

9 From Bainite to Austenite

Many commercial processes cause the steel to revert into the austenitic condition. The transformation of low-temperature ferrite into high-temperature austenite differs from the case where the latter transforms during cooling. Transformation during cooling follows a C curve kinetics in which the overall transformation rate goes through a maximum as a function of the undercooling below the equilibrium temperature. This is because diffusion coefficients decrease but the driving force increases as the temperature is reduced. By contrast, both the diffusivity and driving force for austenite formation increase as a ferritic microstructure is superheated. The rate of transformation increases indefinitely as the temperature is raised, Fig. 9.1.

This kinetic behaviour has several interesting consequences. It is commonly observed that reconstructive transformations can be suppressed by cooling rapidly to a low temperature where the lack of atomic mobility prevents further transformation. Austenite can therefore be retained by rapid cooling to ambient temperature, even though it is not thermodynamically stable. It should be impossible to similarly retain ferrite to high temperatures in the γ -phase field during a heating experiment, since atoms become more mobile at higher temperatures.

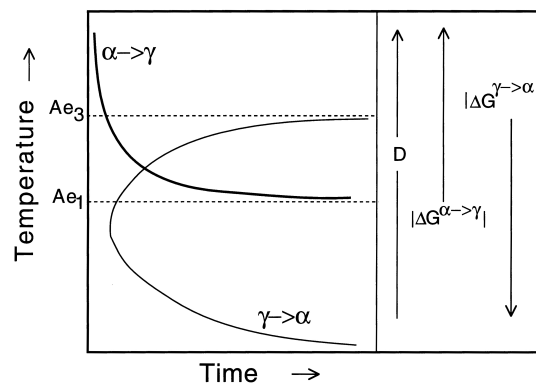


Fig. 9.1 The TTT curves for the $\gamma \rightarrow \alpha$ reaction, and for the reverse $\alpha \rightarrow \gamma$ transformation.

Austenite is not always retained when quenched from an elevated temperature. It may transform by a mechanism which does not require diffusion (martensitic). When the $\alpha \rightarrow \gamma$ transformation occurs during heating, the temperatures involved are usually high enough to permit the rapid reconstructive transformation. It is therefore rare for austenite to grow by a martensitic mechanism.

In compendiums of time–temperature–transformation diagrams, the kinetics of austenite decomposition are presented as a function of the chemical composition and the austenite grain size. The number of variables to be considered when presenting similar data for transformation to austenite is much larger: the initial microstructure can vary widely. The sophistication with which it is necessary to specify the starting microstructure remains to be determined, but factors such as particle size, the distribution and composition of individual phases, homogeneity of the microstructure, the presence of nonmetallic inclusions, etc. should all be important.

There are two particular examples where a detailed knowledge of austenitisation could be exploited to considerable advantage. During fusion welding, an optimum microstructure is required immediately after deposition from the liquid state. The luxury of homogenisation or other thermomechanical treatments is simply not available or practical. The welding process dissipates heat into the surrounding metal, with regions in the immediate proximity of the fusion surface being heated to temperatures high enough to cause austenitisation. Another example where austenitisation theory could be usefully applied is in the development of new wrought steels (Fe–Ni–Ti), where attempts are being made to utilise microstructures which have been partially austenitised.

There clearly is work to be done on all aspects of the formation of austenite, but the discussion in this chapter is confined to the austenitisation of bainitic microstructures.

9.1 Heating a Mixture of Austenite and Upper Bainitic Ferrite

When an iron–carbon alloy is heated to a temperature within the $\alpha + \gamma$ phase field until equilibrium is established, a small rise or fall in temperature leads to the growth or dissolution respectively, of the austenite until the volume fractions once again satisfy the lever rule (Tsuzaki *et al.*, 1988). The transformation of austenite into allotriomorphic ferrite is in this sense reversible, and exhibits little or no hysteresis. A much larger hysteresis occurs for the martensite to austenite transformation because the martensite tempers during heating and because its growth involves dissipation in the form of irreversible plastic

deformation. A substantial hysteresis effect is found experimentally when a mixture of bainitic ferrite and austenite is heated (Fig. 9.2).

If carbides precipitate during the bainite reaction then the final microstructure is unlikely to contain retained austenite. The sample must then be heated into the $\alpha + \gamma$ phase field before austenite can nucleate first and then grow. Of course, if austenite exists in the starting microstructure, and if it remains stable during heating, then it can begin growth as soon as the free energy change becomes negative. For bainite, this may nevertheless require a substantial superheat because the transformation remains incomplete, i.e. it stops when $x_\gamma \simeq x_{T_0}$ rather than when $x_\gamma = x_{Ae3}$. The temperature therefore has to be raised until the carbon concentration of the residual austenite becomes equal to that given by the $Ae3$ phase boundary before the austenite can grow (Fig. 9.2).

These concepts can be revealed in steels which transform to bainite without the precipitation of carbides (Yang and Bhadeshia, 1989b). We have argued above that when a mixture of bainitic ferrite and carbon-enriched austenite is heated sufficiently rapidly to a high enough temperature, the existing austenite can grow without there being a need for nucleation (Fig. 9.3).[†]

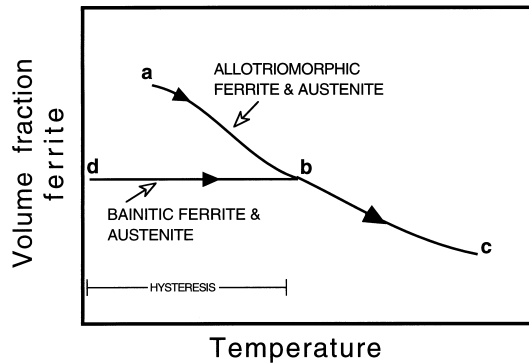


Fig. 9.2 The growth of austenite when mixtures of ferrite and austenite are heated. An equilibrium mixture of allotriomorphic ferrite and austenite begins to transform immediately the temperature is raised, whereas a large superheat is needed when a mixture of bainitic ferrite and austenite is heated.

[†]This does not preclude new nuclei of austenite; thus, Kessler and Pitsch (1965) found that new regions of austenite nucleated when a mixture of martensite and retained austenite was heated. Whether new nuclei form in microstructures which already contain retained austenite must depend on the superheat since nucleation is most difficult at low driving forces.

Bainite in Steels

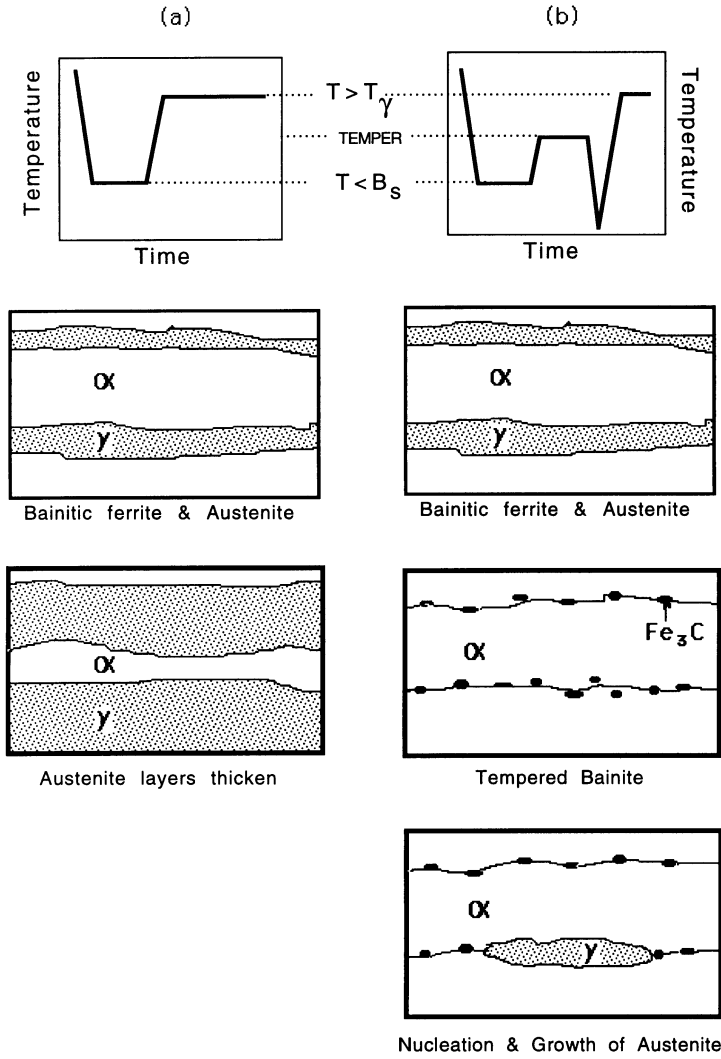


Fig. 9.3 (a) A mixture of bainitic ferrite and austenite is generated by isothermal transformation at a temperature below B_s ; this microstructure is then reheated to an elevated temperature to permit the austenite to grow. (b) A tempering heat-treatment eliminates austenite. It is then necessary to nucleate austenite before it can grow.

Experiments like these have shown that the austenite in low-alloy steels grows by a reconstructive process at all but the fastest of heating rates. A large difference is found between the B_s and A_s (austenite-start) temperatures. The austenite only begins to grow when the Ae_3 temperature of the *residual*

austenite is reached. Its fraction then increases from that at the A_S temperature, to complete transformation at the austenite-finish (A_F) temperature, which is the Ae_3 temperature of the alloy as a whole.

The observed austenitisation behaviour can be understood as follows (Yang and Bhadeshia, 1987, 1988). When carbide precipitation is avoided, bainite stops to form when the $x_\gamma = x_{T'_0}$ (Fig. 9.4). We shall designate this value of x_γ as the initial value x_γ^I obtained by transformation at the temperature T_i , i.e.

$$x_\gamma^I = x_{T'_0}\{T_i\} \quad (9.1)$$

as indicated by the point **a** in Fig. 9.4. Furthermore, we note that:

$$x_\gamma^I \ll x_{Ae3}\{T_i\} \quad (9.2)$$

where $x_{Ae3}\{T_i\}$ is marked as point **b** in Fig. 9.4. Bainite does not form if $x_\gamma > x_{T'_0}$ but at that point, x_γ is far less than the equilibrium or paraequilibrium concentration. Another way of stating this is to say that the fraction of austenite at T_i is greater than expected from equilibrium, so there is no tendency for the austenite to grow. This remains the case until the temperature is high enough to satisfy the equation:

$$x_\gamma^I = x_{Ae3}\{A_S\} \quad (9.3)$$

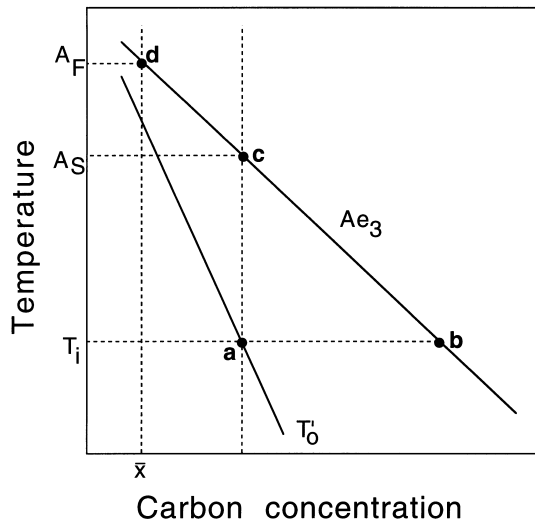


Fig. 9.4 Schematic phase diagram illustrating the theory for austenite growth when the initial microstructure is a mixture of bainitic ferrite and carbon-enriched residual austenite.

The austenite only begins to grow at A_S corresponding to the point **c** in Fig. 9.4. The required superheat $A_S - T_i$ is a direct effect of the incomplete reaction phenomenon.

The theory predicts that when $T_\gamma > A_S$, the $\alpha_b \rightarrow \gamma$ transformation should stop when

$$x_\gamma = x_{Ae3}\{T_\gamma\}. \quad (9.4)$$

Neglecting differences in the densities of austenite and ferrite, and assuming that $x_\alpha = 0$, the equilibrium volume fraction of austenite at T_γ is then given by:

$$V_\gamma\{T_\gamma\} = \bar{x}/x_{Ae3}\{T_\gamma\}. \quad (9.5)$$

The alloy becomes fully austenitic when $x_{Ae3}\{T_\gamma\} = \bar{x}$ (point **d**, Fig. 9.4). The corresponding temperature is designated A_F so for all $T_\gamma > A_F$, the alloy transforms completely to austenite.

9.1.1 One-Dimensional Growth from a Mixture of Austenite and Bainitic Ferrite

As discussed earlier, austenite need not nucleate when mixture of ferrite and austenite is heated. Transformation can occur by the thickening of the austenite films between the bainite plates. This effectively is one-dimensional growth, which we shall assume is diffusion-controlled. The redistribution of carbon must occur inside the austenite during its growth, assuming that its solubility in ferrite can be neglected. Microanalysis has shown that substitutional solutes may partition during austenite formation (Yang and Bhadeshia, 1987, 1988, 1989b). The extent of partitioning decreases as the transformation temperature T_γ , and hence the driving force, increases. It could be assumed that local equilibrium exists at the interface for low T_γ , with a tendency towards zero bulk partitioning, (i.e., negligible-partitioning local equilibrium or paraequilibrium) as $T_\gamma \gg Ae_3$. This makes a full analysis difficult because there are many possibilities between the states of local equilibrium and paraequilibrium.

If local equilibrium is achieved at the interface, the growth rate assuming carbon diffusion-control may give an estimate of the factors influencing the kinetics of transformation. This is the basis of the model presented below, which assumes that substitutional solute gradients do not affect the carbon (Kirkaldy, 1958). There is a further assumption that the tie-line of the equilibrium phase diagram, which determines the interface compositions, passes through the average composition of the alloy, which is unlikely in concentrated alloys. Any effects of soft impingement are also neglected.

One-dimensional diffusion-controlled growth leads to a parabolic thickening of films of austenite. The increase in the half-thickness q of the film is given by:

$$dq = \frac{1}{2} \alpha_1 t^{-\frac{1}{2}} dt \quad (9.6)$$

where α_1 is the one-dimensional parabolic thickening rate constant. The geometry assumed for the thickening of austenite layers is based on the shape of the bainite or acicular ferrite plates which bound the layers. If c is the largest dimension of such a plate, idealised as a rectangular parallelepiped with sides of length a , b and c , with $c = b \gg a$, then when both of the sides of a ferrite plate are penetrated by the growing austenite, the total area of the γ/α interface which advances into the plate of ferrite is $2c^2$. This reduces the thickness of the plate by $\Delta a/2$ from either side. If the minimum detectable change in volume fraction is ΔV_v , then it follows that:

$$\Delta V_v = 2N_v c^2 \int_0^{\Delta a/2} dq \quad (9.7)$$

where N_v is the initial number of particles of austenite per unit volume, and Δa_m is the minimum detectable increase in thickness. It follows that

$$\Delta V_v = 2N_v c^2 \int_0^\tau \frac{1}{2} \alpha_1 t^{-\frac{1}{2}} dt \quad (9.8)$$

where τ is the time taken to achieve the minimum detectable degree of transformation. After integration, equation 9.8 becomes:

$$\Delta V_v = 2\alpha_1 N_v c^2 \tau^{\frac{1}{2}} \quad \text{or} \quad \tau = \left(\frac{\Delta V_v}{2\alpha_1 N_v c^2} \right)^2 \quad (9.9)$$

$$\text{Since} \quad 2N_v c^2 = S_v = 2/\bar{L}_3 \quad \text{it follows that} \quad \tau = \left(\frac{\Delta V_v}{\alpha_1 S_v} \right)^2 \quad (9.10)$$

where S_v is surface area of γ/α boundary per unit volume, and $1/\bar{L}_3$ is the mean number of intercepts of γ/α boundary per unit length of test line (DeHoff and Rhines, 1968). It is clear from equation 9.10, that τ is dependent not only on α_1 but also on the surface area of γ/α interface per unit volume S_v for a specific amount of austenitisation.

For the same initial microstructure and a fixed degree of transformation, τ should decrease rapidly T_γ . The microstructure affects τ via S_v . This explains experimental observations of the different rates at which mixtures of $(\alpha_b + \gamma)$ and $(\alpha_a + \gamma)$ austenitise (Fig. 9.5a). The distribution of plates in an acicular ferrite microstructure is such that S_v is smaller than in bainite, making the transformation to austenite relatively slow. It has also been verified experimentally that τ is proportional to α_1^{-2} (Fig. 9.5c).

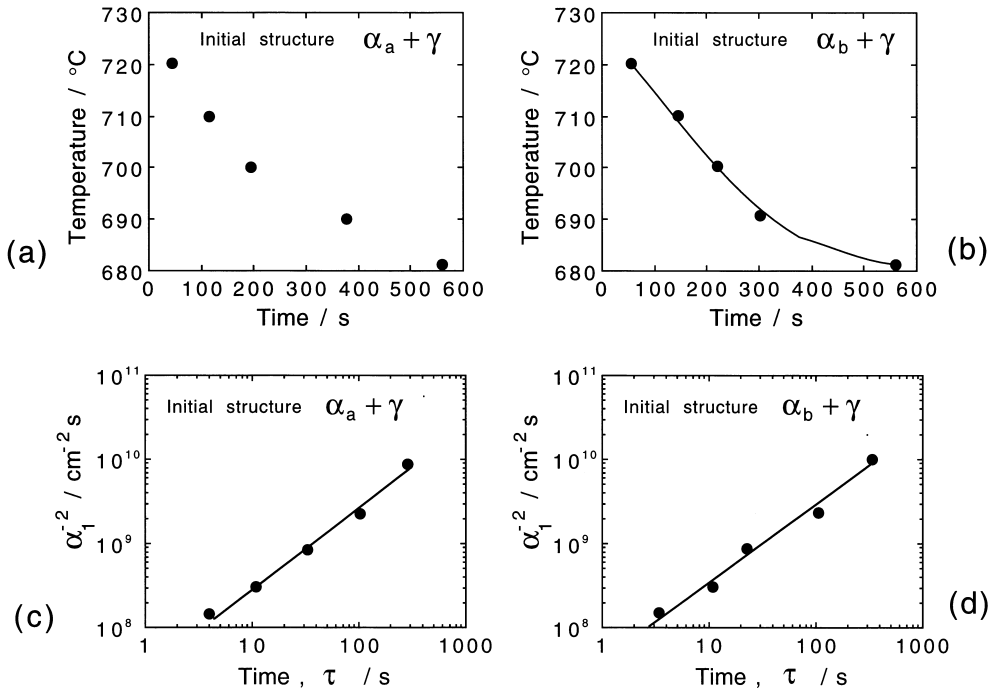


Fig. 9.5 (a,b) TTT diagrams for the growth of austenite from equivalent mixtures of acicular ferrite/austenite, and bainitic ferrite/austenite. (c,d) Linear relationship between the time τ taken for a constant volume fraction of austenite growth, versus α_1^{-2} , where α_1 is the one-dimensional parabolic thickening rate constant for austenite growth (Yang and Bhadeshia, 1989b).

9.1.2 Estimation of the Parabolic Thickening Rate Constant

The parabolic rate constant α_1 can be calculated using existing theory for the $\gamma \rightarrow \alpha$ transformation (Zener, 1949; Dubé, 1948; Bhadeshia, 1985b). Fig. 9.6 shows the carbon concentration profiles in α and γ before and during austenite growth. The austenite must become more dilute in carbon as it grows, the rate of interface motion being determined by the diffusion of carbon in the austenite behind the interface. In Fig. 9.6, x_γ^I is the initial carbon concentration in the austenite, given by $x_\gamma^I = x_{T_0}^I$. The carbon concentration of γ at the γ/α interface during austenitisation is $x_\gamma^{I\alpha}$ and that of austenite far away from the interface is assumed to remain constant at x_γ^I , as is x_α^γ . The coordinate z is normal to the γ/α interface.

The flux of carbon in the austenite, towards the γ/α interface, at the position of interface is from Fick's first law given by:

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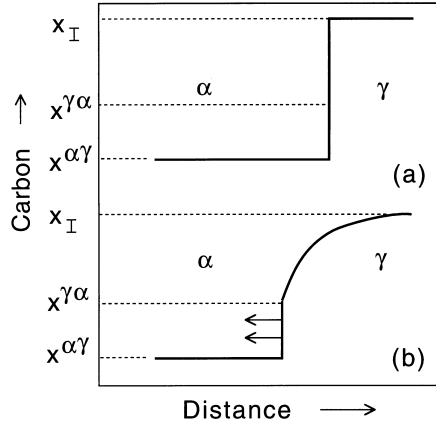


Fig. 9.6 The distribution of carbon, (a) before austenitisation from a mixture of bainitic ferrite and austenite, and (b) during the growth of austenite.

$$J = -D\{x^{\gamma\alpha}\} \left. \frac{\partial x}{\partial z} \right|_{z=Z} \quad (9.11)$$

The rate at which the carbon concentration of austenite is diluted is:

$$R_d = V_d(x_\gamma^I - x^{\alpha\gamma}) \quad (9.12)$$

where V_d is the velocity of interface (the diffusion-field velocity). Given that

$$Z = \alpha_1 t^{\frac{1}{2}},$$

it follows that:

$$V_d = \frac{dZ}{dt} = \frac{1}{2} \alpha_1 t^{-\frac{1}{2}} \quad (9.13)$$

Consequently, the rate at which the carbon concentration of austenite is diluted is given by:

$$R_d = \frac{1}{2} \alpha_1 t^{-\frac{1}{2}} (x_\gamma^I - x^{\alpha\gamma}) \quad (9.14)$$

Making the approximation that the concentration dependence of the diffusion coefficient of carbon can be represented by its weighted average diffusivity \bar{D} , conservation of mass at the interface requires that:

$$\frac{1}{2} \alpha_1 t^{-\frac{1}{2}} (x_\gamma^I - x^{\alpha\gamma}) = -\bar{D} \left(\frac{\partial x}{\partial Z} \right) \Big|_{z=Z} \quad (9.15)$$

This equation expresses the condition that as the austenite becomes dilute as its size increases, the change in concentration at the interface is compensated by a diffusion flux of carbon towards the γ/α interface. The differential equation for the matrix is:

$$\frac{\partial x}{\partial t} = \frac{\partial(\bar{D}\partial x/\partial Z)}{\partial Z} \quad (9.16)$$

subject to the boundary conditions $x = x^{\gamma\alpha}$ at $z = Z\{t\}$, and $x = x_{\gamma}^I$ at $t = 0$. Its solution leads to the following relationship from which α_1 can be determined (Zener, 1949; Dubé, 1948; Atkinson, 1967):

$$f_1 = \frac{x_{\gamma}^I - x^{\gamma\alpha}}{x_{\gamma}^I - x^{\alpha\gamma}} = H_1\{\bar{D}\} \quad (9.17)$$

where

$$H_1\{\bar{D}\} = \left(\frac{\pi}{4\bar{D}}\right)^{\frac{1}{2}} \alpha_1 \operatorname{erfc}\left\{\frac{\alpha_1}{2\bar{D}^{\frac{1}{2}}}\right\} \exp\left\{\frac{\alpha_1^2}{4\bar{D}}\right\} \quad (9.18)$$

9.2 Anisothermal Transformation

Heat treatments are rarely isothermal in commercial practice. A continuous heating curve can be expressed as a series of small isothermal steps i , each occurring at a successively higher temperature, with a time interval t_i associated with each step. With Scheil's rule, a specified increment of transformation is achieved during continuous heating when the sum of all the ratios of time steps to incubation periods equals unity:

$$\sum_{i=1}^n \frac{t_i}{\tau_i} = 1 \quad (9.19)$$

where τ_i is the time required to reach the specified fraction of transformation at the temperature T_i . This additivity rule assumes that the reaction is *isokinetic*, meaning that the fraction transformed is dependent only on the time and on a single function of temperature. This is unlikely to be true except in special cases where for example, nucleation is stifled by site saturation.

9.3 Heating a Mixture of Cementite and Bainitic Ferrite

Austenite grows with an equiaxed shape when the initial microstructure is pearlite, but in the form of layers between plates of ferrite when the initial microstructure is bainite or martensite (Nehrenberg, 1950). However, there are contradictory observations showing austenite nucleating at the prior austenite

grain boundaries from initial microstructures which are bainitic or martensitic (Law and Edmonds, 1980). What is clear, is that when the austenite forms as layers between the ferrite plates, the steel exhibits a *memory effect*. In this, the original austenite grain structure is regenerated when the transformation to austenite is completed, both with respect to shape and crystallography (Sadovskii, 1956; Kimmins and Gooch, 1983). Naturally, the austenite grain structure cannot be refined by repeated thermal cycling of the sample into the austenite phase field when the memory effect operates.

The memory arises from films of retained austenite in the starting bainitic or martensitic microstructure (Kimmins and Gooch, 1983). The films grow and coalesce to regenerate the original austenite grain structure. The memory effect vanishes if the initial microstructure is first annealed to eliminate any retained austenite. Allotriomorphs of austenite are nucleated when these annealed samples are heated into the austenite phase field (Wada and Eldis, 1982).

Retained austenite can decompose during slow heating to the austenitisation temperature, thus destroying the memory effect. Very rapid heating can also eliminate the memory by inducing the nucleation of new austenite grains (Kimmins and Gooch, 1983). The memory is enhanced if the steel contains impurities such as arsenic, phosphorus or tin, which segregate to the prior austenite grain boundaries (Kimmins and Gooch, 1983). The segregation reduces the grain boundary energy, making them less likely as heterogeneous nucleation sites.

9.4 Irradiation-Induced Rapid Heating

Surface layers of a steel containing ferrite and pearlite, when irradiated with high-energy electrons, transform into austenite. The effective heating and cooling rates are large because the irradiation effect is confined to a thin surface layer. As a consequence, the carbon concentration in the austenite which grows from pearlite is found to be much larger than in the remainder of the austenite because the rapid thermal cycle does not permit homogenisation over the scale of the microstructure. During cooling, martensite forms in the high carbon regions and bainite in the regions which were originally ferrite (Choi *et al.*, 1999). Rapid inductive heating also leads to an inhomogeneous distribution of carbon in the austenite, so that cooling produces mixed microstructures of ferrite and bainite (Weidig *et al.*, 1999).

9.5 Summary

Microstructures containing a mixture of bainitic ferrite and austenite when heated do not require the nucleation of new austenite. Nevertheless, they have to be superheated over a large temperature range before the austenite

begins to grow. This is because the bainite reaction stops before equilibrium is achieved so that the fraction of austenite in the initial microstructure is greater than required by equilibrium.

The nucleation of austenite is necessary when the original microstructure does not contain retained austenite. In this case, nucleation generally occurs preferentially at the prior austenite grain boundaries rather than between the ferrite plates. When a microstructure containing retained austenite is heated rapidly, the austenite grows and regenerates the original austenite grain structure, giving the so-called memory effect. This memory can only be destroyed by eliminating the retained austenite either by slow heating or by suitably tempering the initial microstructure (Fig. 9.7).

For any reasonable heating rate, the austenite grows by a reconstructive mechanism with the diffusion of substitutional solutes. The extent of solute partitioning decreases with the superheat above the equilibrium transformation temperature, but cannot as yet be predicted theoretically.

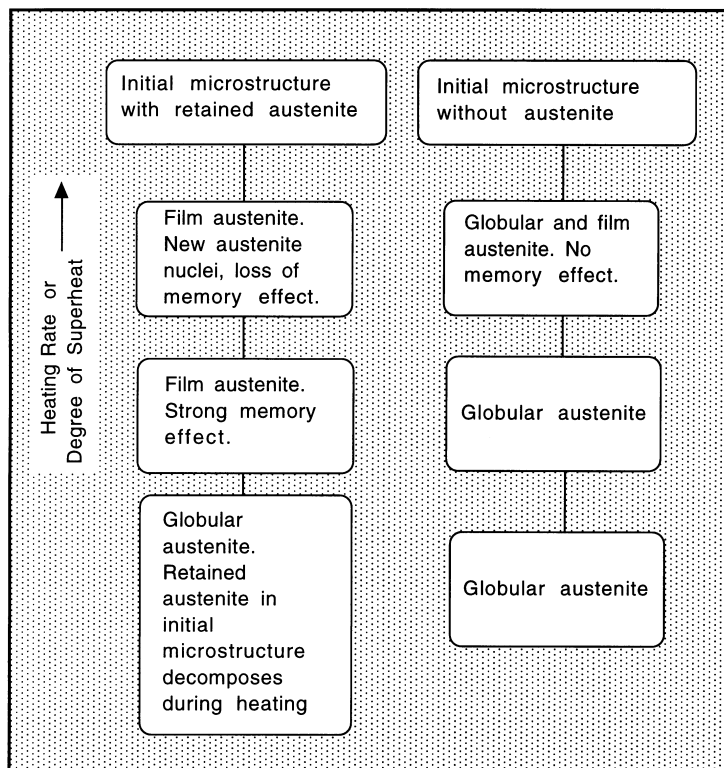


Fig. 9.7 The effects of heating rate and starting microstructure on the morphology of austenite and on the tendency for a memory effect.