

THE MECHANISM OF BAINITE FORMATION IN STEELS

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Abstract—New microstructural evidence, together with a thermodynamic analysis, of the bainite reaction in steels are presented in support of a growth mechanism involving the propagation of displacive sub-units. The results are shown to be consistent with a displacive mechanism, rather than with one involving reconstructive growth. It is also possible to account not only for sympathetic nucleation, which is a characteristic of ferrous bainites, but also to explain the nature of the incomplete reaction phenomenon. Furthermore, the analysis suggests that much of the partitioning of carbon into the residual austenite occurs subsequent to transformation, rather than during the growth process. An explanation is also developed for the limited size of the displacive sub-units in terms of the friction stress associated with the motion of the transformation interface.

Résumé—Nous présentons des observations microstructurales nouvelles et une analyse thermodynamique de la réaction bainitique dans les aciers, qui témoignent en faveur d'un mécanisme de croissance mettant en jeu la propagation de sous-unités. Les résultats sont en accord avec un mécanisme de déplacement plutôt qu'avec un mécanisme qui mettrait en jeu une croissance reconstructive. On peut rendre compte non seulement de la germination induite, qui caractérise les bainites de fer, mais également des phénomènes de réaction incomplète. De plus, l'analyse suggère que la répartition du carbone dans l'austénite résiduelle se produit après la transformation plutôt qu'au cours de la croissance. Nous proposons une explication pour la taille limitée des sous-unités, à partir des forces de frottement associées au déplacement de l'interface de transformation.

Zusammenfassung—Es waren neue mikrostrukturelle Hinweise für die Bainitreaktion zusammen mit einer thermodynamischen Analyse vorgelegt, die einen Wachstumsmechanismus durch Wanderung verschiebbarer Untereinheiten unterstützen. Es wird gezeigt, daß die Ergebnisse mit diesem Verschiebungsmechanismus verträglich sind, jedoch nicht mit dem des rekonstruktiven Wachstums. Nicht nur die für Eisenbainit typische mitlaufende Keimbildung, sondern auch die Natur der unvollständigen Reaktion kann in diesem Modell erklärt werden. Des weiteren legt die Analyse nahe, daß ein großer Teil der Verteilung des Kohlenstoffes in den restlichen Austenit anschließend an die Umwandlung auftritt, also nicht während des Wachstumsprozesses. Schließlich wird die begrenzte Größe der verschiebbaren Untereinheiten anhand von einer bei der Bewegung der Umwandlungsgrenzfläche auftretenden Reibungsspannung erklärt.

NOMENCLATURE

c = Semi-thickness of oblate spheroid representing bainite sub-unit.
 r = Radius of oblate spheroid representing bainite sub-unit.
 R = Gas constant.
 T = Absolute temperature.
 T_0 = Equilibrium temperature for stress free austenite and ferrite of identical composition.
 E = Standard state constant for ferrite.
 $\Delta\bar{H}_{\alpha,\gamma}$ = Partial molar heat of solution of carbon in ferrite, austenite
 x = mole fraction of carbon in ferrite and in austenite.
 Y = mole fraction of substitutional alloying element in austenite or in ferrite, disregarding the presence of carbon.
 \bar{w} = Weighted average of the pairwise interaction energy of carbon atoms in the alloyed austenite.
 ΔT_{magi} = Displacement in the temperature at which the free energy change accompanying the austenite to ferrite transformation in pure iron is calculated in order to account for the changes (per atomic percent) due to alloying element effects on the magnetic and non-magnetic components of free energy change, respectively.

A = Strain energy factor.
 $\Delta F^{\gamma-\alpha}$ = Chemical free energy change accompanying the austenite to ferrite transformation in pure iron.
 $\Delta F^{\gamma-\alpha'}$ = Chemical free energy change accompanying the austenite to ferrite transformation in alloyed iron, given that the austenite and ferrite have identical compositions.
 V_B = Volume fraction of bainitic ferrite.
 x_y = Weight percent of carbon in austenite.
 τ = Friction stress associated with the displacement of the austenite-ferrite interface.
 $\Delta L/L$ = Length change per unit length of dilatometric specimen.
 s = Weight percent of carbon remaining in bainitic ferrite at the termination of transformation.
 \bar{x} = Weight percent of carbon in the alloy as a whole.
 $\Delta S_{\alpha,\gamma}$ = Partial molar non-configurational entropy of solution in ferrite, austenite.

INTRODUCTION

The mechanism of formation of the ferritic component of bainite in steels remains controversial [1-4]; the conflict primarily arises in the choice

between the displacive and reconstructive modes of solid state transformation. The former mechanism implies the existence of an atomic correspondence across the transformation interface whereas the latter allows diffusion controlled processes to destroy any such correspondence.

By the displacive mechanism [1], bainitic ferrite is considered to grow by the propagation of discrete martensitic sub-units, the aggregate of these sub-units constituting the classical sheaf of upper bainite. The sheaf itself has a plate like morphology and its formation involves sympathetic nucleation of successive sub-units. The bainitic ferrite is considered to form with a carbon super-saturation, with partitioning of the carbon into the residual austenite occurring immediately after formation. The width of the sub-units observed by Oblak and Hehemann [1] was *ca.* 0.5 μm , with a length of *ca.* 10 μm .

On the other hand, Aaronson and Kinsman [2] maintain that the observed sub-units are in fact ledges propagating along the broad faces of bainite plates. Based on thickening rates obtained from the *in situ* thermionic emission microscopy observations following the work of Rouze and Grube [5, 6], they calculated an expected inter-ledge spacing of *ca.* 0.5 μm which was claimed to be consistent with the 0.3–8.0 μm ledges observed in the thickening of proeutectoid ferrite plates in the Fe–C system [7]. They therefore further claimed a direct correspondence between ledges and sub-units. Additionally, they suggest that if the ledge mechanism is accepted, the apparent lengthening kinetics are consistent with a diffusion controlled growth mechanism. Aaronson and Kinsman [8] have also presented a thermodynamic analysis which was claimed to invalidate the displacive sub-unit mechanism.

Bhadeshia and Edmonds [3] recently undertook an examination of the structure of bainite in a silicon steel and found that the sheaf of upper bainite (which had the appearance of a plate on the scale of optical microscopy) in fact consisted of smaller sub-units (*ca.* 0.2 μm wide). Furthermore, they showed that these could not possibly be compared with ledges since the individual sub-units were, for the most part of their interfaces, isolated by carbon enriched retained austenite films.

In the present paper, the authors hope to resolve some of the apparent discrepancies connected with the displacive mechanism. New microstructural observations together with a detailed thermodynamic analysis are presented in support of the displacive mechanism and to evaluate the details of the mechanism itself.

EXPERIMENTAL

A 65 g melt of the experimental alloy was prepared from pure constituents in an argon-arc furnace, and the final chemical analysis was Fe–0.43C–3.00Mn–2.12Si. The ingot was sealed in a quartz capsule under

a partial pressure of argon and homogenized for three days at 1250°C, before hot-swaging to 3 mm diameter rod.

Austenitizing (1200°C for 10 min) was carried out with the specimen sealed in a quartz capsule. Subsequent isothermal transformation was carried out by fracturing the quartz and quenching the specimen into a tin bath (286°C for 30 min) covered with a layer of active charcoal before finally quenching into water.

Thin foil specimens for transmission electron microscopy were prepared from 0.25 mm thick discs slit from the heat-treated 3 mm diameter rod under conditions of flood lubrication. The discs were subsequently thinned and electropolished in a twin-jet polishing unit using a 25% glycerol, 5% perchloric acid and 70% ethanol mixture at room temperature. The foils were examined in a Philips EM300 transmission electron microscope operated at 100 kV.

RESULTS AND DISCUSSION

(a) Microstructural observations

It was decided to attempt the examination of a complete sheaf of upper bainite by electron microscopy, with the aim of observing the detailed arrangement of sub-units within a given sheaf and also characterising the tip of the sheaf in order to be able to compare the events at the apparent transformation front with those near the original sheaf nucleation site. It was additionally hoped to gain information on the thickening of the sheaf, the nature of sympathetic nucleation and the morphology of individual sub-units.

The normal foil preparation technique had to be modified since the examination described above necessitated a very large electron transparent area on the thin foil. The conventional technique did not give sufficient thin area and also suffered from electron beam induced foil buckling. However, by adjusting the electropolishing conditions such that the bainite sheaf was preferentially attacked relative to the martensite matrix, it proved possible to examine a complete sheaf, with the relatively thicker martensite matrix providing foil support to prevent buckling.

Figure 1 is a montage of electron micrographs of an isolated sheaf and the following points are to be noted:

1. The sheaf consists of an arrangement of sub-units such that the aggregate itself has a plate like morphology on a macroscopic scale, as is also obvious from the accompanying optical micrograph.
2. Observation of a large number of thin foils indicated that the subunits are themselves plate shaped.
3. The sub-units exhibit pronounced lenticularity, a feature most prominent near the edge of the sheaf, where their sharply pointed tips are also apparent.
4. Consistent with previous observations [3], the sub-units were found to be separated by films of retained austenite and are in the same crystallographic orientation relative to the austenite.

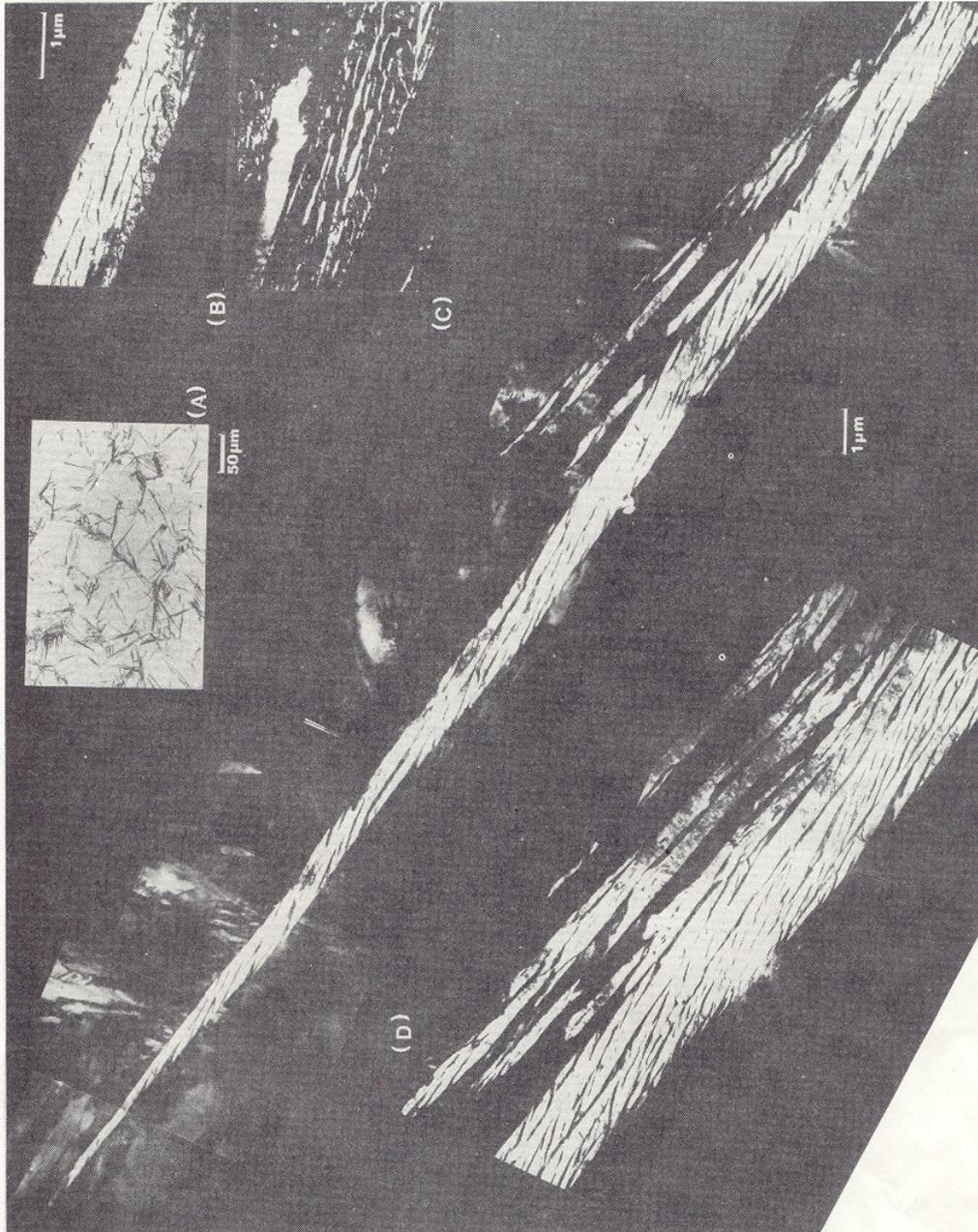


Fig. 1. (A) Optical micrograph showing upper bainite sheaves in a martensite matrix. (B) Electron micrograph of a sheaf of upper bainite. (C) Corresponding retained austenite dark field image. (D) Montage of electron micrographs illustrating morphology and substructure of a sheaf of upper bainite.

5. The apparent width of the sub-unit at the tip of the sheaf is approximately the same as that near the original nucleation site.

6. Examination of the regions near the tip of the sheaf indicate that the new sub-units are sympathetically nucleated near the tips of the prior sub-units.

(b) *Thermodynamic analysis*

Assuming that the free energy surface of austenite can be reasonably extrapolated into the temperature regions where the austenite is not in thermodynamic equilibrium, a T_0 temperature can be defined [9] such that stress free austenite and ferrite of the same composition (with respect to both the interstitial and the substitutional alloying elements) are in metastable equilibrium. Thus, any displacive transformation involving a full supersaturation of carbon can occur only below the appropriate T_0 temperature.

An advanced model for the calculation of the T_0 temperature in steels was presented by Aaronson *et al.* [10] and was subsequently corrected by Shiflet *et al.* [11]. This model is based on the Lacher [12] and Fowler and Guggenheim [13] formalism and resolves the problem of overlapping regions of influence of individual interstitial atoms and is considered to be reliable at low temperatures [10]. The model can be adapted for alloy steels by the methods of [8, 14]. However, the T_0 equations of Shiflet *et al.* contain a serious assumption that the activity of carbon in ferrite can be represented by an approximation (equation 8, Ref. [11]) which requires the carbon content of the ferrite to be small. This is not valid for the purposes of T_0 evaluation, in which the ferrite is supposed to inherit the full carbon content of the austenite. Hence Bhadeshia [15] has recently modified the equations of [11], using the exact activity expressions given by [11], thus obviating any prior assumption about the carbon content of the ferrite. The equations of [15] also take account of Zener ordering [16].

According to [15], the T_0 temperature (when element partitioning does not occur) is given by $\Delta F^{\gamma-\alpha'} = 0$, where

$$\begin{aligned} \Delta F^{\gamma-\alpha'} = & 2xRT \ln x + x(\Delta H_x - \Delta H_\gamma) \\ & - (\Delta S_x - \Delta S_\gamma)T + 4\omega_x - 6\omega_\gamma \\ & - 4RT(1-x)\ln(1-x) \\ & + 5RT(1-2x)\ln(1-2x) \\ & - 6RTx \ln \left| \frac{\delta_\gamma - 1 + 3x}{\delta_\gamma + 1 - 3x} \right| \\ & - 6RT(1-x) \ln \left| \frac{1 - 2\tau_\gamma + (4\tau_\gamma - 1)x - \delta_\gamma}{2\tau_\gamma(2x - 1)} \right| \\ & + 3RTx \ln(3-4x) + 4RTx \ln \left| \frac{\delta_\alpha - 3 + 5x}{\delta_\alpha + 3 - 5x} \right| \\ & + (1-x) \left| 141 \sum_i Y_i (\Delta T_{magi} - \Delta T_{nmi}) \right. \\ & \left. + \Delta F^{\gamma-\alpha} \left\{ T - 100 \sum_i Y_i \Delta T_{magi} \right\} \right| + \Delta f^* \quad (1) \end{aligned}$$

where

$$\delta_\alpha = |9 - 6x(2\tau_x + 3) + (9 + 16\tau_x)x^2|^{1/2}$$

$$\delta_\gamma = |1 - 2(1 + 2\tau_\gamma)x + (1 + 8\tau_\gamma)x^2|^{1/2}$$

$$\tau_x = 1 - \exp(-\omega_x/RT)$$

$$\tau_\gamma = 1 - \exp(-\omega_\gamma/RT)$$

Δf^* = Energy term due to Zener Ordering, evaluated as in [16], ω_γ was calculated using the procedure of [8, 11] and the optimised activity data of [17], ω_x has been recently evaluated by [18], using the accurate and self-consistent activity data of [19]. The other parameters have been recently determined as

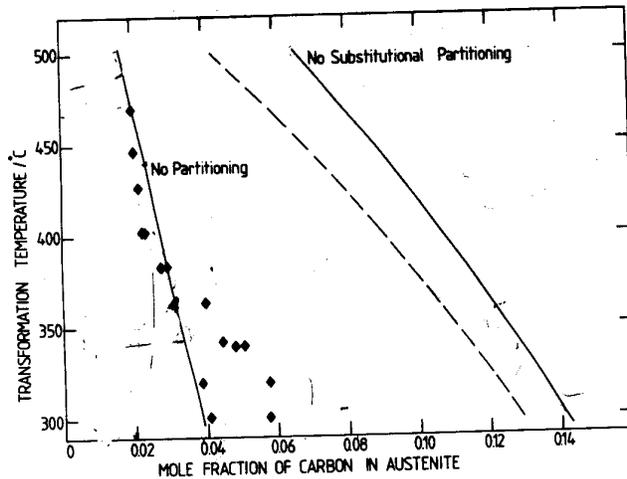


Fig. 2. Plot of temperature vs carbon content in austenite. See text for detailed explanation.

follows:

$$\Delta H_a = 112212 \text{ J mole}^{-1} \quad \text{Ref. [19]}$$

$$\Delta S_a = 51.5 \text{ J mole}^{-1} \text{ K}^{-1} \quad \text{Ref. [19]}$$

$$\Delta H_f = 38575 \text{ J mole}^{-1} \quad \text{Ref. [11]}$$

$$\Delta S_f = 13.48 \text{ J mole}^{-1} \text{ K}^{-1} \quad \text{Ref. [11]}$$

Similarly, the equation for the no-substitutional element partitioning curve (Ae'_3) is given by (11) as

$$141 \sum_i Y_i (\Delta T_{\text{magi}} - \Delta T_{\text{nmi}}) + \Delta F^{r-a} \left\{ T - 100 \sum_i Y_i \Delta T_{\text{magi}} \right\} \\ = 5RT \ln \left| \frac{1-x}{1-2x} \right| + 6RT \ln \left| \frac{1-2\tau_\gamma + (4\tau_\gamma - 1)x - \{1 - 2(1+2\tau_\gamma)x + (1+8\tau_\gamma)x^2\}^{1/2}}{2\tau_\gamma(2x-1)} \right|. \quad (2)$$

In both the above equations, values of ΔF^{r-a} were obtained from the data of Ref. [20]. The calculations were conducted for the alloy used in the experiments of Ref. [3] i.e. 0.43C–3Mn–2.12Si–Fe (wt%). Figure 2 shows the results obtained.

The analysis was taken further in an attempt to understand the nature of the 'incomplete reaction phenomenon' [2, 3]. In this, the bainite reaction stops after a certain amount of transformation at a given temperature and ideally reaches zero transformation at the B_s temperature. Bhadeshia and Edmonds [3], using dilatometry backed by extensive optical and electron microscopy showed that the 0.43C–3Mn–2.12Si–Fe alloy precisely exhibited the incomplete reaction phenomenon described above.

At 363°C, quantitative metallography showed a maximum of 52% bainitic ferrite. Using this as a calibration datum, the dilatometric data of Ref. [3] (which has been thoroughly confirmed by microscopy) was analysed as follows:

If $\Delta L/L$ is the length change, then at 363°C, $\Delta L/L = 2.74 \times 10^{-3}$ with the volume fraction of bainitic ferrite, $V_{B_{363}} = 0.52$. Hence for any other temperature, $V_{B_T} = (0.52/2.74 \times 10^{-3}) (\Delta L/L)_T$ (3). Since it has been shown [3] that the final carbon content of the bainitic ferrite is ca. 0.03 wt%, the carbon content of the austenite is given by

$$x_{\gamma_f} = \bar{x} + \frac{V_B(\bar{x} - s)}{1 - V_B} \quad \begin{matrix} \bar{x} = 0.43 \\ s = 0.03 \end{matrix} \quad (3)$$

Using these equations and the dilatometric data, the carbon content of the austenite at the termination of reaction are also plotted in Fig. 2, as a function of the transformation temperature. The significance of these results will be considered later.

(c) Strain and interfacial energy contributions

The calculated curves using equations (1, 2) do not account for non-chemical components such as interfacial and strain energies. We may compute the strain energy associated with the final platelet as follows:

Considering the sub-unit as an oblate-spheroid of

radius r and semi-thickness c , the strain energy per unit volume is given by Ac/r (Ref. [21]) where A is of the order of 2090 J cm^{-3} [22, 23] at 25°C, $c \approx 10^{-7} \text{ m}$, $r \approx 5 \times 10^{-6} \text{ m}$. Since the value of A quoted above refers to 25°C, and since A is a linear function of the shear modulus of the matrix (24), we can calculate its magnitude at 470°C (B_s) on the basis that the shear modulus decreases by 0.03% per 1°C

rise in temperature [25] to give $A_{470} \approx 1815 \text{ J cm}^{-3}$ or 12910 J mol^{-1} . Thus $Ac/r = 260 \text{ J mol}^{-1}$.

It is important to note that this figure could be significantly reduced if plastic relaxation of elastic strains occurs, [26], as is the case for bainite. The displacement of the transformation interface should, in principle, leave a defect free single crystal region [4] in the absence of periodic twinning or faulting associated with the relevant lattice invariant shears. Nevertheless, both bainitic ferrite and its associated retained austenite have high dislocation densities [3]; these must arise from plastic relaxation effects.

Taking the interfacial energy to be of the order of $0.2\text{--}1.0 \text{ J m}^{-2}$ [22, 27, 28], and using the platelet dimensions quoted earlier, the strain and interfacial energy contributions would amount to approximately 270 J mol^{-1} . Thus the dashed curve in Fig. 2 represents the non-substitutional partitioning curve after allowing for the above value of stored energy.

(d) Further analysis

Considering now the data of Fig. 2, it is evident that the degree of transformation to bainitic ferrite always remains well below that allowed by the $A_{e'_3}$ curve. The results are consistent with the hypothesis that only a very limited amount of carbon partitions into the austenite during transformation, so that the original bainitic ferrite retains much of the carbon content of the parent austenite. The bulk of the partitioning of carbon into the residual austenite must occur after the initial formation event. If this represents the true nature of events, the bainite reaction can be expected to cease as soon as the austenite carbon content reaches a critical value, the magnitude of which would be near the T_0 curve, depending on the exact degree of partitioning occurring simultaneously with transformation. Such an interpretation would also explain the incomplete reaction phenomenon since, with decreasing temperature, the austenite can tolerate successively greater amounts of carbon before the formation of supersaturated ferrite becomes thermodynamically impossible.

An alternative interpretation could arise if it is considered that diffusional growth ceases when the driving force falls below some limiting value. However, in an earlier study [3] we reported that isothermal transformation at a small undercooling below the B_s temperature leads to the formation of a very limited quantity of bainitic ferrite. Continued holding at the isothermal transformation temperature does not lead to any increase in the amount of bainitic ferrite, but after a large time gap (20–30 h) in which no transformation seems to occur, pearlite formation follows. This latter reaction involves the uncoordinated transfer of atoms across the austenite/ferrite interface, in a manner similar to that envisaged by Aaronson *et al.* [2] for the transfer of atoms across the disordered risers of the ledges which are supposed to accomplish the diffusional formation of bainitic ferrite. Indeed, the diffusional school would have it that the bainite transformation is itself simply a degenerate pearlite reaction [29]. Hence the observation on the eventual formation of pearlite [3] supports the view that it is not a limitation on diffusional growth that causes the incomplete reaction phenomenon, since diffusional growth does not cease.

Furthermore, if the aggregate of retained austenite and bainitic ferrite obtained by isothermal transformation at low temperatures ($\sim 300^\circ\text{C}$) is subsequently tempered at 500°C , the austenite diffusively transforms to ferrite and carbides [1]. This is despite the fact that a lower driving force is available for diffusional transformation at higher temperatures, and this not only demonstrates the absence of any limitation to diffusional growth, but stresses the point that there is insufficient thermal activation for diffusional growth at the temperatures of bainite formation. Finally, it should be noted (see Ref. [1]) that the composition of the steel used in the present study is not conducive to the operation of a solute drag effect of the type envisaged by Aaronson [29]. In any case, it is most unlikely that any such effect could be strong enough to completely overcome the transformation driving force and cause zero reaction at the B_s temperature.

In view of the above considerations, it is felt that the incomplete reaction phenomenon observed in the present study is a manifestation of the formation of essentially supersaturated bainitic ferrite, which only subsequently partitions the bulk of its carbon content into the residual austenite.

It is important to note that the use of equation (4) is not valid for those bainite reactions involving carbide precipitation since the latter will necessarily reduce the carbon content of the system, thereby allowing further reaction. In general, the partitioning of carbon into the austenite and the precipitation of carbides (either subsequent precipitation at the interface as in upper bainite or precipitation from a supersaturated matrix as in lower bainite) are competitive processes, the relative rates depending on alloy chemistry (in terms of carbide inhibitors for example) and

transformation temperatures. In such cases the incomplete reaction analysis can only be applied if the carbide precipitation events can be quantitatively accounted for. This is evidently one of the reasons why the incomplete reaction effect is not well pronounced in steels where cementite precipitation dominates, as in plain carbon steels.

In the incomplete reaction analysis it was implicitly assumed that the partitioned carbon is uniformly distributed throughout the residual austenite. Experimental observations [3] show that this is not the case since austenite films entrapped between neighbouring sub-units have a higher carbon content than the islands of residual austenite resulting from the geometrical partitioning of the prior austenite grain by crystallographically different variants of bainite sheaves. This effect is reflected in the martensitic decomposition of the islands of residual austenite when quenching to room temperature from the isothermal transformation temperature whereas the entrapped films are fully retained [3]. Thus, even with $V_B < 0.005$, the films of retained austenite have an X-ray determined carbon content of ~ 0.8 wt% (30) despite the fact that equation (4) would indicate an extremely small average austenite enrichment.

The effect of the inhomogeneous distribution of carbon would be to allow the transformation to proceed to an extent somewhat greater than that allowed by the thermodynamic condition based on a uniform carbon assumption. It is estimated, however, that such an effect would be small since the probability of adjacent sub-unit formation is relatively small. Electron microscopy indicates that the sub-units nucleate sympathetically near the tips of the prior sub-units. This would enable the growth of the new sub-unit to be essentially unaffected by the carbon diffusion field (arising from the post-formation partitioning) of the previous sub-unit. Such growth into relatively lower carbon austenite would clearly be energetically more favourable. Indeed, the carbon concentration near the tip of the prior sub-unit would be lower than that near the broad faces since the shape lenticularity not only implies a higher interface to volume ratio at the tip, but the very geometry of the tip leads to a small carbon concentration per unit interfacial area. The latter effect originates from the fact that a positively curved surface will have a greater volume of austenite per unit distance from the interface compared with a relatively planar interface. It is only when the carbon pile-ups on the broad faces level out that adjacent sympathetic nucleation becomes feasible. The latter would take a relatively longer time, thereby favouring tip nucleation and directly leading to the apparent plate morphology for the sheaf of sub-units as a whole.

(e) Surface relief

Examination of pre-polished, austenitised and isothermally transformed specimens revealed that surface displacements could be associated with the upper

bainite (Fig. 3). The uniformity of interference contrast on the tilted surfaces illustrated in Fig. 3b indicates the invariant plane strain nature of the surface relief. Such displacements are a characteristic of martensitic transformations and are taken to imply that the sub-units propagate in a displacive manner.

In view of this conclusion and the earlier deduction that bainitic ferrite, when first formed, retains much of the parent lattice carbon content, it is of interest to consider why bainite can form at a lower undercooling relative to martensite, which in the present alloy has an M_s temperature of 220°C [3]. It seems that martensite proper requires a high supercooling in order to overcome the high strain and interfacial energies of the fault sequences associated with the spontaneous nucleation of martensite [22, 31–33]. However, this need not apply to the bainite discussed above. The first displacive sub-units to form nucleate at the prior austenite grain boundaries and subsequent subunits, of course, nucleate sympathetically. However, this requires further elaboration since austenite grain boundaries are often taken to stimulate martensite formation as well.

It is proposed that a high angle austenite grain boundary must have an influence on nucleation in a manner other than just acting as a source of defects appropriate to various martensite nucleation theories. For instance, a high angle grain boundary (with its high entropy level) has a relatively large free volume relative to the defect free lattice. Given sufficient thermal activation, it seems conceivable that local grain

boundary atomic configurations arise that may be in some way similar to a ferrite lattice. This would tend to aid b.c.c. nucleation. Bearing in mind that thermal activation is a time-temperature parameter, it is clear that the necessary activation could be compatible with the temperatures at which bainite forms although it need not be available during the rapid quench necessary to form martensite proper. Indeed, in the experience of the authors, half-lens shaped martensite plates are not generally observed to emanate from austenite grain boundaries in high alloy (low- M_s) steels, as would be expected if the grain boundary were the true nucleation site. Additionally, we note that Widmanstätten ferrite (which shows an IPS shape change) does have a half-lens shape when it nucleates from austenite grain boundaries.

It seems that even if the austenite boundary may act as a source of appropriate defects, it need not act as a site for nucleation, as is presently envisaged for the bainite transformation.

(f) Growth termination mechanism

With respect to the limited size of the displacive sub-units, it is noted that a stress is required to drive the interface between the ferrite and austenite in order to accomplish the necessary transformation strain [34]. This stress, τ , is the friction stress and the forces necessary to overcome τ during transformation are provided by the chemical free energy change. Under circumstances when the transformation is perfectly thermoelastic, τ does not increase with the pro-

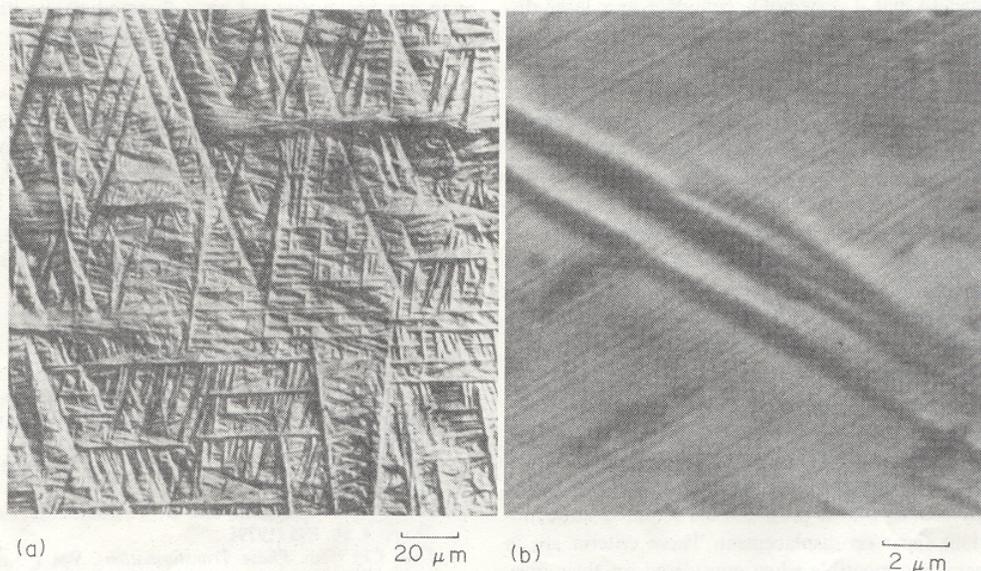


Fig. 3. (A) Nomarski differential interference contrast image showing the general surface displacements due to upper bainite formation in a Fe-3Mn-2.12Si-0.43C alloy. The specimen was pre-polished, sealed in a quartz tube containing a partial pressure of argon and isothermally transformed at 350°C. (B) As above, but showing a high-magnification image of isolated bainite units in an austenitic matrix. The invariant-plane strain character of the relief is evident from the uniformity of the contrast at the tilted regions. The apparent fringes are thought to be due to the use of an oil immersion lens and should not be confused with the interference fringes associated with two-beam interferometry.

gress of transformation. However, when the transformational shape change is plastically accommodated†, the elastic strain energy is relaxed and the frictional stress is greatly increased [34, 35] due to interaction of the interface with accommodation defects to a point where the chemical free energy change will no longer be adequate to further drive the transformation. The interface is thus immobilised by dislocation jamming [35]. Bhadeshia and Edmonds [3] have shown extensive dislocation tangles in the vicinity of the bainitic ferrite-austenite interface, thus providing direct evidence for the above size limiting mechanism.

Additionally, the interfacial diffusion coefficient of carbon at the temperatures where bainite forms will be sufficiently high to allow a Cottrell type atmosphere to form at the transformation interface. This could then stop, or at least hinder the transformation when some critical level of carbon concentration is achieved. The latter would not only pin the interface but would locally increase the carbon content, thereby reducing the available chemical driving force. This would be consistent with the general observation that the sub-units are larger when the average alloy carbon content is lower since the time taken to reach a critical concentration for immobilisation would be greater.

(g) *The nature of the bainite-austenite interface*

The crystallographic theory of martensite is based on the postulate that the habit plane is an invariant plane relative to the macroscopic shape deformation. The parent-product mismatch at the habit plane is considered to be periodically corrected by discontinuities so that it does not accumulate over large distances [4]. These structural discontinuities are needed only to correct the mismatch at the semi-coherent (in general) *planar* boundary. However, for lenticular plates (as observed in the present study) exhibiting a curved boundary this interface is glissile since the additional misfit regions (other than the structural discontinuities) arising from the boundary curvature constitute the transformation dislocations of the martensite pole mechanism [36]. Since the sub-units observed in the present work are indeed lenticular, their interfaces with austenite must also be glissile, in accord with a displacive transformation mechanism.

There are at least three possible criteria [37] for the selection of the interfacial array of mismatch dislocations under transformation conditions; (a) minimum interfacial strain energy, (b) minimum force required to displace the dislocation array, and (c) the array is such that it produces the required lattice invariant shear on displacement. These criteria are in general incompatible when considered on their own. However, we recall that the interfacial energy is only

† In accordance with [35], plastic accommodation refers to all processes beyond those prescribed by the phenomenological theories to establish an invariant-plane condition, i.e. mainly complex slip.

one aspect of the complete energy balance involved in transformation thermodynamics and a relatively small one at that. Hence the choice of any particular criterion must not be made solely on the basis of interfacial energy. It must be borne in mind that post-transformation examination of static interfaces need not be reliable in the case of high temperature transformations (as in the case of [38, 39] as there will exist a driving force for any glissile transformation interface to equilibrate into an energy cusp orientation following transformation. Such criticism also applies to geometrical modelling based solely on minimum interfacial energy considerations [38, 39].

Furthermore, it should be noted that the concept requiring an increase in interfacial energy in order to accomplish transformation is not alien to the ledge mechanism. The latter requires that the riser of a ledge has a disordered structure, thus conferring the necessary mobility. However, Aaronson *et al.* [40] suggest that if the riser does not have a disordered structure, then higher order ledges or jogs must be formed on the riser in order to provide an interfacial structure sufficiently disordered to allow an adequately rapid atom transfer rate. Thus a net increase in interfacial energy is allowed for the sake of an overall reduction in the free energy of transformation.

CONCLUSIONS

It has been shown that the mechanism for bainite formation is consistent with growth by the propagation of essentially supersaturated displacive sub-units. Furthermore, the incomplete reaction phenomenon can be understood on a thermodynamic basis, as can the morphology of the bainitic sheaves.

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