Bainite in steels
H K D H Bhadeshia

The author is in the Department of Materials Science and Metallurgy at the University of Cambridge.

INTRODUCTION

The bainite reaction in steels does not fit neatly into any of the established categories of transformations, and shows characteristics which apparently contradict each other. The subject has been scrutinised in such detail that it can seem to be in a state of confusion. This impression is far from the truth; we present here an assessment of the transformation which is consistent in itself, with all available experimental data and with the other transformations which arise in steels.

To conserve space, it has been necessary to be selective in the topics emphasised and in many cases to refer to review articles rather than the original sources. Many topics have been dealt with in detail in an excellent recent review [1]; these are discussed briefly, with more attention being paid to new developments in the subject.

GENERAL CHARACTERISTICS

Both upper bainite (α_{ub}) and lower bainite (α_{lb}) tend to form as aggregates (sheaves) of small lenticular platelets of ferrite, separated by regions of austenite (γ), martensite (α') and/or cementite (θ). The sheaf itself has a morphology on a macroscopic scale of a wedge shaped plate. Unlike the products of diffusional transformation, the growth of bainite sheaves is stopped by hard impingement with austenite grain or twin boundaries. Even in the absence of such impingement, individual platelets in a sheaf grow to a limiting size, so that the platelets near the tip of a sheaf have approximately the same size as those near the site where the sheaf nucleated [1-3].

The formation of both upper and lower bainite causes a change in the shape of the transformed region; the shape change is known to be an invariant-plane strain (IPS) with a significant shear component [2,4,5]. The IPS shape change is a physical deformation of a type which *necessitates* a coordinated movement of atoms during transformation. It also implies the existence of an atomic correspondence, at least for the iron and substitutional alloying elements, between the parent and product crystals [6]. As a consequence of this shape deformation, which is identical for each platelet within a sheaf, bainitic ferrite has a stored energy of ≈ 400J/mol [7], although this may be reduced by plastic relaxation. This transformation induced plastic deformation is also believed to be responsible for the high dislocation density of bainitic ferrite, and for the stifling of the growth of individual platelets as dislocation tangles block the advance of the transformation interface [8]. Adjacent sheaves of bainite may also accommodate each others shape strains [2], although there is no evidence that such sheaves always grow in a cooperative manner.

The plate morphology of bainite has been attributed to variations in interface mobility as a function of interface orientation [9], but experimental data indicate otherwise. The plate shape in fact arises from the need to minimise the strain energy associated with the IPS shape deformation. For example, *all* kinds of ferrite plates obtained by transformation from austenite are associated with IPS surface relief effects. On the other hand, none of the ferrite obtained by diffusional transformation from γ grows in the form of a plate, even when growth is by a ledge mechanism.
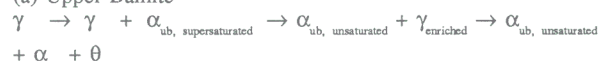
Although it has been suggested that an IPS shape deformation can arise when a sessile semi-coherent interface is displaced by the motion of steps [10], a mechanism for this does not exist. The idea is inconsistent within itself, because the strain energy due to the deformation is then ignored; when the transformation is constrained by its surroundings, strain arises irrespective of the mechanism by which the shape change occurs. It also contradicts the fact that allotriomorphic ferrite, which often grows by a step mechanism (and has the necessary sessile semi-coherent interface) does not show an IPS surface relief. Annealing twins in cubic close-packed crystals should by the hypothesis develop the same IPS surface relief as mechanical twins, but they do not do so. The concept violates the fact that the atomic correspondence implied by the shape deformation is a property of the particle as a whole, irrespective of interface orientation [1].

The ferritic component of bainite forms first and the regions of residual austenite between the platelets of ferrite then either decompose diffusionally to cementite (or other carbides) and ferrite, or partially decompose martensitically during cooling to ambient temperature. In the case of lower bainite, either ε or θ may additionally be found within the bainitic ferrite; experimental data suggest that these carbides precipitate in order to relieve the carbon supersaturation within the bainitic ferrite; the alternative hypothesis that precipitation occurs at the γ/α_{lb} interface is inconsistent with detailed crystallographic evidence and direct observations of the α_{lb}/γ interface [1,11].

Consistent with the IPS shape change accompanying transformation, there is no diffusion of substitutional (or indeed iron) atoms during the growth of bainite. This has been established to a very high resolution in atom-probe experiments which also show that substitutional alloying additions do not segregate to the transformation interface [12,13].

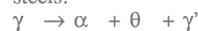The sequence of transformation may be summarised as follows:

(a) Upper Bainite

$$\gamma \rightarrow \gamma + \alpha_{ub, \text{ supersaturated}} \rightarrow \alpha_{ub, \text{ unsaturated}} + \gamma_{\text{enriched}} \rightarrow \alpha_{ub, \text{ unsaturated}} + \alpha + \theta$$

This contrasts with the cooperative growth of cementite and ferrite during the formation of pearlite in plain carbon steels:

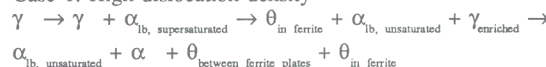$$\gamma \rightarrow \alpha + \theta$$
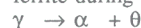
and with the formation of pearlite in substitutionally alloyed steels:

$$\gamma \rightarrow \alpha + \theta + \gamma'$$

Note that γ' differs in composition from γ.

(b) Lower Bainite: the dislocation density of α_{lb} can influence the formation of either ε or θ [11,14].

Case 1: High dislocation density

$$\gamma \rightarrow \gamma + \alpha_{lb, \text{ supersaturated}} \rightarrow \theta_{\text{in ferrite}} + \alpha_{lb, \text{ unsaturated}} + \gamma_{\text{enriched}} \rightarrow$$
$$\alpha_{lb, \text{ unsaturated}} + \alpha + \theta_{\text{between ferrite plates}} + \theta_{\text{in ferrite}}$$

Case 2: Low dislocation density

$$\gamma \rightarrow \gamma + \alpha_{lb, \text{ supersaturated}} \rightarrow \varepsilon\text{-carbide}_{\text{in ferrite}} + \alpha_{lb, \text{ unsaturated}} + \gamma_{\text{enriched}}$$
$$\rightarrow \alpha_{lb, \text{ unsaturated}} + \varepsilon\text{-carbide}_{\text{in ferrite}} + \alpha + \theta_{\text{between ferrite plates}} \rightarrow$$
$$\alpha_{lb, \text{ unsaturated}} + \theta_{\text{in ferrite}} + \theta_{\text{between ferrite plates}} + \alpha$$
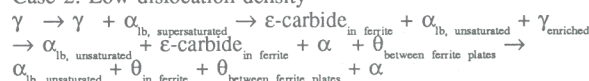
THE INCOMPLETE REACTION PHENOMENON

In steels where the bainite transformation can be studied without interference from other reactions, it is found that the maximum volume fraction of bainite obtained during isothermal transformation from austenite increases from zero as the transformation temperature decreases below the B_s temperature [1,2]. The fact that bainite is not obtained at all for $T>B_s$, together with the observation that for $T<B_s$, the transformation ceases well before the carbon concentration of the residual austenite reaches the equilibrium or paraequilibrium $\alpha+\gamma/\gamma$ phase boundary is the essence of the *incomplete reaction phenomenon*.

A key problem in the study of bainite is to determine the carbon concentration of bainitic ferrite during its growth; as discussed later, at the relatively high temperatures where bainite forms, any excess carbon in the ferrite can rapidly partition into the residual austenite, so that the problem is not amenable to direct measurements. If a platelet of bainitic ferrite grows with the composition of the parent austenite, and *subsequently* rejects its excess carbon into the residual austenite, then the next platelet would have to form from γ which is enriched in carbon. A reaction like this would stop when the carbon concentration of the residual austenite becomes high enough to prevent composition invariant transformation (i.e., when x_γ reaches the T'_o curve on the phase diagram. The curve represents the points on the phase diagram where α, whose free energy is raised by a stored energy term, and γ of identical composition have the same free energy.).

This behaviour is observed experimentally and explains the incomplete reaction phenomenon and at the same time indicates that the growth of bainite is diffusionless.

The phenomenon has not been established experimentally in steels which do not have sufficient hardenability. In such steels, other reactions, like the formation of pearlite, or the precipitation of cementite from austenite, can interfere with the interpretation of results because they alter the composition of the residual austenite. The precipitation of carbides within the bainitic ferrite would also remove carbon from the system and consequently allow the transformation to proceed to a greater extent.

Time-Temperature-Transformation (TTT) diagrams for steels show an apparent slowing down of transformation in the vicinity of the B_s temperature, but only if it is assumed that the entire TTT diagram is due to a single reaction. This is of course not true in reality; reactions which have radically different transformation characteristics can be found over the temperature range concerned. Nonetheless, the apparent bay is attributed to a "solute-drag like effect" [15] which allegedly retards transformation at that temperature. The solute-drag like effect is somehow also supposed to explain the incomplete reaction phenomenon. These concepts are weak [16,17]; the bay really arises due to the overlap of two different C-curves corresponding to diffusional and displacive transformations. The solute-drag like effect relies on the segregation of alloying elements to the transformation interfaces; atom-probe experiments show that this does not happen [12,13]. A detailed assessment [17] of the evidence to date has shown that the effect has not been adequately established either experimentally or theoretically.

Finally, kinetic data confirm the conclusion that bainite grows with a supersaturation of carbon (and without any retardation due to solute-drag), since the lengthening rates of sheaves and of subunits far exceed those calculated on the basis of paraequilibrium, carbon diffusion-controlled growth [18,19].

REDISTRIBUTION OF CARBON

Since bainitic ferrite grows with a non-equilibrium concentration of carbon, then given the opportunity, the carbon should tend to partition into the residual austenite, where it has a lower chemical potential. The time t_d needed to decarburise the ferrite is intuitively expected at least to be comparable to that required for a sub-unit to complete its growth. If t_d is small relative to the time required to relieve the carbon supersaturation by the precipitation of carbides within the ferrite, then upper

bainite is obtained; otherwise, lower bainite forms.

Kinsman and Aaronson [20] first considered the kinetics of the partitioning of carbon from bainitic ferrite of the same composition as the parent phase. For a plate of thickness t_h, the flux of carbon is defined along a coordinate z normal to the α/γ interface, with origin at the interface and z being positive in the austenite. From the mass balance for carbon it follows that:

$$(0.5t_h)(\bar{x} - x^{\alpha\gamma}) = \int_0^\infty [x_\gamma\{z, t_d\} - \bar{x}] \, dz$$

where $\bar{x}$ is the average mole fraction of carbon in the alloy and $x^{\alpha\gamma}$ and $x^{\gamma\alpha}$ are the paraequilibrium carbon concentrations in α and γ respectively; these concentrations should strictly allow for the stored energy of bainitic ferrite. The equation corrects an error in the original treatment; the error had the effect of allowing $t_d\rightarrow 0$ as the upper integration limit $\rightarrow\infty$. The function x_γ is given by:

$$x_\gamma = \bar{x} + (x^{\gamma\alpha} - \bar{x}) \, \text{erfc}\{z \, / \, 2(Dt_d)^{0.5}\}$$

This assumes that for $t<t_d$, the concentration of carbon in the austenite at the interface is given by $x^{\gamma\alpha}$. The concentration dependent diffusivity $D\{x\}$ of carbon in austenite can to a good approximation be taken to be a weighted average diffusivity $\bar{D}$ [16]. On carrying out the integration we get:

$$t_d^{0.5} = t_h(\bar{x}-x^{\alpha\gamma})\pi^{1/2}/ \, 4\bar{D}^{0.5}(x^{\gamma\alpha}-\bar{x})$$

For an Fe-3.0Mn-2.02Si-0.43C wt.% alloy, our calculations show that t_d is in fact much larger than has generally been believed in the past (Fig. 1). It may be significant that t_d increases sharply as T decreases, becoming substantial in the vicinity of the transition from upper to lower bainite; a detailed analysis of the transition must await a working model for the kinetics of carbide precipitation.

The model presented above is still not strictly correct because it does not consider the coupling of fluxes between the ferrite and austenite, a coupling which ensures that there is no build up of solute at the interface. This is difficult to treat analytically, because the ferrite has a finite extent, but an analysis using a finite-difference method shows that the values of t_d calculated above are slightly overestimated (unpublished research). At high volume fractions of bainite, soft-impingement in the austenite also becomes important and this has the effect of increasing t_d. If t_d becomes large when compared with the time needed to nucleate new subunits, then the volume fraction of bainite obtainable at any temperature would increase because the partitioning of carbon would lag behind the progress of the transformation.

THE APPROACH TO EQUILIBRIUM

We have already seen that the bainite reaction stops well before the austenite reaches its equilibrium composition or volume fraction. The transformation is kinetically favoured, but does not lead to the largest possible free energy change during the decomposition of austenite. Given the opportunity, both the bainite and its associated residual austenite tend towards equilibrium.

If an alloy is held at an isothermal transformation temperature below B_s where diffusional transformations are sluggish, the growth of bainite eventually stops as the driving force for displacive transformation reaches zero. However, the alloy is still far from its equilibrium state and the residual austenite is able to transform diffusionally during continued holding at the isothermal transformation temperature, although at a much slower rate. For example, in an Fe-3.0Mn-2.12Si-0.43C wt.% alloy, it is found that the transformation to upper bainite at $452^{\circ}C$ ceases in a matter of minutes, but after holding the specimen at this temperature for 32 hours, a small amount of pearlite forms from the residual austenite [8]. If an Fe-4.08Cr-0.3C wt.% alloy is isothermally transformed at $478^{\circ}C$, the lower bainite transformation stops within some 30 minutes but continued holding of the same specimen at $478^{\circ}C$ for 43 days causes the residual austenite to undergo extremely slow diffusional transformation to two different products [21,22]. One

of these is the alloy pearlite which nucleates at the austenite grain boundaries and develops as a separate transformation. In the other, the original bainite/austenite interfaces move to produce *diffusional* transformation (Fig. 2). The interfaces degenerate into a series of perturbations having an extremely irregular morphology. The perturbations have exactly the same crystallographic orientation as the original bainite (i.e., they are contiguous with the bainite) and they grow with the same substitutional alloy content as the parent austenite; this further transformation is not accompanied by an IPS shape change. During the isothermal heat treatment, the perturbations grew in 43 days to a length comparable to the thickness of the original bainite plate (which took a few seconds to grow). The results are highly significant since they demonstrate unambiguously that the *same* interface accomplishes diffusional transformation at a rate which is many orders of magnitude smaller than when it led to the displacive formation of bainite. They are consistent with the IPS shape change which accompanies the growth of bainite and which implies that the bainite/austenite interface is of a glissile, semi-coherent type with high mobility even at low homologous temperatures. Another interesting point is that the perturbations do not develop into a plate shape, confirming that the orientation dependence of interface energy does not play a key role in determining the shape of the ferrite.

It has long been established that the cementite associated with upper bainite has a substitutional alloy content close to (or slightly higher than) that of the steel as a whole, even though this may be well below the corresponding equilibrium concentration [23]. Chance and Ridley [24] found that for upper bainite in an Fe-0.81C-1.41Cr wt.% alloy, the partition coefficient k_{Cr} (wt.% Cr in θ / wt.% Cr in α_{ub}) could not be distinguished from unity (Fig. 3). Because the carbon concentrations of θ and α or γ are substantially different, a composition invariant transformation to cementite should give k_{Cr} *lower* than unity. Hence, the formation of cementite during the upper bainite transformation does seem to involve a redistribution of elements like Cr to the cementite. On the other hand, it is established that the degree of partitioning is much greater in the case of pearlitic cementite, even when the latter forms at the same transformation temperature (Fig. 3). The explanation for this may lie in the fact that unlike pearlitic cementite, upper bainitic cementite forms from *carbon-enriched* austenite, so that the driving force for cementite formation is higher; the tolerable departure from the equilibrium composition can thus be greater for upper bainitic cementite.

During the tempering of bainite, before any alloy carbides are precipitated, and with the help of diffusion, the cementite tends towards its equilibrium composition [23]. The exact manner in which its composition alters must depend on steel composition, starting microstructure and tempering conditions. There is as yet no theory to relate these variables and the work reviewed below should be interpreted with caution; many of the experiments deal with mixed microstructures (e.g. bainite and allotriomorphic ferrite) produced by continuous cooling transformation, and do not distinguish between the bainitic cementite which precipitates from austenite and that which forms from any lower bainitic ferrite.

The composition of carbides during the tempering of bainite has in industry assumed new significance in recent years and the subject seems ripe for a fresh theoretical study. Where creep resistant bainitic steels are in service at elevated temperatures for many years it is important for safety reasons to know accurately the time-temperature history of the steel at any stage during service. This determines the amount of creep life remaining (i.e., the *remanent life*). It has been proposed that the composition of cementite, monitored using non-destructive microanalytical techniques, can be used as a built in recorder for the time-temperature cycle experienced by the steel. The method was first applied to pearlitic carbides where it was found that the Cr and Mn concentrations in θ varied approximately with $t^{1/3}$, where t is the time at tempering temperature [25]. This relationship has not been justified theoretically.

Afrouz *et al*. [26] reported similar results on the cementite associated with bainite. Some service exposed material (70000hrs @ 565°C) with a mixed microstructure of allotriomorphic ferrite and $\approx$ 20% bainite was tempered further at 550°C (Fig. 4). As expected, the Cr and Mn concentrations of M_3C were found to increase with time, the Mn possibly showing signs of saturation at high t, with the Mo exhibiting considerable scatter. Afrouz *et al*. also reaustenitised (reheat-treated) the service exposed material so that after oil-quenching, a fresh fully bainitic microstructure (probably a mixture of upper and lower bainite) was obtained. The change in M_3C composition was monitored during tempering at 550°C. The starting composition of the cementite was of course leaner than that of the service exposed material and the rate of enrichment was found to be higher for the reheat-treated samples. In both cases, the composition changed approximately as $t^{1/3}$, the slope being greater for the reheat-treated material. The results are interesting and provide a clear method of monitoring the thermal history of creep resistant steels. The details however are confusing; the fact that the service exposed material changes composition at a slower rate implies that the enrichment process does not really follow a $t^{1/3}$ relationship. In addition, the enrichment of the service exposed M_3C, prior to the 550°C temper, seems very low considering that it has been held at 565°C for 70000hrs. The discrepancy could arise because the 20% bainite present in the service exposed material originally formed from *carbon-enriched* austenite; the cementite particles would then be relatively coarse and closely spaced and would consequently enrich at a lower rate. Much theoretical work needs to be done to model the diffusion process and provide a framework for the interpretation of the results.

ACICULAR FERRITE

"Acicular ferrite" (α_a, Fig. 5) is a phase formed by the transformation of austenite during cooling of low-alloy steel weld deposits [27]. It is of considerable commercial importance because it provides a relatively tough and strong microstructure. It forms in a temperature range where diffusional transformations become relatively sluggish and give way to displacive reactions such as Widmanstätten ferrite, bainite and martensite.

The term *acicular* means shaped and pointed like a needle but it is generally recognised that α_a has in three-dimensions the morphology of thin, lenticular plates. The true aspect ratio of such plates has never been measured but in random planar sections, the plates are typically about 10μm long and $\approx$ 1μm wide (this implies a true aspect ratio much smaller than 0.1).

During the early stages of transformation, the plates nucleate on inclusions present in the large columnar austenite grains typical of weld deposits. Subsequent plates may nucleate autocatalytically on these inclusion nucleated plates, so that a one-to-one correspondence between the number of active inclusions and the number of α_a plates is not expected.

The shape change accompanying the formation of α_a has been characterised qualitatively as an invariant-plane strain; other measurements imply that the stored energy of acicular ferrite is $\approx$ 400J/mol [28,29].

Plates of α_a have never been found to cross austenite grain boundaries and the orientation relationship between α_a and the γ grain in which it grows is *always* such that a close-packed plane of the austenite is parallel or nearly parallel to a closest-packed plane of α_a, and corresponding close-packed directions within these planes are within a few degrees of each other [29].

The acicular ferrite transformation obeys the incomplete-reaction phenomenon, the degree of reaction tending to zero as the transformation temperature rises towards the B_s temperature; at a given temperature, the transformation stops as x_γ reaches the T'$_0$ curve. The evidence all indicates that the growth of acicular ferrite is diffusionless, with carbon partitioning into austenite after the transformation event [28,29].

The experimental data to date indicate that acicular ferrite is essentially identical to bainite. Its detailed morphology differs from that of conventional bainite because the former nucleates intragranularly at inclusions within large γ grains whereas in wrought steels which are relatively free of non-metallic inclusions, bainite nucleates initially at γ/γ grain surfaces and continues growth by the repeated formation of subunits, to generate the classical sheaf morphology. Acicular ferrite does

not normally grow in sheaves because the development of sheaves is stifled by hard impingement between plates nucleated independently at adjacent sites. Indeed, conventional bainite or acicular ferrite can be obtained under identical isothermal transformation conditions in the same (inclusion rich) steel; in the former case, the austenite grain size has to be small in order that nucleation from grain surfaces dominates and subsequent growth then swamps the interiors of the γ grains [28,30]. For a larger γ grain size, intragranular nucleation on inclusions dominates, so that α_a is obtained. Hence, the reason why α_a is not usually obtained in wrought steels is because they are relatively free of inclusions and because most commercial heat treatments aim at a small austenite grain size.

RATIONALISATION OF TRANSFORMATIONS [7]

There are several displacive transformation products in steels: Widmanstätten ferrite, bainite and martensite. Widmanstätten ferrite is obtained at relatively low undercoolings and generally involves the cooperative formation of adjacent and mutually accommodating crystallographic variants which are usually (though not always) similarly orientated in space. Because the adjacent variants have slightly different habit planes with the austenite, the plate as a whole has a thin-wedge morphology. Hence, what appears in a light microscope to be a single plate of Widmanstätten ferrite is in reality a composite of two adjacent and mutually accommodating plates. During its growth, Widmanstätten ferrite has an equilibrium or paraequilibrium carbon concentration, lengthening occurring approximately at a rate controlled by the diffusion of carbon in the austenite ahead of the interface.

The growth of bainite is diffusionless, but because the transformation occurs at relatively high temperatures, the excess carbon is rapidly partitioned into the residual austenite, or is precipitated inside the ferrite, in which case, lower bainite is obtained. The fact that the bainite transformation occurs at relatively low driving forces (and high temperatures where the austenite is mechanically soft) means that the growth of platelets (subunits) is stifled by an accumulating debris of deformation induced defects. Consequently, the transformation develops by the diffusionless growth of platelets, each of which is of a limited size. Since after nucleation at an austenite grain surface, new platelets form autocatalytically on those already present, aggregations of platelets are obtained, and these are the classical sheaves of bainite.

Carbon is partitioned into the α during the nucleation of both bainite and Widmanstätten ferrite. However, the activation free energy of nucleation is directly proportional to the chemical driving force, so that the nucleation process is in this sense similar to that in martensitic transformations. Martensite nucleation (and growth) is on the other hand diffusionless.

CONCLUSIONS

The main conclusions are summarised in Table 1. These seem to us to represent the best fit to available theory and experimental data. There is probably no other material for which there is such detailed knowledge about the transformation products. This knowledge, combined with an understanding of the totality of transformations in steels should provide an exciting opportunity for theoretical work on the prediction of the microstructure of real steels.

ACKNOWLEDGMENTS

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Table 1

Key Characteristics of Transformations in Steels

The table below indicates the key characteristics of phase transformations in steels; it may be regarded as a working hypothesis which seems most consistent with the available experimental data. The nomenclature used for the transformation products is as follows: martensite (α'), lower bainite (α_{lb}), upper bainite (α_{ub}), acicular ferrite (α_a), Widmanstätten ferrite (α_w), allotriomorphic ferrite (α), idiomorphic ferrite (α_i), pearlite (P), substitutional alloying elements (X). Consistency of a comment with the transformation concerned is indicated by (=), inconsistency by ($\neq$); cases where the comment is only sometimes consistent with the transformation are indicated by ($\otimes$). The term *parent* γ implies the γ grain in which the product phase grows. Note that it is not justified to distinguish massive ferrite from α.

| Comment	α'	α_{lb}	α_{ub}	α_a	α_w	α	α_i	P
Nucleation and growth reaction	=	=	=	=	=	=	=	=
Plate morphology	=	=	=	=	=	$\neq$	$\neq$	$\neq$
Invariant-plane strain shape change	=	=	=	=	=	$\neq$	$\neq$	$\neq$
Diffusionless nucleation	=	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$
Diffusion only of C during nucleation	$\neq$	=	=	=	=	$\neq$	$\neq$	$\neq$
Reconstructive diffusion during nucleation	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$	=	=	=
Often nucleates intragranularly on defects	=	$\neq$	$\neq$	=	$\neq$	$\neq$	=	$\neq$
Diffusionless growth	=	=	=	=	$\neq$	$\neq$	$\neq$	$\neq$
Reconstructive diffusion during growth	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$	=	=	=
Atomic correspondence (all atoms) during growth	=	=	=	=	$\neq$	$\neq$	$\neq$	$\neq$
Atomic correspondence for Fe, X atoms during growth	=	=	=	=	=	$\neq$	$\neq$	$\neq$
Bulk redistribution of X atoms during growth	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$	$\otimes$	$\otimes$	$\otimes$
Local equilibrium at interface during growth	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$	$\otimes$	$\otimes$	$\otimes$
Local paraequilibrium at interface during growth	$\neq$	$\neq$	$\neq$	$\neq$	=	$\otimes$	$\otimes$	$\neq$
Diffusion of C during transformation	$\neq$	$\neq$	$\neq$	$\neq$	=	=	=	=
Carbon diffusion-controlled growth	$\neq$	$\neq$	$\neq$	$\neq$	=	$\otimes$	$\otimes$	$\otimes$
Cooperative growth of ferrite and cementite	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$	=
High dislocation density	=	=	=	=	=	$\neq$	$\neq$	$\neq$
Incomplete reaction phenomenon	$\neq$	=	=	=	$\neq$	$\neq$	$\neq$	$\neq$
Necessarily has a glissile interface	=	=	=	=	=	$\neq$	$\neq$	$\neq$
Always has a rational orientation with parent γ	=	=	=	=	=	$\neq$	$\neq$	$\neq$
Grows across austenite grain boundaries	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$	=	=	=
High interface mobility at low temperatures	=	=	=	=	=	$\neq$	$\neq$	$\neq$
Displacive transformation mechanism	=	=	=	=	=	$\neq$	$\neq$	$\neq$
Diffusional transformation mechanism	$\neq$	$\neq$	$\neq$	$\neq$	$\neq$	=	=	=
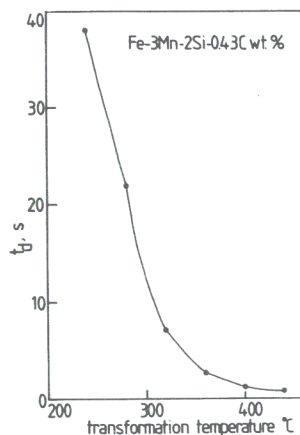
1 Time for decarburisation of Bainitic ferrite.
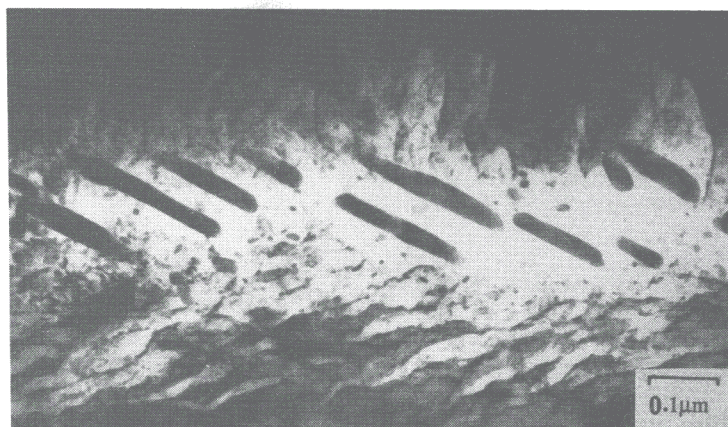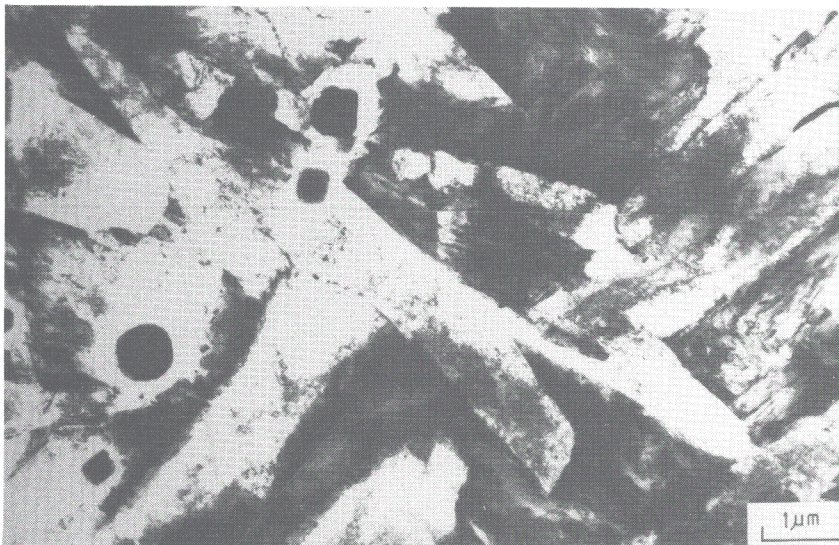

2 Slow diffusional growth of perturbations on lower bainitic ferrite.
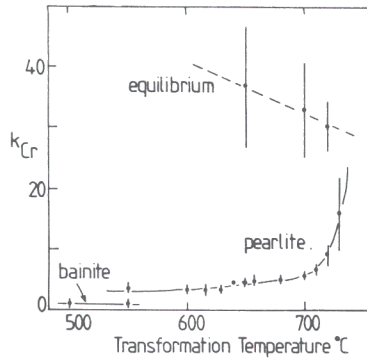
3 Composition of cementite
(Chance and Ridley).

a   b
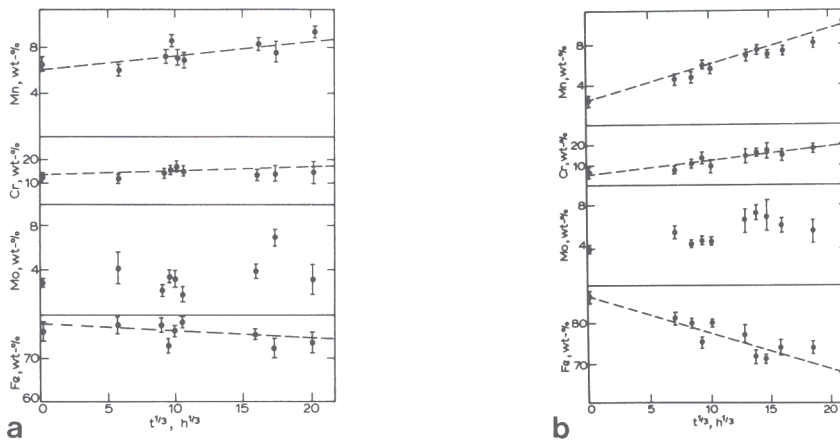
4 Composition changes in bainitic cementite during
tempering: (a) service exposed material,
(b) reheat-treated material. (Afrouz *et al.*).

5 Morphology of
Acicular Ferrite.