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COMMENTS ON THE PAPER "THE STRUCTURE OF THE BROAD FACES OF FERRITE PLATES"

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Introduction

In their note (1) on the structure of Widmanstätten-ferrite/austenite interfaces, Rigsbee et al. misquote the paper by Bhadeshia and Edmonds (specifically, lines 5-11, 2nd column, p.1272, Ref.2). Hence the claim that "Bhadeshia has proposed a transformation of the interfacial dislocation structure from a glissile array at the isothermal reaction temperature to a sessile array at room temperature", and other such claims referring to (2) are incorrect and are not considered further. The comments (1) in the section "Other Glissile Interface Arguments" are discussed below.

The Widmanstätten-Ferrite/Austenite Interface

The nature of interfaces has been thoroughly discussed by Christian and Crocker (3-5); with the help of this work, we proceed to demonstrate that the microstructural observations of Rigsbee and Aaronson (6) can be interpreted to imply that the Widmanstätten-ferrite/austenite (α/γ) interface is glissile. Throughout this paper, the terms 'interface' and 'interface plane' refer to the average interface plane as determined on a macroscopic scale.

A semi-coherent interface containing a single array of intrinsic dislocations is considered to be glissile when the dislocations are able to move conservatively as the interface migrates. The intrinsic dislocations must therefore all be pure screw dislocations, or have Burgers vectors which do not lie in the interface plane. The α/γ interface seems to contain discrete intrinsic dislocations (6) with Burgers vectors ($\underline{b}$) parallel to $(1\ 1\ 1)_\gamma \| (1\ 1\ 0)_\alpha$ – for the remaining discussion, all crystallographic data are referred to the particular α plate which has $(1\ 1\ 1)_\gamma \| (1\ 1\ 0)_\alpha$, $[\bar{1}\ 1\ 0]_\gamma \sim\| [\bar{1}\ 1\ \bar{1}]_\alpha$, the interface plane being $(15\ 21\ 9)_\gamma$ with $\underline{b} \| [\bar{1}\ 1\ 0]_\gamma$, consistent with the first complete set of data given in Table 3 of (6). For these particular data, the interface plane is some 18° from the $(1\ 1\ 1)_\gamma$ and $\underline{b}$ clearly does not lie in the interface plane, so that the intrinsic (or "misfit") dislocations can move conservatively as the interface migrates. This conclusion is true for all the data given in (6). Rigsbee et al. (1) and Rigsbee and Aaronson (6) consistently misinterpret this point by claiming that since $\underline{b}$ lies within the $(1\ 1\ 1)_\gamma$ plane, the interface is sessile, despite the fact that the <u>interface plane</u> is without exception found not to be parallel to $(1\ 1\ 1)_\gamma$.

A glissile interface also requires that the glide planes (of the misfit dislocations) associated with the α lattice must meet the corresponding glide planes in the γ lattice edge to edge in the interface, along the dislocation lines (3). This condition is also satisfied for Widmanstätten-ferrite since the $(1\ 1\ 0)_\alpha$ planes are parallel to the $(1\ 1\ 1)_\gamma$ planes (6).

Since the observed misfit dislocations are all perfect lattice dislocations (6), their glide with the interface is expected to inhomogeneously shear the volume of material swept by the interface without altering the nature of the parent or product lattices, like the lattice-invariant deformation of the phenomenological theories of martensite crystallography. Hence the misfit dislocations cannot be associated with the transformation from γ to α. The α/γ interface consists of stepped planar sections of $(1\ 1\ 1)_\gamma$ planes and the steps have a monoatomic or triatomic height (6). Steps like this can be generated by a series of virtual operations (3,5) which prove that they have a dislocation character with an associated strain field; they could not otherwise be imaged using strain field contrast in the transmission electron microscope. Furthermore, since strain field contrast is observed (6), these particular steps cannot be pure steps. It follows that the Burgers vector associated with each of the discrete (resolvable) steps cannot be a lattice repeat vector. The movement of each of these steps must thus alter the nature of the parent lattice and it is reasonable to call such steps transformation (3) or coherency (7) dislocations. The motion of these steps in the $(1\ 1\ 1)_\gamma$ plane does not lead to the creation or destruction of lattice sites, so that the coherency dislocations are glissile (3,7). Even if their movement causes changes in the interface structure (1), such changes must be periodic and would only lead to a frictional resistance like the Pierls-Nabarro force. Zonal dislocations in twinning suffer from the same problem but are still glissile (3).

The failure to recognise the dislocation character of the steps has another important consequence. There exist α/γ interface models (8-10) in which the orientation relationship between the α and γ lattices is fixed (with say $(1\ 1\ 1)_\gamma \| (1\ 1\ 0)_\alpha$) and the coherency of an interface parallel to $(1\ 1\ 1)_\gamma$ is examined. Attempts are then made to maximise this coherency by introducing steps of atomic height; this procedure ignores the energy of the steps, which are the coherency dislocations discussed above. It is therefore not established that rotating the interface plane away from $(1\ 1\ 1)_\gamma$ actually reduces the total energy of the α/γ interface. For any given orientation relationship, these models (8-10) find only one "low-energy" interface plane (within some 20° of $(1\ 1\ 1)_\gamma$) and hence do not explain the characteristic wedge shape of single crystal Widmanstätten-ferrite plates.

The Shape Change

Case 1: The interpretation of the α/γ interface in terms of coherency and misfit (or anti-coherency, Ref.7) dislocations is consistent with the fact (11) that the formation of Widmanstätten-ferrite plates leads to a change in shape of the transformed region, a change which is macroscopically an invariant-plane strain with a substantial shear component (such a shape change is henceforth referred to as an IPS). The lattice deformation caused by the motion of the coherency dislocations, when combined with the lattice invariant shear due to the anti-coherency dislocations must give a macroscopic shape change which is an IPS. The Widmanstätten-ferrite transformation is envisaged to be a displacive transformation with the morphology controlled by the need to minimise the strain energy associated with the shape change. This also implies the existence of an atomic correspondence (substitutional atoms) across the transformation interface.

On the other hand, for a diffusional transformation involving the unco-ordinated transfer of atoms across the interface, an atomic correspondence would not exist between the parent and product lattices. Accordingly, an IPS shape change would not accompany transformation and the morphology of the product need not be plate shaped. Allotriomorphic ferrite in steels grows diffusionally. Like Widmanstätten ferrite, it can grow without the partitioning of any substitutional alloying elements during transformation, while maintaining a rational orientation relationship with the parent austenite. Despite these similarities with Widmanstätten-ferrite, allotriomorphic ferrite significantly differs in that its shape is not in the form of a thin plate and its formation does not lead to an IPS shape change in the transformed region. This reinforces the hypothesis that Widmanstätten ferrite is not a diffusional transformation product.

Case 2: Christian (12) was the first to demonstrate that it is geometrically possible to obtain an IPS shape change in a transformation where some orderly diffusion of substitutional atoms occurs, so that a partial atomic correspondence is maintained. He pointed out, however, that this would be unlikely since the atomic mobility necessary for the diffusion would lead to recrystallisation, and hence destroy the IPS, its associated strain energy and the atomic correspondence. In the absence of recrystallisation, the surface relief detected would be very similar to that of case 1, and the morphology of the Widmanstätten-ferrite would again be dominated by the need to minimise the strain energy associated with the IPS shape change - the α would again be thin plate shaped.

Case 3: It has been proposed (13) that an IPS surface relief arises even during a diffusional transformation, simply due to the existence of a sessile semi-coherent interface, although a mechanism for this is not stated. Hence, Widmanstätten-ferrite is claimed to be a diffusional transformation product, the plate shape of which is explained in terms of a general theory of precipitate morphology (14). According to this theory, the plate shape arises because a "substantial barrier to growth is present at one orientation of the interphase boundary". The effect of the strain energy due to the IPS shape change is totally ignored, as is the inconsistency with the data on allotriomorphic ferrite. It should be noted that the surface relief of Widmanstätten-ferrite is accurately known (11) to be an IPS and the relief can lead to a single tilt or a tent-shaped tilt (two adjacent invariant-plane deformations). For many other transformations, the surface relief is not accurately established; it is fascinating that in a case (15) where precise measurements of the relief associated with a true diffusional transformation have been made, the relief has been shown not to have the characteristics of an invariant-plane strain. Other difficulties have been discussed elsewhere (16).

From the points presented above, it seems that the experimental data on steels can best be understood in terms of the concepts of Case 1.

Significance of the Curved Interface

The interface models (8-10) mentioned earlier suggest that an interface parallel to the $(1\ 1\ 1)_{\gamma}||(1\ 1\ 0)_{\alpha}$ plane would not be the lowest energy interface orientation. A regularly stepped interface rotated away from this orientation is supposed to have a lower energy due to a higher degree of coherency in the facet sections. If the plate shape of Widmanstätten-ferrite is considered to arise because of the presence of a substantial barrier to the normal migration of this lower energy (and lower mobility) stepped interface, then this particular interface orientation must represent a prominent cusp in the Wulff plot. The observed curvature (6) of the γ/α interface casts doubt on the assumed (and necessary) lack of mobility of the interface plane, whatever the microscopic mechanism by which the curvature arises. If on the other hand, Widmanstätten-ferrite forms by a displacive transformation mechanism, then interface curvature is significant because it would lead to the presence of additional steps (transformation dislocations) in the glissile interface, beyond those already present in the planar interface.

Conclusions

In summary, it is believed that the experimental evidence for Widmanstätten-ferrite formation in steels can best be interpreted in terms of a displacive transformation mechanism.

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Addendum

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The terminology and crystallographic variant of Widmanstätten-ferrite used in this addendum is the same as above. Rigsbee et al. (17) in their reply continue to misquote Bhadeshia and Edmonds (2). The quote (2) "...there will exist a driving force for any glissile interface to equilibrate into an energy cusp orientation..." does not in any way imply transformations from glissile to sessile interfaces.

Considering now the other points in their reply, Rigsbee et al. (17) dispute our interpretation of the results of Ref.6, that the Widmanstätten-ferrite/austenite interface is glissile. This is because they apply the criteria for distinguishing glissile and sessile interfaces to just one component of the interface, the facets on $(1\ 1\ 1)_\gamma || (1\ 1\ 0)_\alpha$. The criteria must be applied to the interface as a whole and we note again that the interface plane deviates substantially from $(1\ 1\ 1)_\gamma$. Applying the criteria to just the $(1\ 1\ 1)_\gamma$ facets only proves that that particular element of the interface is, on its own, sessile with respect to normal migration. However, the interface as a whole is glissile because the $(1\ 1\ 1)_\gamma$ facets do not have to move to enable the interface to migrate; the normal migration of the interface is accomplished by the motion of the atomic height steps (or transformation dislocations) on $(1\ 1\ 1)_\gamma$ and by the migration of the intrinsic dislocations on $(1\ 1\ 1)_\gamma$. Both the steps and the intrinsic dislocations are glissile on $(1\ 1\ 1)_\gamma$, so that there is no restriction to the normal displacement of the interface as a whole.

In the third point that they raise, Rigsbee et al. (17) fail to appreciate the difference between the intrinsic interface dislocations which serve to accommodate misfit, and atomic steps (transformation dislocations) which accomplish the lattice deformation from γ to α. This is thoroughly discussed in (3) and we reiterate that the observed intrinsic dislocations (6) can only produce a lattice <u>invariant</u> deformation on interface migration.

It is not possible to deduce the magnitude of the frictional resistance (which is like the Pierels-Nabarro force) to the motion of the atomic steps but the transformation dislocations are glissile because their motion does not lead to the creation or destruction of lattice sites. The experimentally observed shape deformation is however clear, and indicates that the transformation dislocations are not only glissile but can also glide easily.

Both Rigsbee et al. and Dahmen (18) are incorrect in disputing the point (16) that it is possible to form α plates which have the same crystallographic orientation but different habit planes and shape deformations. In elementary terms, the α habit plane is irrational and there are thus 24 variants of this habit plane; the Nishiyama-Wasserman orientation relation (NW) however has only 12 variants. It follows that it is possible to find two α plates which have different habit planes (and shape deformations) but which can not be distinguished in terms of orientation measurements. Departures from NW do not substantially alter this concept since the two plates would then be very similarly oriented in space and would be separated only by very low angle boundaries. The ledge kinetics arguments (17) are not very relevant since the initial growth of the α plate can be treated in terms of the diffusion of carbon away from the tip of the growing plate (without ledges). In our interpretation, thickening on the superledge scale involves the growth of other adjacent plates (which may be in different orientations), and not the motion of any "growth ledges" (17). This also explains the so-called "complex relief morphologies" (17).

The explanation given in point 9 of (17) is surprising since it rather invalidates the argument that Widmanstätten-ferrite acquires its plate morphology as a result of the existence of low-energy interfaces. Allotriomorphic ferrite, despite being in an identical crystallographic orientation as Widmanstätten-ferrite does not develop a plate morphology.

In point 11 of (17), Rigsbee et al. again do not present a mechanism for the development of surface relief via their model. The statement "..partial or full coherency across the broad faces of the ledges permits preservation of the shape strain.." does not explain how the relief arises when the transformation is actually supposed to occur at disordered risers of superledges. If their mechanism is correct, then in contradiction to experimental evidence, annealing twins in FCC crystals should develop the same IPS surface relief as mechanical twins in these crystals. Both kinds of twins have the same crystallography and develop coherent interfaces on the same planes and yet the diffusionally formed annealing twins do not show the IPS relief.

The contradictions which indicate that the strain energy due to the IPS relief of Widmanstätten-ferrite is often ignored (point 12, Ref.17) have been dealt with in (16). Fig.4 and the discussion on p.372 of Ref.6 clearly demonstrate the curved interfaces detected between Widmanstätten-ferrite and austenite.

In summary, we suggest that none of the earlier conclusions of this paper are altered by the arguments presented in (17).

References

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