 | RLH 2 | SS 43 |
| 4306 | X 7 CrNi 8 9 | R1M 290 | (SS 3 L) |
| 4463 | 5X CrNiMo 18 12 | R1M 217 | SS 24 L |
| 4436 | 6X CrNiMo 18 12 | R1M 215 | SS 24 |
| 4460 | 8X CrNiMo 27 5 | R1M 214 | SS 44 |
| 4541 | X 10 CrNiTi 18 9 | R1M 204 | SS 33 |
| 4590 | X 10 CrNiN 18 | R1M 295 | SS 63 |
| 4571 | X 10 CrNiMoTi 18 10 | R1M 213 | SS 34 |

Constructional steels |
<p>| 1141 | Ck 15 | B 4 V |
| 4922 | X 22 CrMoV 12 1 | ROP 43 |
| 5864 | 35 NiCr 18 | R 75 |
| 6882 | 34 CrNiMo 6 | CRO 861 |
| 7218 | 25 CrMo 4 | RO 653 |
| 7220 | 34 CrMo 4 | RO 752 |
| 7225 | 42 CrMo 4 | RO 952 |
| 7228 | 24 CrMo 5 | RO 653 |
| 7337 | 16 CrMo 4 4 | RO 346 |
| 7761 | 32 CrMo 12 | RO 1155 |
| 7380 | 10 CrMo 9 10 | RO 211 |
| 7723 | 24 CrMoV 5 5 | ROP 5462 |
| 8507 | 34 CrAlMo 5 | (ARO 75) |
| TABLE 8.4  SWEDISH STEEL STANDARD SPECIFICATIONS |
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### Table 8.7 ROUND AND SQUARE BARS, METRIC UNITS

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*Multi-Material High Speed Steel (shank) weigh approximately 10% more.*

*High Speed Steel of 44.5-15.9 (pipe weigh approximately 10% less).*
### Table 8.8 Hexagonal and Octagonal Bars, Metric Units

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### Table 8.9 Flat Bars, Width 10 to 49 mm, Metric Units

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Table 8.10 Flat Bar Width 45 to 130 mm, Metric Units

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*Molybdylon High Speed steels weigh approximately 15% more. High Speed Steels of 18-4-1 type weigh approximately 10% more.*
### Table 8.11 FLAT BARS, WIDTH 140 to 350 MM, METRIC UNITS

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### Table 8.12 ROUND AND SQUARE BARS, INCH UNITS

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Notes:
- Units: Length in feet, weight in pounds.
- Material: High-speed steel.
- Density: 0.00287 lb/ft$^3$.

Multihole: High-speed steel weight approximately 30% more.

High-speed steel of HSS-1 type weight approximately 10% more.
### Table 8.13 Hexagonal and Octagonal Bars, Inch Units

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</tr>
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<td>0.23</td>
</tr>
<tr>
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<tr>
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### Table 8.14 Flat Bars, Width 8 to 15 in. Inch Units

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<th>Width 1</th>
<th>Width 1 1/4</th>
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<td>0.37</td>
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### Table 8.15 Flat Bars, Width 8 to 15 in. Inch Units

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Molybdenum High Speed Steels weigh approximately 7% more.
High Speed Steels of 18-41 type weigh approximately 10% more.
### Table 8.17 Conversion Table for Celsius (Centigrade) and Fahrenheit

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The known number of degrees centigrade or Fahrenheit is found in the line of figures marked O, whatever the desired number of degrees is read off in the left or right column of figures.
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Conversion Table for "Celsius (Centigrade) and Fahrenheit (32°F to 212°F)"

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**The known number of degrees centigrade or Fahrenheit is found in the line of figures marked O, whereas the desired number of degrees is read off in the left or right column of figures.**

### TABLE 8.18
Inches to Millimetres (1 in = 25.40 mm)

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**Note:** The table provides the conversion of inches to millimetres, rounded to the nearest tenth of a millimetre. For precise conversions, consult a more detailed reference.
Table 8.18b: INCHES TO MILLIMETRES
(1 in = 25.40 mm)

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Table 8.19: DECIMALS OF INCHES TO MILLIMETRES

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### 8.5 Conversion Tables for Weight (Mass)

#### Table 8.20 Pounds to Kilograms

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### Table 8.21 British Units to Kilograms

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#### Notes

- 16 ounces = 1 pound (lb)
- 28 pounds = 1 quarter (qr)
- 4 quarters = 1 hundredweight (cwt)
- 20 hundredweights = 1 ton

- 1 ounce (oz) = 0.028 kg
- 1 pound (lb) = 0.454 kg
- 1 quarter (qr) = 12.7 kg
- 1 hundredweight (cwt) = 50.8 kg
- 1 ton = 1016.05 kg
### Table 8.23 Tons per square inch to newtons per square millimetre

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<th>Tons/in² (tonf/in²)</th>
<th>N/mm² (N/m²)</th>
<th>Tons/in² (tonf/in²)</th>
<th>N/mm² (N/m²)</th>
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- **Note:** The table lists conversions from tons per square inch to newtons per square millimetre (tons/in² to N/mm²), providing values for various pressure intensities.
### Table 8.24 Kiloponds per Square Millimetre to Newtons per Square Millimetre

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### Table 8.25 Pounds per Inch Square to Kiloponds per Square Millimetre

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(1000 lb per sq. inch = 0.703 kp per \(\text{mm}^2\))
### 8.7 Conversion Tables for Energy

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#### Table 8.27 Footpounds to Kilopond Metres

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- 556 TABLPS
### Table 8.28: Kilopond Metres to Joules

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### 8.8 Conversion Table for Fracture Toughness Units

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<td>16114.5</td>
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<tr>
<td>10</td>
<td>9.010</td>
<td>316.23</td>
<td>32235.0</td>
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1 hbar $\sqrt{cm} = 1 MN m^2$
### Table 8.31 Common SI Units to British Units

<table>
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<th>Length</th>
<th>Density</th>
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<tr>
<td>1 km</td>
<td>1 kg/m³</td>
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<tr>
<td>1 m</td>
<td>1 N</td>
</tr>
<tr>
<td>1 mm</td>
<td>0.039 370 in</td>
</tr>
<tr>
<td>1 µm</td>
<td>0.001 591 5 in²</td>
</tr>
<tr>
<td>1 mm²</td>
<td>1 N/m²</td>
</tr>
<tr>
<td>1 m²</td>
<td>1 Pa</td>
</tr>
<tr>
<td>1 m³</td>
<td>1 G</td>
</tr>
<tr>
<td>1 cm³</td>
<td>0.001 591 5 in³</td>
</tr>
<tr>
<td>1 mm³</td>
<td>0.001 550 000 in³</td>
</tr>
</tbody>
</table>

### Table 8.32 Common Non-SI Metric Units to British Units

| Length       |  
|--------------|------------------|
| 1 Å          | 0.003 937 015 µm |
| 1 cm         | 0.39 370 in      |
| 1 m          | 3.28 14 ft       |
| 1 km         | 0.62 1 371 mi    |

| Area         |  
|--------------|------------------|
| 1 ha         | 247 105 acres    |
| 1 acre       | 0.40 468 6 yd²   |
| 1 m²         | 16 145.9 ft²     |

| Force        |  
|--------------|------------------|
| 1 N          | 0.22 4 809 lbf   |
| 1 kgf        | 2.2 4 809 lbf    |
| 1 dyn         | 0.22 4 809 lbf   |

| Pressure     |  
|--------------|------------------|
| 1 bar        | 14 503.8 lb/in²  |
| 1 atm (technical atmosphere) | 14 503.8 lb/in²  |

| Torque       |  
|--------------|------------------|
| 1 lbf·ft     | 1.3 8 21 N·m    |
| 1 ft·lbf     | 1.3 8 21 N·m    |

| Energy       |  
|--------------|------------------|
| 1 J          | 0.737 56 lbf·ft  |
| 1 kgf·m      | 0.737 56 lbf·ft  |
| 1 lbf·ft     | 0.737 56 lbf·ft  |

* The US gallon has been included for convenience.
### Table 8.34 ATOMIC NUMBER AND ATOMIC WEIGHT

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<tr>
<th>Symbol</th>
<th>Name</th>
<th>Atomic number</th>
<th>Atomic weight</th>
<th>Symbol</th>
<th>Name</th>
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<th>Atomic weight</th>
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**Notes:**
1. The above atomic weights are based on the exact number 12 for the carbon isotope 12, as agreed between the International Union of Pure and Applied Physics and of Physico and Applied Chemistry (1961).
2. Atomic weights of isotopes found in nature. Particular isotopes are chosen as the values for hydrogen, boron, carbon, oxygen, silicon and sulphur, when the decision shown is due to variations in relative concentrations of isotopes.
Index

Ageing, 253
Artificial, 523
Dimensional changes, 520–523
Air cooling, 226
Cooling curves, 227
Alloy steels, 324
TTT diagrams for isothermal transformation, 95
Alloying elements, 82–126, 264, 375
Austenite-forming, 82
Carbide-forming, 85
Carbide-stabilizing, 85
Effect on eutectoid point, 89
Effect on ferrite hardness, 87
Effect on grain growth, 88
Effect on martensite formation temperature, 90–93
Effect on pearlite and bainite formation, 93
Effect on resistance to tempering, 102–106
Ferrite-forming, 83
Inductive effects, 95
Nitriding, 86, 379, 395
α-iron, 1, 2
Aluminium, 83, 88, 113, 253, 259, 395
Aluminium-alloyed nitriding steels, 407
Ammonia, nitriding with, 86 Nitriding
Annealing, 108, 110, 177–191
Full, 177, 188
High-speed steels, 181–183
Homogenizing, 191
Calculation, 28
Hydrogen, 108, 191
Isothermal, 179, 188
Optimum machinability, 183–184
Quench, 188–190
Recrystallization, 184
Spheroidizing, 177–181
Stress-relief, 184–185
Structure, 177–181
Temperature, 184
Aquaquench, 226
Arrest lines, 59
ASTM Grain Size Number, 34
Atomic number, 562
Atomic structure, 2
Atomic weight, 562
Austempering, 217–219
Hold time at temperature, 217
Austenite, 1, 2, 6, 7
Grain size, 14, 132
Retained, 86 Retained austenite
Austenite-forming elements, 82
Austenitic thermochemical treatments, 377
Austempering, 188
Bainite, formation, 14, 22, 93–95, 245
Lower, 15, 162
Upper, 15, 162
Ball-bearing steels, 473
Barium chloride baths, 195
Blanking dies, 267, 483, 491
Blanking tools, 287, 319, 462
Body-centred cubic lattice, 2, 47
Borax, 198
Boron, 102
Brielli test, 42
Brielli fracture, 73, 76, 113
Brittleness, nitriding, 389
Ranges, 246–248
Temper, 246–248
Carbide, 209–212, 319, 340, 471
Distribution, 398
Double or complex, 85, 88
Carbide-forming elements, 85
Carbide particle size, 38
Carbide precipitation, 183, 188–190, 295, 297, 473
Carbide stabilizers, 85
Carbon, 4, 82, 92, 95, 112, 127, 351, 376, 396, 427
Carbon penetration, depth of, 340, 342, 359
Carbon potential, 340, 356
Carbon steels, 93–94, 177, 261, 266, 320, 454–456, 462, 470, 520
Carbonitrides, 398
Carbonitriding, 427–432
Carburizing, 339
Box, 347
Calculation, 25
Compounds, 339, 340, 345, 347, 348, 370
Depth of, 340, 348
Gas, 355–358, 366, 371
Methods, 345–348
Pack, 345, 364, 370
Protection against, 372–374
Protective pastes, 373
Salt-baths, 350, 370
Selective, 372–374
Temperature, 364, 370
Carpenter torsion impact test, 57–58
Carrier gas, 200
Case-hardened steels, 88, 344
Case hardening, 339–376
Core hardness, 367–370
Core strength, 375–376
Depth, 340, 359–361, 374–376, 506
Dimensional changes during, 506–514
Recommendations, 370–371
Surface hardness, 361–367
Tool steels, 371–372
Case-hardening steels, 38, 141, 183, 188, 342–345, 409
Cast-iron chips, 195, 198, 229
CCT diagram, 20, 102, 103, 128, 152–161, 177, 295
Cementite, 6, 7, 93
Charyp U test, 76
Charyp U value, 54–56
Charyp V test, 75–76
Charyp V value, 54–56
Chromium, 82, 83, 85, 87, 90, 95, 133, 188
Chromium–manganese steels, 364
Chromium–molybdenum nitriding steels, 407
Chromium–molybdenum steels, 324
Chromium–nickel–molybdenum steels, 324–326
Chromium steels, 330
Cold hardening, 309
Cold punch annealing, 275
Cold-work steels, 261–290
dimensional changes during hardening and tempering, 475
Compresive stress, 171, 173, 174
Connecting sleeve, 511
Constructional steels, 39, 93, 125, 258, 319–339, 529–533
Continuous-heating curves, 11
Contour hardening, 169, 268, 448
Controlled atmospheres, 199
Cooling curves, 102, 227
Air cooling, 227
Oil quenching, 223
Water quenching, 221
Copper plating, 373–374
Corrosion, intergranular, 189
Corrosion resistance, nitrided steels, 424–427
Crack formation, 220, 232–233, 244–245, 258, 323, 524
Crack initiation, 74, 174–175
Crack Opening Displacement, 77
Creep, 69–70
Creep curve, 72
Creep rupture strength, 70
Creep strength, 70
Creep test, 71–72
Critical diameter, 130, 132, 160
Critical Stress Intensity, 77
Crystal structure, 2, 47
C-T (Compact Tension) test bar, 76
Curie Point, 2
Cutting tools, 317, 318
Cyamid bath, 297, 350–355, 374
Cyamide salts, 355
Decarburization, 27, 194, 198, 202, 258
δ-iron, 1
Decoxidation, 113
Design for heat treatment, 523–527
Desulphurization, 121
Dew point, 355, 356
Die block, 240, 274
Die-casting dies, 297, 304
INDEX 665

Dyes, 216, 476, 486, 498, 499, 504
Diffusion, 23–29, 106, 110, 113, 427–428
coefficient, 24
distance calculation, 25
equation, 23–24, 342, 401
nature of, 23
rate of, 24

Dimensional changes, 529–523
aging, 529–523
case hardening, 506–514
dimensioning and tempering of tool steels, 474–506
nitrizing, 394, 514–519
tempering, 472–474

Dimples, 73

Dislocations, 29–32, 250, 253
Dissoociation pipette for nitrizing, 381
Drawing tools, 287
Drop-forging dies, 274, 292
DVM impact specimen, 55–56

Edge dislocation, 29
Electric-arc furnace, 116
Electrodes for refining, 117
Elongation, 51, 53, 113
Emulsions, 223
Endothermic gas, 200
Endurance limit, 64, 175, 338
and hardness, 65
and surface finish, 66–68
and ultimate tensile strength, 65
influence of change of section, 69
ways of increasing, 69
Energy conversion tables, 555
Eutectoid point, 89–90
Eutectoid steels, 5, 13
Exothermic gas, 200
Extrusion dies, 310

Face-centred cubic lattice, 2, 74
Fatigue, 58
formation, 93
hardness, 87
Fatigue-forming elements, 83
Ferrite, 1

INDEX 665

Fine-grained steels, 88, 89
Flame hardening, 210, 451–464
blanking tools, 462
factors influencing hardness and depth of hardening, 454–458
grains, 458
long objects, 459–461
machine components and tools, 458–464
manual, 451
methods, 451–454
moulds, 462–464
plates, 461
progressive, 452, 454
spin hardening, 452, 454
steel grades for, 434

Fracture, 304
Fracture, brittle, 73, 76, 113
cleavage, 72
ductile, 73, 76
fatigue, 59, 62, 375, 422
mechanism, 76
shear, 72
stress, 74
transition from ductile to brittle, 74–75
Fracture test for grain size, 36
Fracture toughness, 76–80
and yield strength, 77–80
implication of, 80
Fracture toughness units, conversion table for, 558
Furnaces, 3

gas, 195

Iron, 2
Gas, 8

controlled atmosphere, 199–201
endothermic, 200
exothermic, 200
in steel, 106–116
inert, 200
Gears, 360, 407, 438, 452, 458, 513
Glue-nitriding, 534

Goodman diagram, 64, 66
Grain boundaries, 32
Grain-boundary carbide, 5, 295
Grain growth, 88, 182, 183, 190, 294
Grain-refining, 181, 193
Grain size, 32–41, 132, 160, 181, 184, 206, 209, 294, 398
ASTM standard classified, 34
austenite, 34
case-hardening steels, 38
constructional steels, 39
fracture test for, 36
high-speed steels, 39
impurity method, 36
methods of determining, 34–38
tool steels, 39
Großmann hardenability test, 130–137
Hadfield steel, 82, 83, 185, 189, 249, 339
Hardenability, 127–176, 259, 264, 361, 428–431, 506
calculations, 132, 133
high, 160
low, 163–169
practical application, 161–169
Hardenability band, 140

Hardenability factor, 134
Hardenability test, Großmann, 130–137
Jominy end-quench, 137–151
Hardening, 194–233, 339
Hardenability test, 233, 339
Case Hardening, 169, 268, 448
controlled atmosphere, 199–201
cooling methods, 212–219
depth of, 127–128, 135, 167, 168, 169,
174, 267, 268, 123
dimensional changes during, 466–471
Flame, 194–202
heating rate, 202
heating time at temperature, 7, 209, 211
holding time vs. heating-up time, 212
induction, 202
See Induction hardening
Insufficiency, 245
maching before, 258
on mandrels, 510
packing media, 198
protective agents and pastes, 198–199
quenching equipment, 230–233
quenching media, 219–230
quenching range, 209
secondary, 282
temperature, 182, 183, 206
vacuum, 201–202
Hardness, 401
carbon content and martensite relationship, 127
conversion table, 49
ferrite, 87
normalized plain-carbon and low-alloy steels, 193
test, 42–51
H-factor, 120
High-alloy steels, 324
High-duty steels, 306
High-speed steels, 209, 211, 310–319, 416
annealing of, 181–183
dimensional changes during hardening and tempering, 502
grain size, 39
High-strength steels, 320, 335
High-temperature properties, 294, 302, 306
Homogenizing annealing, 28, 191
Hot-work steels, 290–310
dimensional changes during hardening and tempering, 496
H-steels, 141
Hydrogen, 106–110
annealing, 108, 191
diffusion, 106
embrittlement, 191
nitrizing, 383
removal, 108, 191
solubility, 107
Hyper-eutectoid steels, 6, 7, 13, 180
Hypo-eutectoid steels, 6, 177
Impact strength, 124, 295
Impact stress, 175
Impact test, 53–57, 75–76, 113
Inclusions, 113–114, 119, 122
Induction hardening, 210, 432–451
advantages and disadvantages, 450–451
depth of penetration, 433
equipment, 435
factors influencing hardness and depth of hardening, 400–447
flank, 449
frequency selection, 435
fundamental principles, 432
gears, 447–449
machine components, 447–449
procedure, 437–440
progressive, 439–440, 449
shocks, 449
single-shot, 438, 449
INDEX 567

568 INDEX

Nitriding (cont)
salt bath (cont) comparison with gas nitriding, 387–395
spalling during, 400
stopping-off agents, 382
surface appearance, 400
surface hardness, 388
surface smoothness, 389
white layer, 378, 404
with ammonia, 379
with ammonia and hydrocarbons, 383
with ammonia and nitrogen or hydrogen, 383
Nitriding box, 380–382
Nitriding steels, 259, 395, 407–418
Nitrogen, 86, 110–113, 200, 351, 379, 427
activity during nitriding, 403–404
diffusion, 410, 427–429
harmful effects, 112
influence on properties of steel, 111–113
solubility, 110
Non-deforming steels, 400
Normalizing, 180, 181, 191–194
double, 194
Notch sensitivity, 69, 376, 422
Notched test specimen, 74–75
Oil quenching, 221
cooling curves, 223
Oxidation, 202, 454
Oxygen, 113–116
diffusion, 113
harmful effects, 113
influence on properties of steel, 113–116
Packaging media, 198
Patenting, 219
Pearlite, 4, 7, 13–14, 93–95
Peeving, 257
Pilger roll, 268, 271
Plane strain, 76
Plane stress, 76
Planning tools, 313
Plates, 461, 476, 485, 487, 490, 500, 502
Potassium nitrite, 227
Powder nitriding, 387
Precipitation hardening, 87, 253–257
Prehardening steels, 227
Preheating, 202, 205
Proof stress, 51
Protective agents and pastes, 198–199, 373
Punching dies, 266
Punching tools, 218, 287, 319
Quenching, 214
direct, 339
double, 339
equipment, 230–233
jig, 231
press, 231
single, 339
temperature, 361, 364
Quenching media, 219–230
agitation, 230
cooling capacities, 230
temperature, 230
Quenching tanks, 230
Recrystallization annealing, 184
Relaxation during tempering, 473
Residual stress, 169
Retained austenite, 18–20, 92, 206, 245,
269, 283–286, 317, 361, 364, 366, 457,
470–471
decomposition on tempering, 20–23
mechanical transformation, 249
stabilization, 250–252
transformation, 249–252
Rings, 473, 475, 481, 482, 499, 506, 508,
516, 517
River pattern, 73
Rockwell test, 44
Rolls, 163, 164, 167, 175, 216, 435, 436,
437, 440, 443, 444, 493
Salt-bath carburizing, 350, 370
Salt-bath furnaces, 194, 234
Salt-bath nitriding, 385
advantages and disadvantages, 394
comparison with gas nitriding, 387–395
Salt-bath quenching, 227
Schaeffler diagrams, 45
Sclerometry test, 47
Sendzimir rolls, 290
Shafts, induction hardening, 449
Shallow-hardening, 266
Shear, 29, 72
Shear bands, 376
Shear stress, 374–375
Shepherd’s fracture number, 38
Shore sclerometer test, 48
SI designations, tensile properties, 51
SI units, conversion, 559
Silicon, 83, 84, 93, 95, 106, 135
Size, conversion tables for, 546

INDEX 567

Induction hardening (cont)
steel grades, 435
Induction heating, coils, 436
tempering, 447
Inert gas, 200
Intercept method for grain size, 36
 Ionitriding, 384
Iron, crystal structure, 2
magnetism, 2
transformations, 1
Iron carbide, 4
Iron-carbon equilibrium diagram, 4–6, 90, 178
Iron-nickel equilibrium diagram, 83
Iron-nitrogen equilibrium diagram, 378
ISO-index G, 34
Isothermal transformation, pearlite and bainite, 93–95
TTT diagram, 93–102
Isotropic material, 471
Izod impact specimen, 55–56
Jernkontoret’s (JK) fracture number, 38–41
Jominy curves, 140, 164–169, 432
practical applications, 141–151
Jominy end-quench hardenability test, 137–151
Knoop test, 44
Larsson–Miller parameter, 186
Lattice, body-centred cubic, 2, 74
face-centred cubic, 2, 74
Lattice parameter, 2
Low-alloy steels, 273, 324
Ludwig–Dawidenko–Orowan hypothesis, 74
Machinability, 183–184
Machinability tests, 125
Machining allowances, 258–259, 372
McQuaid–Ehne method, 38
Mandrels, 280, 510, 525
Manganese, 82, 85, 90, 93, 95
Manganese steels, 190, 249, 339
Maraging steels, 257, 338
Martempering, 18, 175, 214–217, 257, 500
Martensite, acicular, 17
decomposition on tempering, 20–23
Martensite (cont)
formation, 15, 127, 225, 250, 258
temperature, 90–93
massive, 17
structure, 17
untempered, 245
Master curve, 240, 243
Master parameter curve, 22–23
Materials testing, 42–81
Messenner impact specimen, 55–56
Micro-alloyed steels, 89
Milling cutters, 319
Molybdenum, 83, 88, 95, 133, 253
Moulds, 462–464, 485
Nickel, 82, 85, 93, 95
Nickel-base alloys, 253–257
Nimonic alloys, 254
Niobium, 88, 89
Nitridability, 395–406
influence of micro-structure, 396
Nitride formation, 389
Nitride-forming elements, 86, 379, 395
Nitried steels,
corrosion resistance, 424–427
fatigue strength, 422
high temperature to tempering and high hardness at elevated temperatures, 420
notch sensitivity, 422
properties, 418–427
wear resistance, 418–420
Nitrides, 112, 378–379, 385
Nitriding, 86, 112, 377–407
brittleness, 389
degree of dissociation, 404–406
depth, 406
development, 388
dimensional changes during, 514–519
dimensional stability, 394
dissociation pipe, 381
effect of time on depth of hardness, 401
gas, 379
advantages and disadvantages, 394
comparison with salt-bath, 387–395
glow discharge, 384
method, 379–395
nitrogen activity during, 403–404
powder, 387
preparation of parts, 382
properties imparted to steel by, 378
salt-bath, 385
advantages and disadvantages, 394
INDEX 568

Slag Inclusion Chart, 114
Sling plant, 29
Slitting knives, 270
Sodium cyanide, 350, 355
Sodium nitrite, 227
Solid solution, 23
Substitutional, 23
Solution treatment, 253, 254
Spalling during nitriding, 400
Specific volume of phases, 469
Spheroidizing, 7, 177–181
Split-ring test, 473
Spring steels, 333
Stacking faults, 250
Stainless steels, 188, 330, 416, 528–532
Steel bars, weight tables for, 534
Steel standard specifications, 528
Steelmaking, new processes, 116–126
Straightening, 257–258
Strain-ageing, 112
Strain-hardening, 74
Stress concentration factor, 68
Stress condition during tempering, 473
Stress conversion tables, 551
Stress diagram, 170–171, 176
Stress distribution, 171, 172, 174
Stress pattern, influence of depth of hardening, 169
Stress relaxation, 185–186
Stress-relief annealing, 184–188
Striations, 59
Structure treating, 304
Sub-zero treatment, 18, 249, 250–252, 284, 317, 479, 521
Sulfur treatment, 385–386, 419, 420, 426
Sulphides, 122
Sulfur, effect on properties of steel, 120–126
removal, 121
Surface finish and fatigue stress, 66–68
Tantalum, 253
Temper brittleness, 246–248
Temper colour, 235
Temperature conversion tables, 542
Tempering, 20–23, 233–249
after grinding or electrical spark machining, 246
case-hardened steels, 371
dimensional changes during, 472–474
double, 244
Tempering (cont)
effect of alloying elements on resistance to, 102–106
hardness and temperature relationship, 22
heating to temperature, 234
heating rate, 234–240
holding time, 240–244
induction heating, 447
relaxation, 473
stress conditions, 473
temperature calculation, 105
tool steels, dimensional changes, 474–506
volume changes, 472
Tempering diagram, 241, 242, 245
Tempering parameter, 23, 240
Tensile strength, 51
Tensile stress, 171–175
Tensile test, 51–53
Ternary phase diagrams, 84
Thermal stress, 170, 172, 185, 220, 466, 474
Through-hardening steels, 525
Time-temperature-transformation, 6–20
Titanium, 88, 89, 253, 395
tool steels, 85, 93, 195, 198, 210, 258,
261–319, 412, 456, 469, 528–533
 ageing, 520
case hardening, 371–372
cold-work, 261
designations, 262
dimensional changes during hardening and tempering, 474–506
grain size, 39
high-alloy cold-work, 276
hot-work, 290
Torsion impact test, 57–58, 318
Tough-hardening, 320
Toughness, 75, 217, 219, 233, 275, 294,
317, 395
Transformation, changes in volume during, 470
Transformation stress, 170, 220, 467
Transformer sheet, 84
Transition temperature, 55, 75, 113, 124,
247
TTT diagram, 8, 12, 19, 20, 21, 93–102,
151–152, 177, 179, 214, 218, 245, 247,
428, 489
Tufftride treatment, 385–389
Tungsten, 83, 88, 95
Twining boundaries, 33

570 INDEX

Ultimate tensile strength, 51
and endurance limit, 65
Units, conversion, 559–561
Water quenching, 219, 323
cooling curves, 221
special additives, 225
Water turbines, 333
Wear resistance, nitried steels, 418–420
Weight (mass) conversion tables for, 549
Weight tables for steel bars, 534
Whisker, 29
White layer, 378, 404
Wöhler diagram, 62
Work hardening, 30

Yield point, 51
Yield stress, 51, 65
Yoder rolls, 271
Young's Modulus of Elasticity, 51, 53