BAINITE : MOBILITY OF THE TRANSFORMATION INTERFACE

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Abstract. - Experimental evidence is provided to show that the interface responsible for the austenite to bainite transformation exhibits the essential characteristics of a glissile, semi-coherent interface. The experiments involve a comparison between the rate of formation of bainite, and the mobility of an interface whose motion does not lead to the formation of invariant-plane strain surface relief effects. It is demonstrated that true thermally activated motion of the latter kind of interface is extremely slow at the temperatures where bainite forms.

Introduction. - An FCC (face centered cubic) lattice cannot be homogeneously deformed into a BCC (body centered cubic) lattice by a strain which is an invariant-plane strain (IPS), and it follows (1) that these crystals cannot be joined by an interface which is fully coherent, except when the interfacial area involved is very small. For larger areas of contact, the structure of the interface will in general consist of coherent regions separated by periodic discontinuities which prevent the misfit in the interface plane form accumulating over large distances, in order to minimise elastic strains associated with the interface. There are two kinds of semi-coherency (1-3) - if the discontinuities discussed above are screw dislocations, or dislocations whose burgers vector does not lie in the interface plane, then the semi-coherency is of the type associated with <u>glissile</u> martensitic interfaces, whose motion is <u>conservative</u>, and leads to an IPS shape change. Alternatively, the dislocations which constitute the discontinuities may have burgers vectors which lie in the interface plane, not parallel to the dislocation line, giving "epitaxial" semi-coherency (1-3), Fig.1. Clearly, the normal displacement of such an interface necessitates the thermally activated climb of the misfit dislocations, so that the interface can only move in a <u>non-conservative</u> manner, <u>with relatively restricted mobility</u> at low temperatures.

The nature of the shape change that accompanies the motion of an epitaxially semi-coherent interface is more difficult to assess. As discussed by Christian (1-3), the upwards non-conservative movement of the boundary AB (Fig.1) to a new position C'D' would change the shape of a region ACDB of the parent crystal to a shape AC'D'B of the product phase. The shape change amounts to a uniaxial contraction normal to AB together with a shear component parallel to the interface plane (i.e., an invariant-plane strain). Because of the dislocation climb implicit in the process, the total number of atoms in regions ACDB and AC'D'B will not be equal, the difference being removed by diffusion <u>normal</u> to the interface plane. Atom movements are therefore necessary (at least) over a distance equal to that moved by the boundary, corresponding to the <u>thickness</u> of the transformed region. If this constitutes the only diffusional flux that accompanies interface motion, than the shear component of the shape change will

not be destroyed, and the transformation will exhibit the surface relief effects (and corresponding strain energy) normally associated with displacive transformations. The mobility will, of course, be limited by the climb process. This means (3) that the detection of an IPS surface relief effect, with a shear component, does not necessarily imply the existence of a glissile martensite type semi-coherent interface. However, Christian (3) has pointed out that since atoms have to migrate over large distances when an eptaxially semi-coherent interface moves, they should also be able to produce a net flow parallel to the interface, thus eliminating the shear component of the shape change, and its associated strain energy. Referring to Fig.1, this would involve the diffusion of matter contained in the region BF'D' to region AFC', in a direction parallel to the interface. Hence, it has been considered improbable (1-3) that an atomic correspondence can be maintained during non-conservative interface motion.
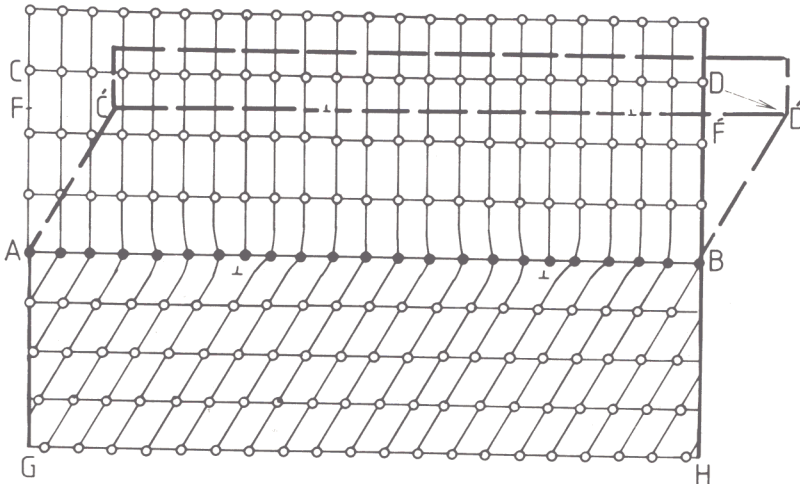

Figure 1 Diagram illustrating the nature of the shape change accompanying the movement of an epitaxially semi-coherent interface.

Such an argument does, however, ignore the fact that the climb motion of the misfit dislocations occurs through a distance equal to the thickness of the plate AC'D'B (Fig.1), whereas the flow necessary to eliminate the shear component of the shape change has to occur through a much greater distance equal to the length of the plate. The thickness to length ratio of such plates is always very small (for good elastic accommodation), and hence it seems feasible that IPS shape changes with shear components will be observed, even when the interface motion is non-conservative in the sense discussed above.

There are some important reservations to this statement, the first being that the free energy change driving the interface must clearly be greater than the strain energy due to the shape change. Secondly, the degree of semi-coherency (in terms of the spacing of the misfit dislocations) must also be important, since the total flux of atoms normal to the interface, necessary for the climb of the misfit dislocations, will increase as the dislocations become more closely spaced. There will clearly be a stage when the

time required to achieve the diffusional <u>flux</u> normal to the interface plane becomes equal to that necessary to diffuse matter parallel to the interface (in order to eliminate the shear component of the shape change), despite the difference in diffusion distances pointed out earlier.

In summary, the examination of surface relief effect may not enable distinction between the two kinds of semi-coherency. The purpose of the present work was to resolve whether the interfaces which accomplish bainitic tranformation in steels are glissile, by examining interface mobility.

<u>Results and Discussion</u>. - The alloy used was prepared as in Ref.4, and the final composition (wt.pct.) was Fe-0.30C-4.08Cr. Most of the experimental details have been published elsewhere (4,5), and some earlier results of relevance (5) to the present problem are summarised below.

It has been demonstrated (5) that the mechanism of bainitic transformation in the above alloy is consistent with the theory (6,7) that the growth of bainitic ferrite occurs by the propagation of martensitic sub-units, with the redistribution of excess carbon occurring <u>after</u> the actual transformation event. Hence, the formation of bainite has been shown to cease (5-7) when the carbon content of the residual austenite (enriched by the post-transformation partitioning of carbon from ferrite to austenite) reaches a level which makes further <u>diffusionless</u> transformation of austenite thermodynamically impossible. This does not, however, mean that <u>diffusional</u> transformation cannot occur beyond this stage (5,6), even though bainitic tranformation may cease.

Figure 2 Transmission electron micrograph of lower bainite, whose original interface with the austenite has degenerated into a series of perturbations.

Hence, Bhadeshia (6) found that a specimen of Fe-0.30C-4.08Cr austenitised at 1100°C for 10 mins and then isothermally transformed at 478°C first transformed to bainite, but

the bainite reaction terminated within some 30 minutes, when the carbon content of the residual austenite was such that diffusionless transformation became thermodynamically impossible. However, on continued holding of the same specimen at the isohtermal transformation temperature (478°C) for 43 days, at least two different diffusional transformations could be detected (one of these, the formation of alloy-pearlite from the austenite grain boundaries is not considered further). Under the influence of the prolonged isothermal heat-treatment, the bainite-austenite interfaces degenerated (Fig.2) into a series of perturbations.

These perturbations grew diffusionally (6) in the sense that they did not exhibit an IPS shape change (unlike the original bainite plates), although microanalysis indicated that they grew without any substitutional alloying element partitioning. It should also be noted that the morphology of the perturbations (Fig.2) is clearly inconsistent with the existence of any macroscopically obvious invariant plane. The ferrite corresponding to the perturbations was found to be in the same crystallographic orientation as the original bainite (implying continuity). Despite all this, it took 43 days for the perturbations to develop and accomplish roughly the same degree of tranformation as the bainite, which developed very rapidly.

Growth Rate Analysis. - The perturbations illustrated in Fig.2 could be a consequence of morphological instability, of the type first discussed by Mullins and Sekerka (8). They showed that a planar interface, moving under the conditions of diffusion controlled growth, can sometimes become unstable and degenerate into a series of perturbations. It is therefore necessary to examine whether the perturbations observed in the present work grew at a carbon diffusion controlled rate.

The theory of diffusion controlled growth (for plate like morphologies) has been fully discussed elsewhere (9), but the following crucial points should be pointed out:

1) The diffusion coefficient of carbon is concentration dependent, and the growth rate was calculated on the basis of a constant diffusion coefficient evaluated as a weighted average over the composition range in the matrix, as discussed by Trivedi and Pound (9,10). The relevant composition range (which delimits the carbon concentration profile ahead of the growing edge) was taken to be 0.0138 - 0.0959 mole fraction of carbon, the upper limit corresponding to the value given (at 478°C) by the Ae'_3 curve of the phase diagram (Fig.1b of Ref.6). The lower limit is the average alloy carbon content.

2) To obtain the weighted average diffusion coefficient ($\bar{D}$), its carbon concentration dependence was evaluated as in (11), taking full account of the influence of Cr on the activity of carbon in austenite. $\bar{D}$ was thus found to be 3.925×10^{-11} cm^2 per second.

3) The interfacial energy was taken to be 0.2 J/m^2, and the growth rate was calculated on the basis of Zener's assumption (see Ref.9) that the plate tip radius adjusts itself to a value consistent with the occurrence of a maximum growth velocity.

4) The growth rate was evaluated using Trivedi's analysis (Eqn.14 of Ref.12) and the Gibbs-Thompson effect (see Ref.9) was taken into account using equation 10 of Ref.12.

All these conditions should give an accurate estimate of the maximum possible growth rate under carbon diffusion control, and this was calculated to be 9.044×10^{-7} m/s.

Taking the experimental time period for the growth of the perturbations to be 43 days, and choosing the maximum length of observed perturbations (about 0.5 microns, Fig.2), the experimental perturbation growth rate turns out to be many orders of magnitude

smaller, at 1.4×10^{13} m/s. These results have two important implications; firstly, the perturbations cannot be rationalised in terms of the Mullins and Sekerka morphological instability theory, since growth is not at a carbon diffusion controlled rate. Secondly, the results clearly demonstrate that thermally activated interface movement at the temperatures where bainitic transformation occurs, is far too slow to account for the very rapid initial growth of the bainite plates. It follows that the interface responsible for the bainite transformation must be of a glissile semi-coherent type. It should be noted that the results are particularly significant since the perturbations have the same orientation relation with the parent austenite as the original bainitic ferrite.

It remains to explain the origin of the perturbations. This can be done in terms of the earlier discussion of the diffusion necessary to eliminate the shear component of the shape change. Clearly, the diffusion distance over which the atoms have to be transported to eliminate the shear component would be minimised if the original nearly planar bainite-austenite interface degenerates into thin perturbations (the thinness being in a direction parallel to the orninginal interface plane).

Conclusions. - It is believed that the present experiments provide strong evidence to suggest that the interface which accomplishes bainitic transformation in steels has a glissile structure.

It is also suggested that under circumstances where the driving force for transformation can fully account for the strain energy that accompanies IPS effects, and when the degree of interfacial coherency is high, the movement of epitaxially semi-coherent interfaces is not likely to lead to an elimination of the shear component of the shape change.

References.

1) CHRISTIAN J.W., 'Theory of Tranformations in Metals and Alloys', Part 1, 2nd Ed., p.359, Pergamon Press, Oxford, (1975).
2) CHRISTIAN J.W., 'Mehcanism ofPhase Transformations in Crystalline Solids', Institute of Metals (London), Monograph 13, (1969), p.129.
3) CHRISTIAN J.W., 'Physical Properties of Martensite and Bainite', Special Rep.93, Iron & Steel Inst. (London) (1965), p.1.
4) BHADESHIA H.K.D.H., Acta Metallurgica,28 (1980) 1103.
5) BHADESHIA H.K.D.H., 'Phase Transformations', Pittsburgh Conference, (1981), in press.
6) BHADESHIA H.K.D.H., EDMONDS D.V., Acta Metallurgica 28 (1980) 1265.
7) BHADESHIA H.K.D.H., WAUGH A.R., Acta Metallurgica 30 (1982) 775.
8) MULLINS W.W., SEKERKA R.F., J. Appl. Phys., 34 (1963) 323.
9) CHRISTIAN J.W., 'Theory of Tranformations in Metals and Alloys', Part 1, 2nd Ed., p.472, Pergamon Press, OXford, (1975).
10) TRIVEDI R., Metall. Trans., 1 (1970) 921.
11) BHADESHIA H.K.D.H., Metal Science, 15 (1981) 477.
12) HILLERT M., Metall. Trans. 6A (1975) 7.