12_____ The Heat-Treatment of Plain Carbon Steels—(II)

12.10 In the previous chapter those heat-treatment processes were discussed in which the steel component was permitted to reach a state of thermal equilibrium at ambient temperature. That is, cooling took place sufficiently slowly to allow a pearlitic type of microstructure to form. Such treatments are normally only useful for improving the toughness and ductility of a steel component, and when increased hardness is required it is necessary to quench, or cool, the component sufficiently rapidly, in order to prevent the normal pearlitic structure from being formed. If a combination of strength and toughness is necessary then a further 'tempering' process may follow quenching. Alternatively one of the isothermal treatments may be used to replace the dual treatments of quenching and tempering.

12.11 Prior to the development of metallurgy as a science many of the processes associated with the hardening of steel were clothed in mystery. For example, it was thought that the water of Sheffield possessed certain magical properties, and it is said that an astute Yorkshire business man once exported it in barrels to Japan at considerable profit. In point of fact the high quality of Sheffield steel was a measure of the craftsmanship used in its production. Similarly, it is reported that Damascus steel swords were hardened by plunging the blade, whilst hot, into the newly decapitated body of a slave and stirring vigorously. Some metallurgists have suggested, possibly more out of cynicism than scientific accuracy, that hardening would be assisted by nitrogen absorption from the blood of the slave during this somewhat gruesome procedure. James Bowie, originator of the Bowie knife in the days of the 'Wild West', is said to have quenched his knives *nine times* in succession in panther oil.

In this chapter, then, we shall deal with the production of structures, other than pearlite, in plain carbon steels, and seek to explain the relationship which exists between the mechanical properties and the crystal structure produced by the treatments employed.

Hardening

12.20 When a piece of steel, containing sufficient carbon, is cooled rapidly from above its upper critical temperature it becomes considerably harder than it would be if allowed to cool slowly. The degree of hardness produced can vary, and is dependent upon such factors as the initial quenching temperature; the size of the work; the constitution, properties and temperature of the quenching medium; and the degree of agitation and final temperature of the quenching medium.

12.21 Whenever a metallic alloy is quenched there is a tendency to suppress structural change or transformation. Frequently, therefore, it is possible to 'trap' a metallic structure as it existed at a higher temperature and so preserve it at room temperature. This is usually an easy matter with alloys in which transformation is sluggish, but in iron-carbon alloys the reverse tends to be the case. Here, transformation, particularly that of austenite to pearlite, is rapid and is easily accomplished during ordinary air-cooling to ambient temperature. This is due largely to the polymorphic transformation which takes place but also to rapid diffusion of carbon atoms in the face-centred cubic lattice of iron. The rapid diffusion of carbon atoms is a result of their smaller size and the fact that they dissolve interstitially (This also leads to the absence of coring with respect to carbon in cast steels.)

When a plain carbon steel is quenched from its austenitic range it is not possible to trap austenite and so preserve it at room temperature. Instead, one or other phases is obtained intermediate between austenite on the one hand and pearlite on the other. These phases vary in degree of hardness, but all are harder than either pearlite or austenite.

Water quenching of a steel containing sufficient carbon produces 12.22 an extremely hard structure called *martensite* which appears under the microscope as a mass of uniform needle-shaped crystals (Plate 12.1A). These 'needles' are in fact cross-sections through lens- or discus-shaped crystals—another instance of the misleading impression sometimes given by the two-dimensional image offered by the metallurgical microscope. Since martensite is of uniform appearance even at very high magnifications it follows that the carbon is still in solution in the iron and has not been precipitated as iron carbide as it would have been if the steel had been cooled under equilibrium conditions. However, X-ray crystallographic examination of martensite shows that despite very rapid cooling which has prevented the precipitation of iron carbide, the lattice structure has nevertheless changed from FCC (face-centred cubic) to something approaching the BCC (body-centred cubic) structure which is normally present in a steel cooled slowly to ambient temperature. This BCC type structure is considerably supersaturated with carbon since at ambient temperatures only 0.006% carbon is retained in solution under equilibrium conditions. Consequently the presence of dissolved carbon in amounts of, say, 0.5% can be expected to cause considerable distortion of the structure and in fact produces one which is body-centred tetragonal.

The transformation of a single crystal of martensite from austenite appears to be achieved in about 10^{-7} seconds. How can this change in structure take place so rapidly? It is suggested that a process of diffusionless phase transformation is involved, that is, there is an extremely limited movement of iron and carbon atoms into positions more nearly approaching equilibrium at the lower temperature. The lattice structure of austenite is represented in Fig. 12.1. This is the FCC structure with carbon atoms able to occupy interstitial positions as indicated*. If we regard the superimposed figure, indicated in heavier line with a base ABCD, this shows how FCC austenite can be regarded as a body-centred tetragonal structure and it is thought that martensite transformation involves a change from this structure to a true body-centred tetragonal structure with very little consequent movement of the iron atoms. In Fig. 12.2 the alteration in dimensions of a unit cell from the original austenite to the new BCT (body-centred tetragonal) is indicated. Thus the unit becomes more 'squat' in shape as the 'a₀' axis shrinks and the $a_0/\sqrt{2}$ axis expands. But for the presence of the carbon atoms inherited from the austenite the structure would transform to simple BCC ferrite. The interstitially dissolved carbon atoms have little chance to move and must fit into the new structure where they will cause considerable distortion since in a body-centred structure there are far fewer interstitial sites available. Since not all of the interstitial sites will be occupied by carbon atoms a structure something like that shown in Fig. 12.3 will form.



Fig. 12.1 The face-centred cubic structure of austenite showing its relationship to a bodycentred tetragonal cell based on *ABCD*.

The actual change from FCC to BCT involves a very small movement of atoms and probably proceeds in a manner similar to that in mechanical twinning. Movement of dislocations due to the shear involved can have an effect akin to severe work-hardening. This, in conjunction with the great distortion produced by the interstitial carbon atoms helps to explain the great hardness and negligible ductility of martensite. The presence of any carbon in excess of 0.02% will frustrate the formation of a simple BCC structure when such a steel is quenched from the austenitic range. The

* In steel a maximum of less than one in six of these positions are ever occupied by a carbon atoms.



Fig. 12.2 The transformation of FCC austenite to BCT martensite. The austenite 'tetragonal unit' shown above is the one outlined in Fig. 12.1. The sides, a_0 , have contracted to 'c', and the sides $a_0/\sqrt{2}$ have expanded to a_1 .



Fig. 12.3 The possible shape of a BCT martensite cell containing only one interstitial carbon atom (i). In any one martensite crystal all carbon atoms occupy the same interstitial position on the 'c' axis. (Iron atom X is displaced by the carbon atom of the cell 'above' it). In (ii) a normal BCC ferrite cell is shown for comparison.

degree of distortion existing in the resulting tetragonal martensite will be proportional to the overall carbon content. Consequently as the carbon content increases so does hardness.

Less severe quenching gives rise to a structure known as *Bainite*. This phase appears under the microscope, at magnifications in the region of \times 100, as black patches (Plates 12.1 B and c), but a higher magnification of \times 1000 shows that it is of a laminated nature something like pearlite. The growth of bainite (Fig. 12.4) differs from that of pearlite in that ferrite nucleates first followed by carbide, whereas in pearlite it is the carbide



Plate 12.1 12.1A 0.5% carbon steel, water quenched from 850°C.
Entirely martensite. × 100. Etched in 2% nital.
12.1B 0.5% carbon steel, oil quenched from 780°C.
Bainite (dark) and martensite. × 750. Etched in picral-nital. (*Courtesy of United Steel Companies Ltd. Rotherham*).
12.1C 0.2% carbon steel, water quenched from 870°C on a falling gradient.

Acicular bainite (dark) and martensite. \times 1000. Etched in picral-nital. (Courtesy of United Steel Companies Ltd. Rotherham).



Fig. 12.4 The growth of bainite. As the ferrite crystals grow, so the concentration of carbon in the surrounding austenite increases until a point is reached where carbide is rejected.

which nucleates first (8.43). Bainite growth takes place quickly because the driving force is increased by a greater degree of non-equilibrium at the lower temperatures at which it is formed. Consequently particle size is too small to be seen by low-power microscopy.

Still slower rates of cooling produce normal pearlite, the coarseness of the ferrite and cementite laminations depending upon the rate of cooling. Thus, normalising leads to the formation of a fairly fine-grained structure whilst annealing produces coarse-grained structures.

In practice, factors such as composition, size and shape of the 12.23 component to be hardened dictate the rate at which it shall be cooled. Generally no attempt is made to harden plain carbon steels which contain less than 0.25% carbon since the increase in hardness produced would be so small and non-uniform for reasons which will become apparent later in this chapter (12.45). Large masses of steel of heavy section will obviously cool more slowly than small articles of thin section when quenched, so that whilst the surface skin may be martensitic, the core of a large section may be bainitic because it has cooled more slowly. If, however, small amounts of such elements as nickel, chromium or manganese are added to the steel, it will be found that the martensitic layer is much thicker than with a plain carbon steel of similar carbon content and dimensions which has been cooled at the same rate. Alloving elements therefore 'increase the depth of hardening', and they do so by slowing down the transformation rates. This is a most important feature, since it enables an alloy steel to be hardened by much less drastic quenching methods than are necessary for a plain carbon steel. The liability to produce quench-cracks, which are often the result of water-quenching, is reduced in this way. Design also affects the susceptibility to quench-cracking. Sharp variations in crosssection and the presence of sharp angles, grooves, notches and rectangular holes are all likely to cause the formation of quench-cracks. Consequently

when mass production is involved it is often more satisfactory to use a low-alloy steel containing small amounts of the cheaper elements like manganese which can then be oil quenched on a conveyor-belt system. This not only cuts labour costs but eliminates the human element from quenching, as well as minimising distortion and cracking and providing a more uniform product.

12.24 The quenching medium is chosen according to the rate at which it is desired to cool the steel. The following list of media is arranged in order of quenching speeds:

5% Caustic soda 5--20% Brine Cold water Warm water Mineral oil Animal oil Vegetable oil

The very drastic quench resulting from the use of caustic soda solution is used only when extreme hardness is required in components of simple shape. For more complicated shapes an oil-quenched alloy steel would give better results. Originally animal oils obtained from the blubber of seal and whale were used for this purpose, but the near extinction of the whale has brought to an end the extremely barbaric practice of whaling by all civilised nations. (We must be vigilant nevertheless that whaling does not begin again as apparently there are considerable numbers of so-called gourmets in some countries who wish to eat these noble mammals.) Most quenching oils are now of mineral origin and are obtained during the refining of crude petroleum.

In addition to the rate of heat abstraction such factors as flash point, viscosity and chemical stability are important. A high flash point is necessary to reduce fire risks, whilst high viscosity will lead to loss of oil by 'drag-out', ie oil clinging to the work piece as it is withdrawn from the quenching bath. Atmospheric oxidation and other chemical changes generally led to a thickening of whale oil and the formation of thick scum. On the other hand some mineral oils 'crack' or break down to simpler compounds of lower boiling point which will volatilise in use leaving a thicker, more viscous mixture behind.

Water solutions of synthetic polymers such as polyalkalene glycol are now replacing oils for many quenching operations. Not only do they eliminate fire risks, smoke and unpleasant fume but are generally less expensive. Moreover, having lower viscosities, loss by drag-out is reduced. Less contamination of the work results and degreasing prior to subsequent operations is unnecessary.

12.25. To harden a piece of steel, then, it must be heated to between 30 and 50°C above its upper critical temperature and then quenched in some medium which will produce in it the desired rate of cooling. The medium used will depend upon the composition of the steel and the ulti-

mate properties required. Symmetrically shaped components are best quenched 'end-on', and all components should be agitated in the medium during quenching.

Tempering

12.30 A fully hardened carbon tool steel is relatively brittle, and the presence of stresses set up by quenching make its use, in this condition, inadvisable except in cases where extreme hardness is required. Hence it is customary to re-heat—or 'temper'—the quenched component so that internal stresses will be relieved and brittleness reduced. Medium-carbon constructional steels are also tempered but here the temperatures are somewhat higher so that strength and hardness are sacrificed to some extent in favour of greater toughness and ductility.

12.31 During tempering, which is always carried out *below* the lower critical temperature, martensite tends to transform to the equilibrium structure of ferrite and cementite. The higher the tempering temperature the more closely will the original martensitic structure revert to this ferrite-cementite mixture and so strength and hardness fall progressively, whilst toughness and ductility increase (Fig. 12.5). Thus by choosing the appropri-



Fig. 12.5 The relationship between mechanical properties and tempering temperature for a steel containing 0.5% carbon and 0.7% manganese in the form of a bar 25mm diameter, previously water quenched from 830°C.

ate tempering temperature a wide range of mechanical properties can be achieved in carbon steels.

12.32 The structural changes which occur during the tempering of martensite containing more than 0.3% carbon, take place in three stages:

1st stage At about 100°C, or possibly even lower, the existing martensite begins to transform to another form of martensite, containing only 0.25% carbon, together with very fine particles of a carbide. However, this carbide is not ordinary cementite but one containing rather more carbon and of a formula approximately Fe₅C₂. It is designated ε -carbide. No alteration in the microstructure is apparent under an ordinary optical microscope because the ε -carbide particles are so small, but the electron microscope reveals them as films about 2×10^{-8} m thick. At this stage a slight increase in hardness may occur because of the presence of the finely-dispersed but hard ε -carbide. Brittleness is significantly reduced as quenching stresses disappear in consequence of the transformation. At 100°C the transformation proceeds very slowly but increases in speed up to 200°C.

2nd stage This begins at about 250°C when any 'retained austenite' (12.43) begins to transform to bainite. This will cause the martensite 'needles' to etch a darker colour and formerly this type of structure was known as *troostite*. A further slight increase in hardness may result from the replacement of austenite by much harder bainite.

3rd stage At about 350°C the ε -carbide begins to transform to ordinary cementite and this continues as the temperature rises. In the meantime the remainder of the carbon begins to precipitate from the martensite—also as cementite—and in consequence the martensite structure gradually reverts to one of ordinary BCC ferrite. Above 500°C the cementite particles coalesce into larger rounded globules in the ferrite matrix. This structure was formerly called *sorbite* but both this term and that of troostite are now no longer used by metallurgists who prefer to describe these structures as 'tempered martensite'.

Due to the increased carbide precipitation which occurs as the temperature rises the structure becomes weaker but more ductile, though above 550°C strength falls fairly rapidly with little rise in ductility (Fig. 12.5).

12.33 Tempering can be carried out in a number of ways, but, in all, the temperature needs to be fairly accurately controlled. As the steel is heated, the oxide film which begins to form on a bright, clean surface first assumes a pale-yellow colour and gradually thickens with increase in temperature until it is dark blue. This is a useful guide to the tempering of tools in small workshops where pyrometer-controlled tempering furnaces are not available and is the time-honoured method of heat-treating high-quality hand-made wood-working tools. Table 12.1 shows typical colours obtained on clean surfaces when a variety of components are tempered to suitable temperatures. Such a colour-temperature relationship is only applicable to plain carbon steels. Stainless steels, for example, oxidise less easily, so that the colours obtained will bear no relationship to the temperatures indicated in the table. Moreover, the oxide film colour is only a reliable guide when the component has been progressively raised in

temperature. It does not apply to one which has been maintained at a fixed temperature for some time, since here the oxide film will be thicker and darker in any case. In addition, the human element must also be taken into account, so that, in general, tempering in a pyrometer-controlled furnace is more successful.

12.34 Furnaces used for tempering are usually of the batch type (13.20 —Part II). They employ either a circulating atmosphere or are of the liquid-bath type. Liquids transfer heat more uniformly and have a greater heat capacity, and this ensures an even temperature throughout the furnace. For low temperatures oils are often used, but higher temperatures demand the use of salt baths containing various mixtures of sodium nitrite and potassium nitrate. These baths can be used at about 500°C, but above that temperature either mixtures of chlorides or lead baths are necessary. Another popular furnace, in which the temperature can be varied easily and controlled thermostatically, is the circulating-air type. Here, uniform temperatures up to 650°C can be obtained by using fans to circulate the atmosphere, first over electric heaters, and then through a wire basket holding the charge.

Failing a pyrometer-controlled furnace, temperature-indicating paints and crayons are useful in determining the tempering temperature of small components, provided some method of uniform heating is available. Such indicators do indeed record the *actual temperature reached by the component*, which is more than can be said for a pyrometer controlling a furnace which is in the hands of an unskilled operative.

Temperature (°C)	Colour	Type of component
220	Pale yellow	Scrapers; hack saws; light turning and parting tools
230	Straw	Screwing dies for brass; hammer faces; planing and slotting tools
240	Dark straw	Shear blades; milling cutters; paper cutters; drills; boring cutters and reamers; rock drills
250	Light brown	Penknife blades; taps; metal shears; punches; dies; wood-working tools for hard wood
260	Purplish-brown	Plane blades; stone-cutting tools; punches; reamers; twist drills for wood
270	Purple	Axes; gimlets; augers; surgical tools; press tools
280	Deeper purple	Cold chisels (for steel and cast iron); chisels for wood; plane cutters for soft woods
290	Bright blue	Cold chisels (for wrought iron); screw-drivers
300	Dark blue	Wood saws; springs

Table 12.1 Tempering Colours for Plain-carbon-steel Tools



Plate 12.2 12.2A 0.5% carbon steel, water quenched from 850°C and then tempered at 600°C.

Spheroid carbide in ferrite. \times 250. Etched in 2% nital.

12.2B 0.5% carbon steel, normalised and then annealed for 48 hours at 670°C

The pearlite cementite has become spheroidised. \times 750. Etched in picral-nital. (Courtesy of United Steel Companies Ltd., Rotherham).

12.2C 0.5% carbon steel, water quenched and then tempered for 48 hours at 670°C. Spheroidised carbide has in this case been precipitated from martensite; making the distribution more even than in 12.2B. \times 750. Etched in picral-nital. (Courtesy of United Steel Companies Ltd. Rotherham).

Isothermal Transformations

12.40 As pointed out earlier in this chapter (12.22), the microstructure and properties of a quenched steel are dependent upon the rate of cooling which prevails during quenching. This relationship, between structure and rate of cooling, can be studied for a given steel with the help of a set of isothermal transformation curves which are known as TTT (Time-Temper-ature-Transformation) curves. The TTT curves for a steel of eutectoid composition are shown in Fig. 12.6. They indicate the time necessary for transformation to take place and the structure which will be produced when austenite is *supercooled* to any predetermined transformation temperature.

12.41 Such curves are constructed by taking a large number of similar specimens of the steel in question and heating them to just inside the austenitic range. These specimens are divided into groups each of which



Fig. 12.6 Time-temperature-transformation (TTT). Curves for a plain carbon steel of eutectoid composition.

Martensitic transformation is not complete until approx -50° C. Consequently a trace of 'retained austenite' may be expected in a steel quenched to room temperature.

is quickly transferred to an 'incubation' bath at a different temperature. At predetermined time intervals individual specimens are removed from their baths and quenched in water. The microstructure is then examined to see the extent to which transformation had taken place at the holding temperature. Let us assume, for example, that we have heated a number of specimens of eutectoid steel to just above 723°C and have then guenched them into molten lead at 500°C (Figs. 12.6 and 12.7). Until one second has elapsed transformation has not begun, and if we remove a specimen from the bath in less than a second, and then quench it in water, we shall obtain a completely martensitic structure, proving that at 500°C after one second ('A' on Figs. 12.6 and 12.7) the steel was still completely austenitic. The production of martensite in the viewed structure is entirely due to the final water-quench. If we allow the specimen to remain at 500°C for ten seconds ('B' on Figs. 12.6 and 12.7) and then water-quench it, we shall find that the structure is composed entirely of bainite in feather-shaped patches, showing that after ten seconds at 500°C transformation to bainite was complete. If we quenched a specimen after it had been held at 500°C for five seconds ('C' on Figs. 12.6 and 12.7) we would obtain a mixture of bainite and martensite, showing that, at the holding temperature (500°C), the structure had contained a mixture of bainite and austenite due to the incomplete transformation of the latter. By repeating such treatments at different holding temperatures we are able, by interpreting the resulting microstructures, to construct TTT curves of the type shown in Fig. 12.6.

Because of the very rapid transformations, austenite \rightarrow martensite (and even austenite \rightarrow pearlite) which are involved, it is obvious that considerable practical difficulties arise during laboratory investigations of this type. Since it is impossible to change the temperature from 730°C to 500°C (in the example described above) *in zero time* and again from 500°C to 0°C *in zero time*, we must do the best we can by using very small specimens which are *thin* enough to reach quenching bath temperatures as quickly as possible. For this reason small specimens about the size of a 1p piece are appropriate (Fig. 12.8). These can be attached to suitable 'handles' to facilitate their very rapid transfer between baths. If the incubation bath is of molten metal this will also provide the maximum quenching rate on transfer from the austenitising bath.

12.42 The horizontal line (Fig. 12.6) representing the temperature of 723°C is, of course, the lower critical temperature above which the structure of the eutectoid steel in question consists entirely of stable austenite. Below this line austenite is unstable, and the two approximately C-shaped curves indicate the time necessary for the austenite \rightarrow ferrite + cementite transformation to begin and to be completed following rapid quenching to any predetermined temperature. Transformation is sluggish at temperatures just below the lower critical, but the delay in starting, and the time required for completion, decrease as the temperature falls towards 550°C. In this range the greater the degree of undercooling, the greater is the urge for the austenite to transform, and the rate of transformation reaches a maximum at 550°C. At temperatures just below 723°C, where transformation takes place slowly, the structure formed will be coarse pearlite, since



Fig. 12.7 The extent to which transformation takes place during incubation for different time intervals at a fixed temperature.



Fig. 12.8 The thermal treatment sequence used in the derivation of a set of TTT curves. The thin specimens used are about the diameter of a 1p coin.

there is plenty of time for diffusion to take place. In the region just above 550°C, however, rapid transformation results in the formation of very fine pearlite.

12.43 At temperatures between 550 and 220°C transformation becomes more sluggish as the temperature falls, for, although austenite becomes increasingly unstable, the slower rate of diffusion of carbon atoms in austenite at lower temperatures outstrips the increased urge of the austenite to transform. In this temperature range the transformation product is bainite. The appearance of this phase may vary between a feathery mass of fine cementite and ferrite for bainite formed around 450°C; and dark acicular (needle-shaped) crystals for bainite formed in the region of 250°C.

The horizontal lines at the foot of the diagram are, strictly speaking, not part of the TTT curves, but represent the temperatures at which the formation of martensite will begin (M_s) and end (M_f) during cooling of austenite through this range. It will be noted that the M_f line corresponds approximately to -50° C. Consequently if the steel is quenched in water at room temperature, some 'retained austenite' can be expected in the structure since at room temperature transformation is incomplete. This retained austenite, however, will amount to less than 5% of the austenite which was present at the M_s temperature. In fact, at 110°C (Fig. 12.6) 90% of the austenite will have transformed to martensite.

12.44 These TTT curves indicate structures which are produced by transformations which take place *isothermally*, that is, at a fixed single temperature and specify a given 'incubation' period which must elapse before transformation begins. There is no direct connection between such isothermal transformations and transformations which take place under continuous cooling at a constant rate from 723°C to room temperature. Thus it is not possible to superimpose curves which represent continuous cooling on to a TTT diagram. Modified TTT curves which are related to continuous rates of cooling can, however, be produced. These are similar

in shape to the true TTT curves, but are displaced to the right, as shown in Fig. 12.9. On this diagram are superimposed four curves, A, B, C and D, which represent different rates of cooling.

Curve A represents a rate of cooling of approximately 5°C per second such as might be encountered during normalising. Here transformation will begin at X and can be completed at Y, the final structure being one of fine pearlite. Curve B, on the other hand, represents very rapid cooling at a rate of approximately 400°C per second. This is typical of conditions prevailing during a water-quench, and transformation will not begin until 220°C, when martensite begins to form. The structure will consist of 90% martensite at 110°C and so contain a little retained austenite at room temperature. The lowest rate at which this steel (of eutectoid composition) can be quenched, in order to obtain a structure which is almost wholly martensitic, is represented by curve C (140°C per second). This is called the critical cooling rate for the steel, and if a rate lower than this is used some fine pearlite will be formed. For example, in the case of the curve D, which represents a cooling rate of about 50° C per second, transformation would begin at P with the formation of some fine pearlite. Transformation, however, is interrupted in the region of Q and does not begin again



Fig. 12.9 The relationship between TTT curves and curves representing continuous cooling.

until the M_s line is reached at R, when the remaining austenite begins to transform to martensite. Thus the final structure at room temperature is a mixture of pearlite, martensite and traces of retained austenite.

12.45 The TTT curves illustrated in Fig. 12.6 are those for a steel of eutectoid composition. If the carbon content is either above or below this, the curves will be displaced to the left so that the critical cooling rate necessary to produce a completely martensitic structure will be greater. In order to obtain a structure which is entirely martensitic the steel must be cooled at such a rate that the curve representing its rate of cooling does not cut into the 'nose' of the modified 'transformation begins' curve in the region of 550°C. Obviously, if the steel remains in this temperature range for more than one second, then transformation to pearlite will begin. Hence the need for drastic water-quenches to produce wholly martensitic structures in plain carbon steels.

For a steel containing less than 0.3% carbon the transformation-begins curve has moved so far to the left (Fig. 12.10(i)) that it has become impossible to obtain a wholly martensitic structure however rapidly it is cooled. Large quantities of ferrite will inevitably precipitate when the transformation-begins curve is unavoidably cut in the upper temperature ranges. The resulting structure will be most unsatisfactory since hard martensite will be interspersed with soft ferrite.

Fortunately, the addition of alloying elements has the effect of slowing down transformation rates so that the TTT curves are displaced to the right of the diagram. This means that much slower rates of cooling can be used, in the form of oil- or even air-quenches, and a martensitic structure still obtained. Small amounts of elements, such as nickel, chromium and manganese, are effective in this way and this is one of the most important effects of alloying. In Fig. 12.10(ii) representing the TTT curves for a low alloy steel (covered by BS970:945M38) the 'nose' of the transformation-



Fig. 12.10 (i) TTT curves for a 0.35% carbon steel, indicating that it is impossible to produce a wholly martensitic structure even by drastic water quenching. (ii) TTT curves for a low-alloy steel. Since the curves are displaced significantly to the right it is now possible to obtain a completely martensitic structure by oil quenching.

begins curve is displaced well to the right and in fact a 'double nose' is formed. Even when a continuous-cooling curve (representing an oilquench) is superimposed on this isothermal diagram, it will be seen that there is no transformation until the M_s line is reached and the structures will be wholly martensitic. Since this diagram represents the TTT curves for a hypo-eutectoid steel, ferrite precipitation will begin before pearlite formation as indicated by the broken line.

We will now consider one or two practical applications arising 12.46 from this study of modified isothermal transformation curves. Let us first examine the conditions under which a fairly large body of steel will cool. when quenched. The core will cool less quickly than the outside skin, and since its cooling curve B (Fig. 12.11A) cuts into the nose of the 'transformation-begins' curve, we can expect to find some fine pearlite in the core, whilst the surface layer is entirely martensitic. This feature is usually referred to as the 'mass effect of heat-treatment' (12.50). Even if we are able to cool the component quickly enough to obtain a completely martensitic structure, as indicated in Fig. 12.11B there will be such a considerable time interval CD between both core and surface reaching a martensitic condition that this will lead to quench-cracks being formed. These cracks will be caused by stresses set up as the volume change, associated with the austenite \rightarrow martensite transformation, progresses from the skin to the core of the section. Supposing, however, we cool the steel under conditions of the kind indicated in Fig. 12.11c. Here the steel is quenched into a bath at temperature E and left there long enough to permit it to reach a uniform temperature throughout. It is then removed from the bath and allowed to cool so that martensite will begin to form at F. The net result is that, by allowing the core to attain the same temperature as the surface whilst in the bath at temperature E, we have prevented a big temperature gradient from being set up between the surface and the core of the specimen at the moment when martensite begins to form. The final air-cooling will not be rapid enough to allow a large temperature gradient to be set up, and both core and surface will become martensitic at approximately the same time, thus minimising the tendency towards quench-cracking. The success of this treatment, which is known either as martempering or marquenching lies in cooling the steel quickly enough past the 'nose' of the modified transformation-begins curve. Once safely past that point, relatively slow cooling will precipitate martensite. If we should cut into the 'nose' fine pearlite will begin to form. It is obvious that for plain carbon steels this type of treatment will have its limitations and will be applicable to components of thin section only, otherwise the temperature gradient set up within the section would be too great to prevent pearlite formation in the core.

With suitable steels an *ausforming* operation can be combined with martempering. Whilst the steel is at the holding temperature the austenitic structure is heavily worked—up to 90% sectional reduction in area being applied. The density of dislocations is increased by working and these 'jammed' dislocations are retained because the relatively low temperature does not permit recrystallisation. The steel is then allowed to cool and so transform to martensite which will *retain the high density of* dislocations,



Fig. 12.11 (A) and (B) illustrate the effects of mass during normal quenching. (C) and (D) show how these effects may be largely overcome in martempering and austempering.

so that its strength is further increased. Moreover, the grain is fine due to the low forming temperature, whilst a fine dispersion of carbides, presumably deposited during forming, contributes some degree of particle hardening. Some low-alloy steels, having a deep 'bay' between the pearlite and bainite 'noses' of the TTT curve (Fig. 12.10(ii)) can be treated in this way and will develop tensile strengths in the region of 3000 N/mm².

12.47 Isothermal transformation offers a method by which we can obtain a tempered type of structure without the preliminary drastic water-

quench. Such a treatment, known as *austempering*, is illustrated in Fig. 12.11D. Here the steel is quenched into a bath at a temperature above that at which martensite can be formed and allowed to remain there long enough for transformation to be complete at G. Since transformation to bainite is complete at G, the steel can be cooled to room temperature at any desired rate, but air-cooling is preferable. We have thus succeeded in obtaining a structure which is similar in properties, though not in microstructure, to that of tempered martensite which is obtained by quenching and tempering. The drastic water-quench from above the upper critical temperature, however, has been avoided. Austempering is therefore a process of considerable importance when heat-treating components of intricate section. Such components might distort or crack if they were heat-treated by the more conventional methods of quenching and tempering.

Although most of our modern knowledge of the basic mechanisms of isothermal transformation phenomena stem from the research carried out by Bain and Davenport from the late nineteen-thirties onwards, industrial processes based on what we now understand as isothermal transformation in steel, predated the work of Bain and his associates by many years. Such a process was patenting, a treatment employed in the manufacture of high-tensile steel wire. This wire, containing between 0.35 and 0.95% carbon, is first austenitised at temperatures up to 970°C (depending of course on the carbon content) by passing it through heated tubes in order to minimise decarburisation. From the austenitising furnace it passes into a bath of molten lead at 400-500°C where it remains for a period long enough for a structure of feathery bainite to develop. This structure is sufficiently ductile to permit cold-working of up to 90% sectional reduction in area, without annealing being necessary. Tensile strengths of over 2000 N/mm² are attainable as a result of the combination of heat-treatment and cold-work. Wire for piano, guitar and other musical instrument strings can be made in this way as well as high-tensile wire ropes.

The heat-treatment of spades, forks and other thin-section garden tools by a method akin to austempering also pre-dated Bain by many years. There are many such instances where industrial processes were developed by trial and error over many years, long before the underlying scientific explanation was forthcoming. Nevertheless modern austempering processes are finding increasing use. For example, even the steel toecaps of industrial boots are heat-treated in this manner.

12.48 Finally, mention must be made of *isothermal annealing*. In this process the steel is heated into the austenitic range and then allowed to transform as completely as possible in the pearlitic range. The object of such treatment is generally to soften the steel sufficiently for subsequent cold-forming or machining operations.

The nature of the pearlite formed during transformation is influenced by the initial austenitising temperature. An austenitising temperature which is little above the upper critical for the steel promotes the formation of spheroidal pearlitic cementite during isothermal annealing, whilst a higher austenitising temperature favours the formation of lamellar pearlitic cementite. The pearlite structure is also influenced by the temperature at which isothermal transformation takes place, as would be expected. Transformation just below the lower critical temperature leads to the formation of spheroidal pearlitic cementite since precipitation is slow, whilst at lower temperatures transformation rates are higher and lamellar cementite tends to form. A structure containing spheroidal cementite is generally preferred for lathe work and cold-forming operations, whilst one with lamellar cementite is often used where milling or drilling are involved. It is claimed that isothermal annealing gives more uniform properties than does an ordinary annealing process.

Hardenability and Ruling Section

12.50 Brief mention of the workshop term 'mass effect' in connection with the heat treatment of steel has already been made (12.46). If a piece of carbon steel is of heavy cross-section it will probably be impossible to cool it quickly enough to produce a uniformly martensitic structure throughout, even by the most severe quenching. Such a section would be likely to have a soft un-hardened core due to its relatively slow cooling rate (Fig. 12.11A), whilst a piece of steel of thin section quenched in a similar way would be martensitic throughout (Fig. 12.12). This difficulty may be remedied to some extent by adding alloying elements to the steel. These reduce the critical rates of austenite transformation and make it possible to get a martensitic structure throughout quite thick sections even when the less drastic oil- or air-quenching processes are used.

12.51 This is one of the most important functions of alloying, but to avoid the misuse of steels due to lack of understanding of their properties it became necessary for manufacturers to specify the maximum diameter or *ruling section* of a bar, up to which the stated mechanical properties would apply following heat-treatment. That is, some users failed to appreciate that these low-alloy steels still had a critical cooling rate even though it was much lower than that of a plain carbon steel of similar carbon



Fig. 12.12 The effects of mass produced in the structure on quenching. The heavier sections will cool too slowly to be entirely martensitic.

content. If the ruling section is exceeded then the properties across the section will not be uniform since hardening of the core will not be complete.

12.52 The Jominy end-quench test is of great practical use in determining the hardenability of steel. Here a standard test piece is made (Fig. 12.13A) and heated up to its austenitic state. It is then dropped into position in a frame, as shown in Fig. 12.13B, and quenched at its end only, by means of a pre-set standard jet of water at 25°C. Thus different rates of cooling are obtained along the length of the bar. After the cooling, a 'flat', 0.4 mm deep, is ground along the side of the bar and hardness determinations made every millimetre along the length from the quenched end. The results are then plotted as in Fig. 12.14.

These curves show that the depth of hardening of a nickel-chromium steel is greater than that of a plain carbon steel of similar carbon content, whilst the depth of hardening of a chromium-molybdenum steel is greater than that of the nickel-chromium steel.

With modifications, the results of the Jominy test can be used as a basis in estimating the 'ruling section' of a particular steel. There is no simple mathematical relationship between the two, however, and it is often more satisfactory to find by trial and error how a particular section will harden, after a preliminary Jominy test has been conducted.



Fig. 12.13 The Jominy end-quench test. (A) The standard form of test piece used. (B) A simple type of apparatus for use in the test.



Fig. 12.14 The relative depth of hardening of three different steels as indicated by the Jominy test.

British Standard Specifications for Carbon Steels

12.60 British Standard Specifications for wrought steels containing carbon and manganese (including some free-cutting steels) are dealt with in Part 1 of BS 970. The original 'En' designation which was used to identify individual steels within BS 970 was abandoned as long ago as 1972 but 'old habits die hard' and one still finds these En numbers being quoted. Some claim that the new numbers are 'too cumbersome' but in fact they are very useful in that the six digit designation used indicates both carbon and manganese contents as well as other important information relating to supply of the steel.

The series 000 to 199 has been allocated for the first three digits in respect of steels containing carbon and manganese. These digits represent one hundred times the mean manganese content. The letter 'A' or 'M' has been introduced as the fourth digit to indicate whether the steel is to be supplied to analysis (A) or mechanical property (M) requirements. When hardenability requirements are included in the specification the letter 'H' is used as the fourth digit. Finally the fifth and sixth digits represent one hundred times the mean carbon content. (Since only one steel in this Part of the standard contains 1.00% carbon, the two digits '99' are used to describe it so as to avoid the use of a three-digit suffix throughout the standard.)

As an example BS 970:060A52 describes a steel containing 0.50-0.55% carbon; 0.50-0.70% manganese supplied according to analysis (A) requirements.

The free-cutting carbon-manganese steels (6.63) are prefixed by the

		_	Composition (%)			_	Typical Mechanical properties				
Туре		BS970:	с	Mn	Heat-treatment	- Ruling Section (mm)	Tensile strength (N/mm ²)	Yield strength (N/mm ²)	Elongation (%)	Izod (J)	Hardness (Brinell)
Low carbon		015A03 050A12 060A17	0.03 0.12 0.17	0.15 0.50 0.60	Not heat-treatable—used where ductility and formability are required						
Medium carbon—mainly constructional steels.	'20' carbon	070M20	0.20	0.70	Normalised at 880–910°C. Water-quenched 880–910°C; tempered 550–660°C.	150 20	430 620	215 355	21 20	41	150 180
	'30' carbon	080M30	0.30	0.80	Normalised at 860-890°C. Water quenched 860-890°C; tempered 550-660°C.	150 20	495 695	250 415	20 16		165 205
	'40' carbon	080M40	0.40	0.80	Normalised at 830-860°C. Water quenched 830-860°C; tempered 550-660°C.	150 20	540 770	280 465	16 16	20 34	180 230
	'50' carbon	080M50	0.50	0.80	Normalised at 810–840°C. Oil quenched 810–840°C; tempered 550–660°C.	150 12.7	620 925	310 570	14 12		205 275
	'55' carbon	070M55	0.55	0.70	Normalised at 810–840°C. Oil quenched 810–840°C; tempered 550–660°C.	63.5 20	695 930	355 575	12 12		230 275
High carbon		080A62 080A72 080A83 060A86 060A99	0.62 0.72 0.83 0.86 1.00	0.80 0.80 0.80 0.60 0.60	Heat-treated to give required combination of hardness and toughness—generally used as tool steels		-				-

Table 12.2 Representative Compositions and Properties for Some Steels Covered in BS970: Part I

(Compositions for some of the carbon-manganese free-cutting steels will be found in Table 6.2) For typical uses of steels in the above Table refer to Table 7.1 and Fig. 7.7.

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series 200 to 240 where the second and third digits are roughly one hundred times the sulphur content. Thus BS 970:216M28 describes a freecutting steel containing 0.24-0.32% carbon and 0.12-0.20% sulphur supplied according to mechanical property (M) requirements.

A few representative carbon steels along with their BS 970 designations, mean compositions, heat-treatments and typical mechanical properties are shown in Table 12.2.

Exercises

- 1. Examine Table 12.2 and explain why the following steels (covered in BS 970): 070M20; 080M30; 080M40 and 080M50 have progressively lower normalising and quenching temperatures. (12.20)
- 2. Outline, using sketches, the theory which seeks to explain the development of the martensite structure when a carbon steel is water quenched. (12.22)
- 3. An annealed 0.4% C steel bar is cold-worked and placed with one end in a furnace at 900°C whilst the other end is maintained at room temperature. After a few hours the bar is quenched in cold water. Describe the structures you would expect to find along the length of the bar. (11.51 and 12.21)
- 4. Sketch and label the 'steel part' of the iron-carbon thermal equilibrium diagram. With reference to the diagram describe the structural changes which occur when a cast 0.5%C steel is:
 - (i) slowly heated to 900°C;
 - (ii) slowly cooled from 900°C;
 - (iii) quenched from 900°C;
 - (iv) quenched from 900°C and reheated to 650°C.
 - Sketch each microstructure, including that of the steel in the cast condition, and comment *qualitatively* on the mechanical properties you would expect as a result of each treatment. (11.53, 12.21 and 12.32)
- 5. Both annealing and tempering are processes used to soften steel. Outline the conditions when these treatments would be used, and indicate any difficulties that may be encountered in practice. (11.50 and 12.30)
- 6. Four thin pieces of the same 0.8% C rolled-steel rod are heat-treated differently as follows:
 - (i) heated to 680°C for twenty-four hours and cooled in still air;
 - (ii) water-quenched from 750°C;
 - (iii) heated to 730°C, quenched into molten lead at 400°C, allowed to remain there for five minutes and then cooled;
 - (iv) heated to 1200°C and cooled in still air. Sketch the type of microstructure produced in each specimen and explain the mechanism of its formation. (11.52; 12.20; 12.46 and 11.21)
- 7. Using diagrammatic TTT curves, explain the reasons for the addition of alloying elements to steels to overcome the limitations of carbon steels in heat-treatment. (12.50)
- 8. Explain fully what is meant by the 'ruling section' of a steel and discuss its significance in the choice of steels for engineering design. Outline *one* experimental procedure which is helpful in assessing the ruling section of a steel. (12.50 and 12.52)
- 9. The following results were obtained during Jominy end-quench tests carried out under similar conditions on two steels of similar carbon content;

Dist. from quenched end (mm)	2	5	7	10	15	20	25	30	40	50
Hardness (H _V), steel. 1.	604	512	321	290	254	250	246	241	236	235
Hardness (H _V), steel. 2.	605	590	585	580	577	556	549	535	500	437

Draw, on squared paper, the Hardness/Distance-from-quenched-end curve for each steel and comment on them.

What would you deduce regarding the possible compositions of each steel? (12.52)

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