

7 *Upper & Lower Bainite*

Although there have been attempts at generalising the definition of bainite, the most appropriate description remains that the microstructure consists of a non-lamellar mixture of ferrite and carbides, which can be classified further into upper and lower bainite. This latter distinction is of value because there are clear differences in the mechanical properties of upper and lower bainite. The two microstructures can easily be distinguished using transmission electron microscopy and hence can be discussed in the context of mechanical properties and the growth mechanism.

Lower bainite is obtained by transformation at relatively low temperatures. Both upper and lower bainite form as aggregates of small plates or laths (sub-units) of ferrite. The essential difference between them is in the nature of the carbide precipitates. Upper bainitic ferrite is free of precipitation, any carbides growing from the carbon-enriched residual austenite between the plates of ferrite. In addition to this kind of precipitation, there are carbide particles present inside lower bainitic ferrite. We shall see that the precipitates in lower bainitic ferrite can be any of the carbides reported to occur during the tempering of martensite, for example, ϵ , η , χ or cementite.

7.1 The Matas and Hehemann Model

The transition between upper and lower bainite is believed to occur over a narrow range of temperatures. It is possible for both forms to occur simultaneously during isothermal transformation near the transition temperature (Pickering, 1967). Matas and Hehemann (1961) proposed that the difference between upper and lower bainite comes from a competition between the rate at which carbides can precipitate from ferrite and the speed with which carbon is partitioned from supersaturated ferrite into austenite (Fig. 7.1). Upper bainite forms at higher temperatures, permitting the excess carbon to partition before it can precipitate in the ferrite. In lower bainite, the slower diffusion associated with the reduced transformation temperature provides an opportunity for some of the carbon to precipitate in the supersaturated ferrite.

A corollary is that upper bainite should not form when the carbon concentration is large. This is indeed found to be the case in a Fe–7.9Cr–1.1C wt% alloy which has a B_s temperature of just 300 °C (Srinivasan and Wayman,

1968a), and in a Fe–4.08Cr–0.3C wt% alloy which has a B_s temperature of 490 °C. Ohmori and Honeycombe (1971) have shown that in a series of high purity Fe–0.16–0.81C wt% alloys, lower bainite is not obtained when the carbon concentration is less than about 0.4 wt%. Tsuzaki *et al.* (1991) found a similar result for Fe–Si–C alloys; only upper bainite formed in a Fe–2Si–1Mn–0.34C wt% steel, whereas both upper and lower bainite could be observed when a higher carbon variant (0.59 wt%) was examined. A thorough piece of work by Oka and Okamoto (1986) on high purity, high carbon Fe–0.85–1.8C wt% steels has shown the absence of upper bainite in all cases. The formation of pearlite was in each case found to give way directly to that of lower bainite.

The model illustrated in Fig. 7.1 has been expressed quantitatively by comparing the time required to decarburise supersaturated ferrite against cementite precipitation kinetics (Takahashi and Bhadeshia, 1990).

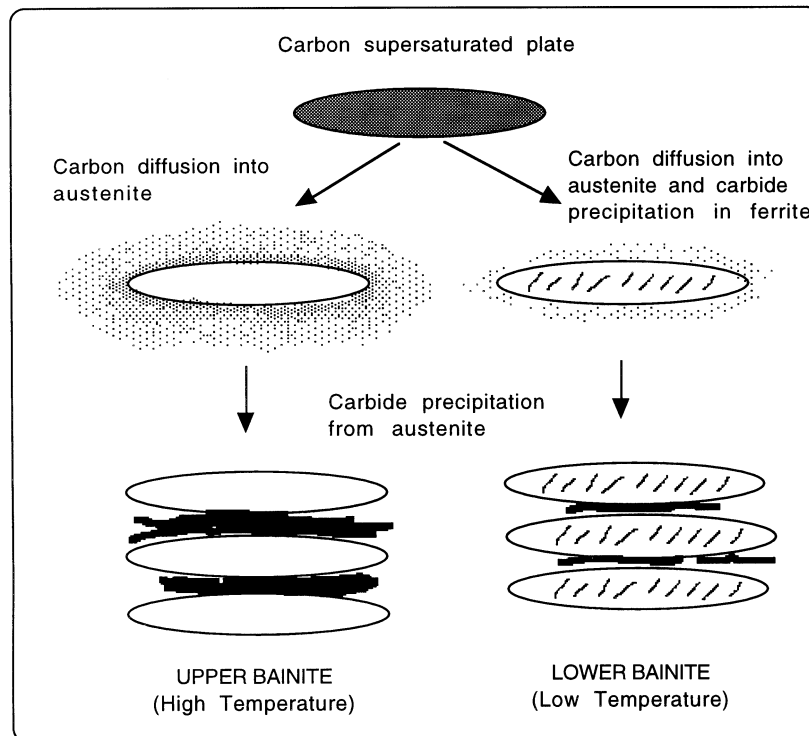


Fig. 7.1 Schematic representation of the transition from upper to lower bainite. After Takahashi and Bhadeshia (1990).

7.2 Quantitative Model

7.2.1 Time to Decarburise Supersaturated Ferrite

The theory for the partitioning of carbon from a supersaturated plate of ferrite has been presented in Chapter 6. The diffusion coefficient of carbon in ferrite is greater than that in austenite. This, together with the assumption that there is local paraequilibrium at the α/γ interface, gives the time t_d required to decarburise a plate of a specified thickness (equation 6.28). Some results for plain carbon steels are presented in Fig. 7.2. In each case, t_d is found to go through a minimum as a function of the transformation temperature. This is because the diffusion coefficient of carbon decreases with temperature (leading to an increase in t_d), while at the same time, the amount of carbon that the austenite can tolerate increases at lower temperatures.

7.2.2 Kinetics of Cementite Precipitation

It is not yet possible to estimate the rate of cementite precipitation from supersaturated ferrite as a function of time, temperature and chemical composition.

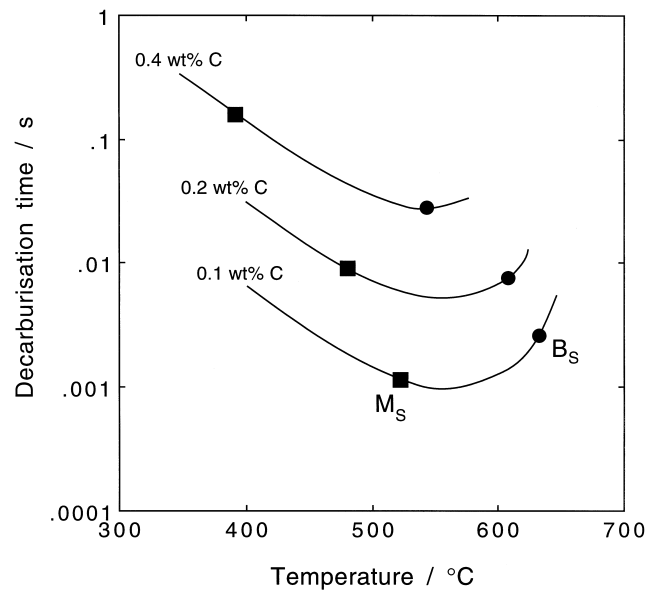


Fig. 7.2 Calculated time for the decarburisation of supersaturated ferrite plates (of thickness $0.2 \mu\text{m}$) in plain carbon steels with 0.1, 0.2 and 0.4 wt% carbon respectively. The calculated martensite-start and bainite-start temperatures are also indicated.

However, for plain carbon steels, and in some cases for alloy steels, martensite tempering data can be adapted to derive reasonable functions for the purpose of predicting the transition from upper to lower bainite (Takahashi and Bhadeshia, 1990, 1991).

The first change that happens during the tempering of supersaturated martensite is that the excess carbon precipitates in the form of carbides. Prolonged annealing can also lead to recovery, recrystallisation and the coarsening of cementite precipitates. To derive a function representing precipitation alone, it is necessary to focus on the early stages of tempering.

Speich (1969) reported that the change in hardness of martensite in plain carbon steels after an hour of tempering at temperatures above 320 °C, includes significant contributions from recovery, recrystallisation and coarsening of cementite particles (Fig. 7.3). The data representing hardness changes during tempering below 320 °C can be used to derive a function which expresses the change in the volume fraction of cementite precipitation as a function of time and temperature. An Avrami equation can then be used empirically to represent the tempering reaction:

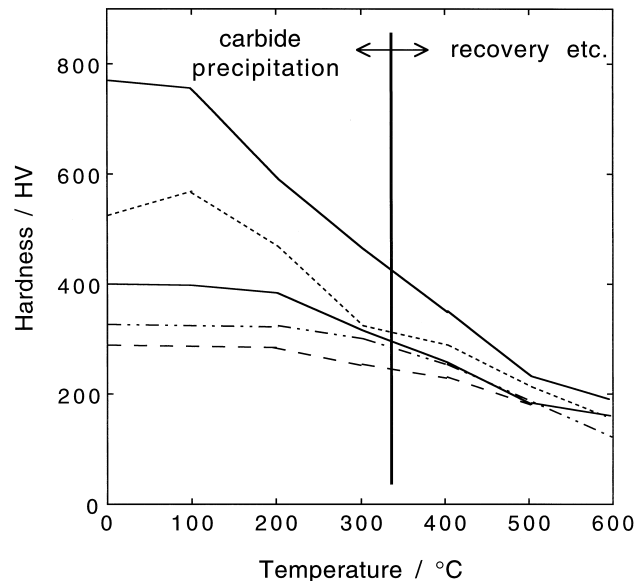


Fig. 7.3 Hardness curves for iron-carbon martensitic samples which were tempered for 1 hour at the temperatures indicated; the five curves represent steels with different carbon concentrations (data due to Speich, 1969). The data to the left of the vertical line reflect changes due to the precipitation of carbides rather than recovery or coarsening processes.

$$\xi\{t\} = 1 - \exp\{-k_A t^n\} \quad (7.1)$$

where $\xi\{t\}$ is the volume fraction of cementite normalised by its equilibrium volume fraction at the reaction temperature, t is the time, and k_A and n are rate constants determined from the experimental data. Since it is assumed that $\xi\{t\}$ is related at any time to the hardness of the martensite, $H\{t\}$, it follows that:

$$\xi\{t\} = (H_0 - H\{t\})/(H_0 - H_F) \quad (7.2)$$

H_0 is the hardness of the as-quenched virgin martensite, H_F is its hardness when all the carbon has precipitated but before any significant recovery, recrystallisation or coarsening has occurred. The assumption here is that the amount of carbon precipitated is related linearly to the change in hardness during the early stages of tempering.

Using the values of hardness for plain carbon martensite tempered for 1 hour at 320 °C, reported by Speich, H_F can be expressed empirically as a function of the initial hardness and average carbon concentration \bar{x} (mole fraction), as follows:

$$H_F = H_0[1 - 1.731\bar{x}^{0.34}] \quad (7.3)$$

This equation is valid for plain carbon steels containing less than 0.4 wt% carbon, the value of H_F becoming constant thereafter. The hardness H_0 of plain carbon martensite (< 0.4 wt% carbon) before tempering can be also be deduced from the data reported by Speich:

$$H_0 = 1267(\text{weight \% carbon})^{0.9} + 240 \quad (7.4)$$

where the hardness of martensite in pure iron is 240 HV (Leslie, 1982). This equation gives the hardness of virgin martensite in plain carbon steels as a function of dissolved carbon. There is evidence that the effect of carbon tends eventually to saturate, so H_0 should be set not exceed about 800 HV irrespective of the carbon concentration (Bhadeshia and Edmonds, 1983a,b). Having established all the data necessary to estimate the amount of cementite precipitated, it remains to evaluate the terms k_A and n of the Avrami equation in order to calculate the time t_θ for the formation of a specified fraction of cementite as a function of time, temperature and carbon concentration. This can easily be done by fitting the Avrami equation to experimental data on the tempering of martensite.

There are more elaborate theories available for the change in the strength of low-carbon martensite due to the precipitation of cementite, making it possible to estimate $H_0 - H_F$ independently of the empirical approach described above. The change can be expressed in terms of the decrease in solid solution strengthening as carbon is incorporated into cementite, and a lesser increase in strength as the cementite particles precipitation harden the martensite. Thus,

the yield strength of martensite, σ_y , is expressed as a combination of the intrinsic yield strength, the effect of the dislocation cell structure, and precipitation hardening by cementite (Daigne *et al.*, 1982):

$$\sigma_y = \sigma_0 + k_\epsilon \epsilon_1^{-1} + k_p \Delta^{-1}, \text{MPa} \quad (7.5)$$

where σ_0 is the intrinsic strength of martensite (including solid solution strengthening due to carbon), ϵ_1 is the average transverse thickness of the cell structure, and Δ is the average distance between a particle and its two or three nearest neighbours. The data needed to evaluate the equation are well-founded. A comparison of the calculated strength and measured strength after tempering should give a good idea of the extent of cementite precipitation. When this is done, the relation between hardness and the amount of the precipitation (thus the decrease in solute carbon) is found not to be linear as was assumed in the empirical approach, but the predicted changes in hardness are found to be remarkably consistent with those measured by Speich for the early stages of tempering. This justifies the assumption that much of the hardness change can be attributed to the precipitation of carbon rather than due to other annealing effects such as tempering.

7.2.3 Quantitative Estimation of the Transition Temperature

Following the gist of the Matas and Hehemann proposal, a comparison of the time t_d required to decarburise a plate of ferrite, with the time interval t_θ necessary to obtain a detectable amount of cementite precipitation in the ferrite should give a good indication of whether upper or lower bainite is expected during isothermal transformation. If $t_d < t_\theta$ then it may be assumed that upper bainite is obtained, and vice versa (Fig. 7.4). A weakness of this theoretical model is that decarburisation and precipitation should really be coupled. A disposable parameter in the model as it stands is the 'detectable amount' of cementite precipitation, which has to be fixed by comparison with experimental data.

Some calculated data on the plain carbon steels are presented in Fig. 7.5. They indicate that lower bainite should not be observed in plain carbon steels with carbon concentrations less than 0.32 wt%. Furthermore, only lower bainite (i.e. no upper bainite) is expected in steels with carbon concentrations exceeding 0.4 wt%. Steels containing between 0.32 and 0.4 wt% of carbon should exhibit both both upper and lower bainite, depending on the reaction temperatures. Finally, at low temperatures where t_θ and t_d both become large, the times required for precipitation or redistribution of carbon exceed that to complete transformation, consistent with the fact that untempered martensite

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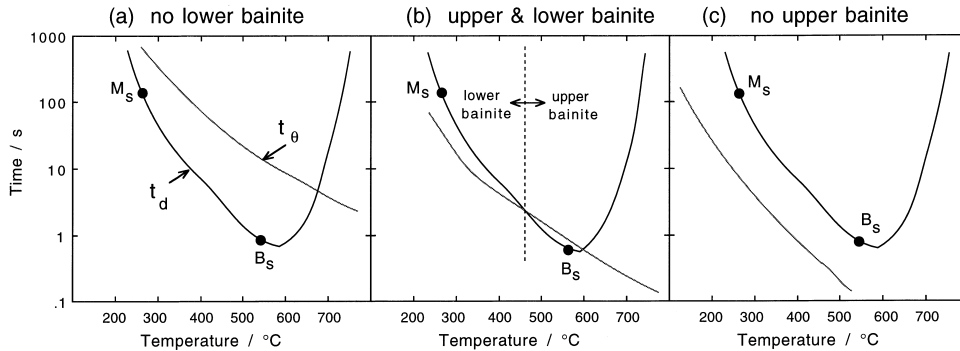


Fig. 7.4 Illustration of how differences in the relative variation of the decarburisation time t_d and the precipitation time t_θ can lead to: (a) a steel which is incapable of transforming to lower bainite; (b) a steel which should, under appropriate conditions, be able to transform to upper or lower bainite; (c) a steel in which bainitic transformation always leads to the formation of lower bainite.

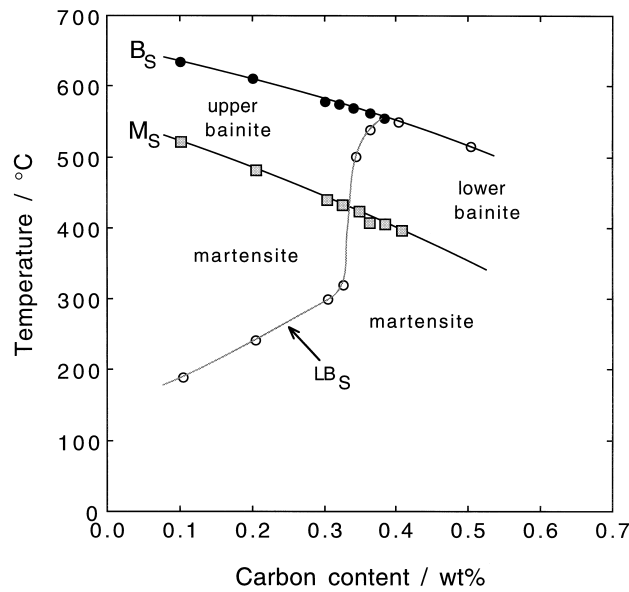


Fig. 7.5 Calculated lower bainite transformation start temperatures for plain carbon steels, as a function of transformation temperature. The calculated martensite-start and bainite-start temperatures are also presented. After Takahashi and Bhadeshia, 1990.

can be obtained at temperatures near M_S , with the degree of autotempering of the martensite decreasing as M_S is reduced.

7.2.4 Comparison of Theory and Experimental Data

The general behaviour indicated by the calculations for plain carbon steels, is found to be that observed experimentally. Some interesting work by Oka and Okamoto (1986) proves that there is no upper bainite in plain carbon steels with more than 0.8 wt% of carbon; the only bainite observed in their experiments was classical lower bainite at all temperatures above the M_S temperature (Fig. 7.6a).

Ohmori and Honeycombe (1971), in a study of plain carbon steels, showed that during isothermal transformation above M_S , only upper bainite could be obtained in samples containing less than 0.4C wt% (Fig. 7.6b). This is consistent with theory, although their observation that upper bainite can be obtained in steels with a carbon concentration up to about 0.85C wt% is not consistent with the theory, nor with the data reported by Oka and Okamoto (1986).

7.3 Mixed Microstructures Obtained By Isothermal Transformation

According to the analysis presented above, only lower bainite is expected in plain carbon steels with more than 0.32 wt% of bulk carbon content. However, the calculations are for ferrite plates whose carbon concentration is at first identical to that of the bulk alloy. As transformation proceeds the austenite becomes enriched in carbon. The bainite which grows from this enriched austenite will inherit a higher than bulk concentration of carbon. This leads to the possibility of upper bainite being followed by lower bainite during isothermal transformation. Mixed microstructures should therefore be obtained in plain carbon steels containing less than 0.32 wt% carbon, especially if the rate of carbide precipitation from the austenite is slow enough to allow the austenite to become enriched.

The maximum carbon concentration that can be tolerated in residual austenite before the bainite reaction stops is given by the T'_0 curve. Therefore if the carbon concentration in residual austenite at the T'_0 curve (i.e. $x_{T'_0}$) is greater than 0.32 wt%, lower bainite can form during the later stages of transformation.

However, the formation of cementite from the residual austenite also becomes possible if $x_{T'_0} > x^{\gamma\theta}$, where $x^{\gamma\theta}$ is a point on the $\gamma/(\gamma + \theta)$ phase boundary, since the austenite will then be supersaturated with respect to the cementite. The fact that a curve showing the carbon concentration in austenite which is in equilibrium with cementite in plain carbon steels crosses the T'_0

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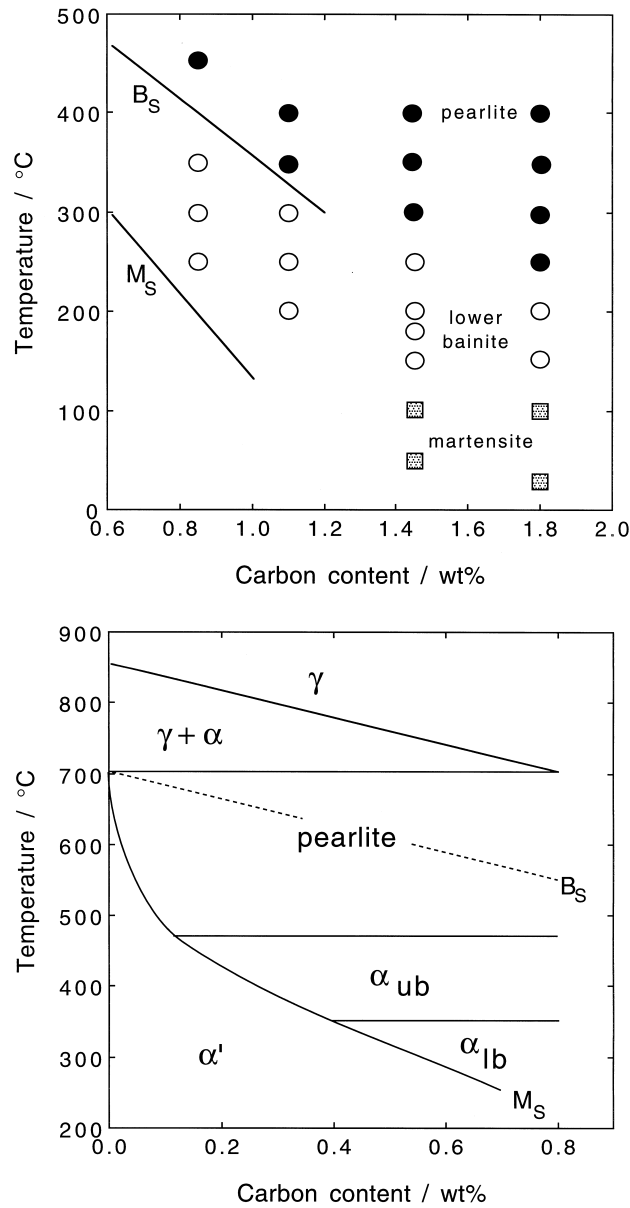


Fig. 7.6 (a) Experimental data (Oka and Okamoto, 1986) illustrating the temperatures at which fine nodules of pearlite, classical lower bainite and martensite were obtained by isothermal transformation of plain carbon steels. The lines represent calculated bainite-start and martensite-start temperatures (Takahashi and Bhadeshia, 1990). (b) The effect of carbon concentration on the temperature range where each microstructure is formed (Ohmori and Honeycombe, 1971).

curve at 0.4 wt% of carbon concentration (560 °C), leads to the identification of three regimes for bainite on the Fe–C phase diagram (Fig. 7.7). In steels with more than 0.4 wt% of the initial bulk carbon content (region B), lower bainite is to be expected from the earliest stages of transformation. For steels whose composition lies in region A, lower bainite is expected to be absent during isothermal transformation at all temperatures above M_S , and this behaviour is valid for any stage of transformation since the austenite cannot be supersaturated with cementite as far as regime A is concerned. The behaviour in the region marked C should be more complex. The residual austenite for these steels (region C) may at some stage of transformation contain enough carbon to precipitate cementite. If the kinetics of cementite precipitation from austenite are rapid, then lower bainite may not be obtained in steels with an average carbon concentration less than 0.32 wt%, but otherwise, a mixed microstructure of upper and lower bainite might arise.

Two of the trends described above have been verified by Chang (1999) who not only found that the lower bainite start temperature L_S could be depressed by retarding the precipitation of cementite, but also showed that a mixture of upper and lower bainite can be obtained by transformation at temperatures just above L_S .

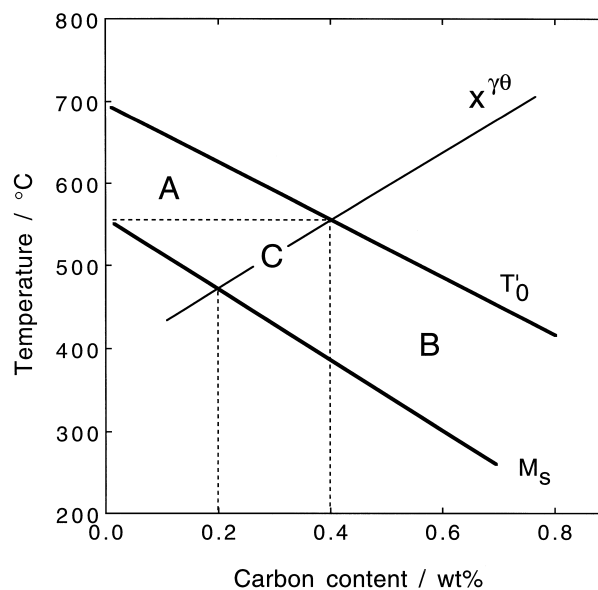


Fig. 7.7 Identification of regimes A, B and C, in which the progress of isothermal transformation can lead to changes in the nature of the transformation product. The line marked $x^{\gamma\theta}$ is the calculated $\gamma/(\gamma + \theta)$ phase boundary.

7.4 Other Consequences of the Transition

The growth of bainite plates stops when the glissile interface is clogged by plastic accommodation induced defects. For a given defect density, lower-bainite plates should be longer than those of upper bainite, since the driving force for transformation increases with the undercooling. At lower transformation temperatures the matrix is able to support higher strains without plastic deformation so that the defect density *in the matrix* is expected to be smaller. Step quenching experiments in which an alloy is first partially transformed to lower bainite and then heated rapidly into the upper bainite transformation range are consistent with this since the growth of lower bainite ceases on reaching the higher temperature (Goodenow and Hehemann, 1965). This also happens when specimens partially transformed to lower bainite experience an increase in temperature within the lower bainite transformation range (White and Owen, 1961).

7.5 Comparison with the Tempering of Martensite

We have seen that the transition from upper to lower bainite can be predicted on the basis of a simple model which compares the rates of decarburisation and precipitation. Lower bainite is in effect generated by a process of auto-tempering. It follows that there should be a strong comparison with the microstructure of tempered martensite.

When high-carbon martensite is tempered, the first precipitate is usually a transition phase such as ϵ -carbide, which is replaced eventually by the more stable cementite. Similarly, when lower bainite forms in high carbon steels, it contains ϵ -carbide which subsequently transforms into cementite during prolonged holding at the isothermal transformation temperature (Matas and Hehemann, 1961).

The chances of obtaining ϵ -carbide (instead of cementite) in lower bainite increase as the transformation temperature is reduced for the same steel (see Table II, Matas and Hehemann, 1961). As the transformation temperature is reduced, the time required to decarburise a supersaturated plate of bainite increases. A high carbon concentration can persist in the ferritic matrix for a time period long enough to allow the formation of ϵ -carbide, which does not form if the carbon concentration is less than about 0.25 wt%, (Roberts *et al.*, 1957). This explains why a medium carbon Fe-0.43C-3Mn-2Si wt% steel transforms to lower bainite containing cementite particles, although when quenched to martensite, gives ϵ -carbide on tempering (Bhadeshia and Edmonds, 1979a, 1983a). Some of the carbon in the former case is lost to the austenite by diffusion, thus preventing the formation of ϵ -carbide.

7.6 Summary

A comparison between the times required to decarburise supersaturated ferrite plates with that required to precipitate cementite within the plates is a reasonable way to interpret the transition from upper to lower bainite. If the decarburisation process dominates, upper bainite is predicted whereas relatively rapid carbide precipitation within the ferrite leads to the microstructure of lower bainite.

A number of predictions from this theory are in agreement with experimental data. Thus, lower bainite cannot form in plain carbon steels containing less than about 0.3 wt% carbon. Similarly, upper bainite is predicted and found to be absent in plain carbon steels containing more than 0.4 wt% carbon.