

## Short Communication

# Metallographic observations of bainite transformation mechanism

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*Metallographic observation has directly revealed the continuous motion of the bainite/austenite interface in a manner consistent with the achievement of thermoelastic equilibrium. Further observation demonstrates that the shear deformation associated with the growth of bainite can induce mechanical twinning in the surrounding microstructure in appropriate circumstances. For the purposes of comparison, analogous observations for martensite are also reported.*

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### Introduction

The purpose of the present paper is to report some metallographic observations made during the course of other investigations on the design of high strength steels. These observations are interesting because they reveal aspects of the mechanism of bainitic transformation in steels.

### Experimental

The chemical compositions of the alloys used in the present study are given in Table 1. For each of the alloys, samples in the form of 10 mm dia. bars were sealed in quartz tubes containing pure argon, and homogenised at 1300°C for three days. The bars were then machined down to 8 mm dia. cylinders of 12 mm length.

Alloy A is an experimental steel which only transforms to martensite. In the present work it was quenched into liquid nitrogen to generate plate martensite. Alloy B is a commercial high strength steel (300M) which was austenitised at 1100°C and quenched in water to generate a martensitic microstructure. Alloys C and D are experimental steels designed for the study of bainite. The large silicon concentration ensures the absence of any carbide precipitation during the formation of upper bainite and in the austenite of lower bainite.

Transformation experiments for alloys C and D were carried out in a thermomechanical simulator (Thermecmaster Z) with rf heating, the temperature being monitored by a Pt/Pt-10Rh thermocouple spot welded to the sample. The samples were heated to 1000°C for 10 min in vacuum, cooled rapidly to an isothermal transformation temperature below the bainite start temperature and, after an appropriate period, quenched to ambient temperature using helium jets. The isothermal transformation conditions were 270°C (2500 s) and 325°C (2000 s) for alloys C and D respectively. For alloy C, the sample was maintained at a uniaxial compressive stress of 700 MN m<sup>-2</sup> during isothermal transformation. This was to encourage the plates of bainite to form close to the planes of maximum shear stress and hence to form with large relative inclinations, for reasons discussed below.

Samples for optical microscopy were prepared using standard metallographic techniques, and etched using 2% nital. Thin foils for transmission electron microscopy (TEM) were prepared by electropolishing; the electrolyte consisted of a mixture of 5% perchloric acid, 25% glycerol, and 70% ethyl alcohol. The polishing potential was 55 V at a current of 20–30 mA; at –5°C. The thin foils were

examined using a Philips EM400ST TEM operated at 120 kV.

### Results

#### THERMOELASTIC EQUILIBRIUM

The shape deformation associated with a displacive transformation of austenite in steel can be described as an invariant plane strain with a relatively large shear component. Christian<sup>1</sup> demonstrated that when the shape strain is elastically accommodated, the strain energy scales with the plate aspect ratio (thickness/length). This is why all of the displacive transformation products in steel, e.g. Widmanstätten ferrite, bainite, and martensite, occur in the form of thin plates.

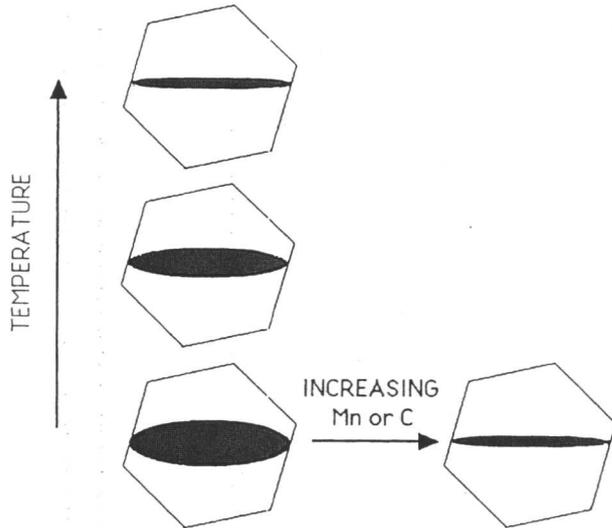
The need to minimise strain energy demands a thin plate, but this also leads to a minimisation of the volume of transformation per plate. Therefore, a plate will tend to adopt the largest aspect ratio consistent with the available free energy change driving the transformation. In ideal circumstances, where the transformation interface remains glissile throughout, and where there is no friction opposing the motion of the interface, thermoelastic equilibrium occurs.<sup>2–5</sup> The aspect ratio of the plate adjusts so that the strain energy is equal to the driving force (Fig. 1). The driving force can be increased by cooling the austenite, which should lead to plates of increased thickness. Alternatively, it can be decreased by the addition of austenite stabilising elements such as manganese or carbon.

The thermoelastic equilibrium illustrated in Fig. 1 has been frequently demonstrated for martensite,<sup>2–5</sup> but not for bainite. One reason for this could be that bainite grows in the form of sheaves, which are clusters of connected platelets which grow in parallel formations. The spacing between the platelets is to a large extent controlled by the carbon diffusion field associated with each platelet,<sup>6</sup> therefore the increase in platelet thickness is restricted by the diffusion fields and by the presence of adjacent parallel platelets (i.e. soft and hard impingement effects).

A further complication is that the bainite transformation occurs at high temperatures where the austenite is mechan-

Table 1 Chemical compositions (wt-%) of alloys studied

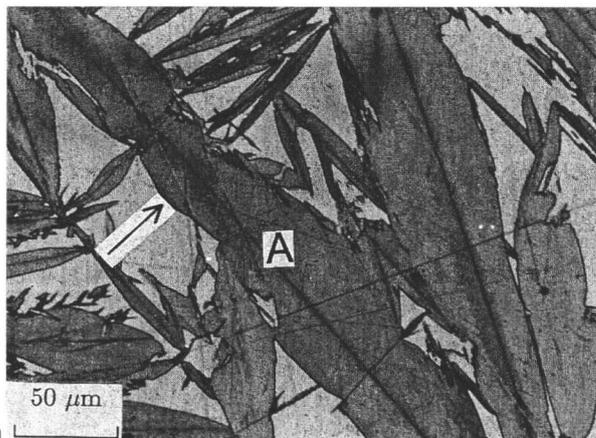
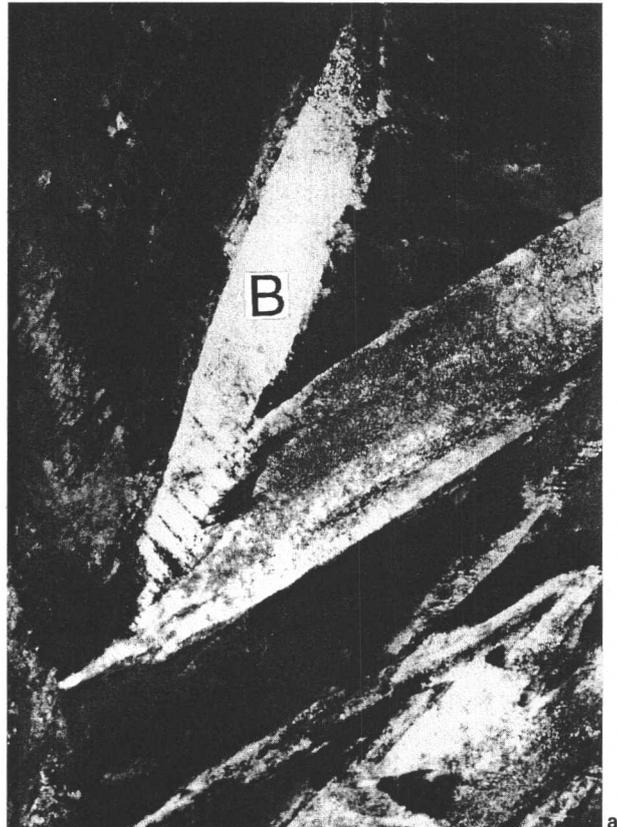
| Alloy | C    | Si   | Mn   | Cr   | Mo   | Ni   | V    |
|-------|------|------|------|------|------|------|------|
| A     | 0.30 | ...  | ...  | ...  | ...  | 30.1 | ...  |
| B     | 0.44 | 1.74 | 0.67 | 0.83 | 0.39 | 1.85 | 0.15 |
| C     | 0.46 | 2.10 | 2.15 | ...  | ...  | ...  | ...  |
| D     | 0.27 | 1.98 | 2.18 | 1.9  | ...  | ...  | ...  |



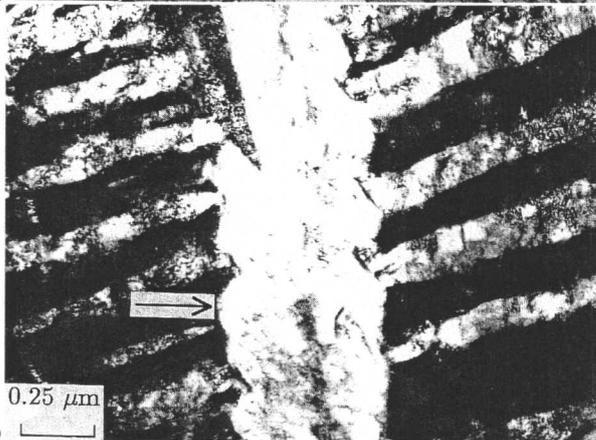
1 Effect of thermoelastic equilibrium on aspect ratio of plate with fixed length

ically weak. The shape deformation therefore causes plastic deformation, and the resulting dislocation debris eventually blocks the transformation interface, which loses coherency. This is why platelets of bainite are arrested in their growth even when their size is much smaller than the austenite grain size.<sup>7,8</sup>

It should nevertheless be possible to observe the increase in thickness of individual bainite platelets if the process is captured at a suitably early stage of growth. Some published



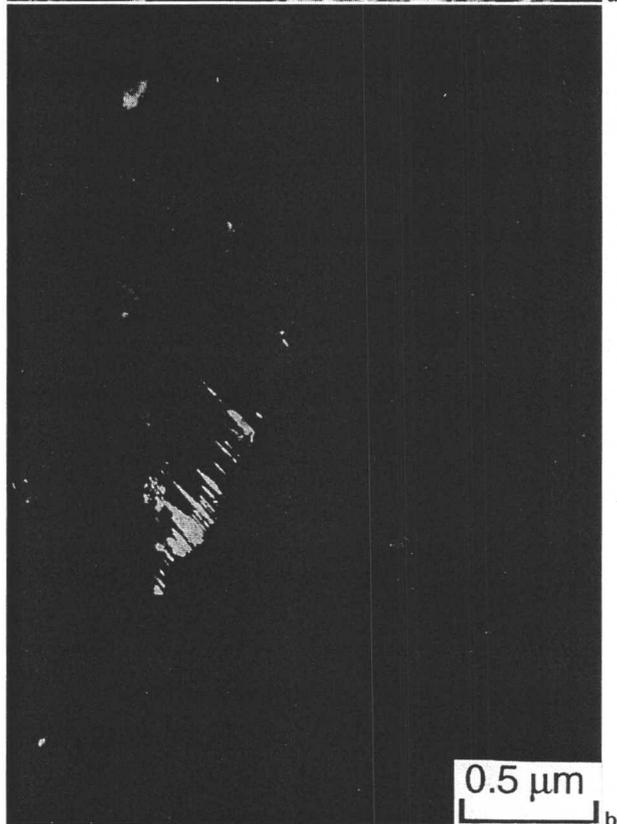
a



b

a martensite in alloy A; b bainite in alloy C

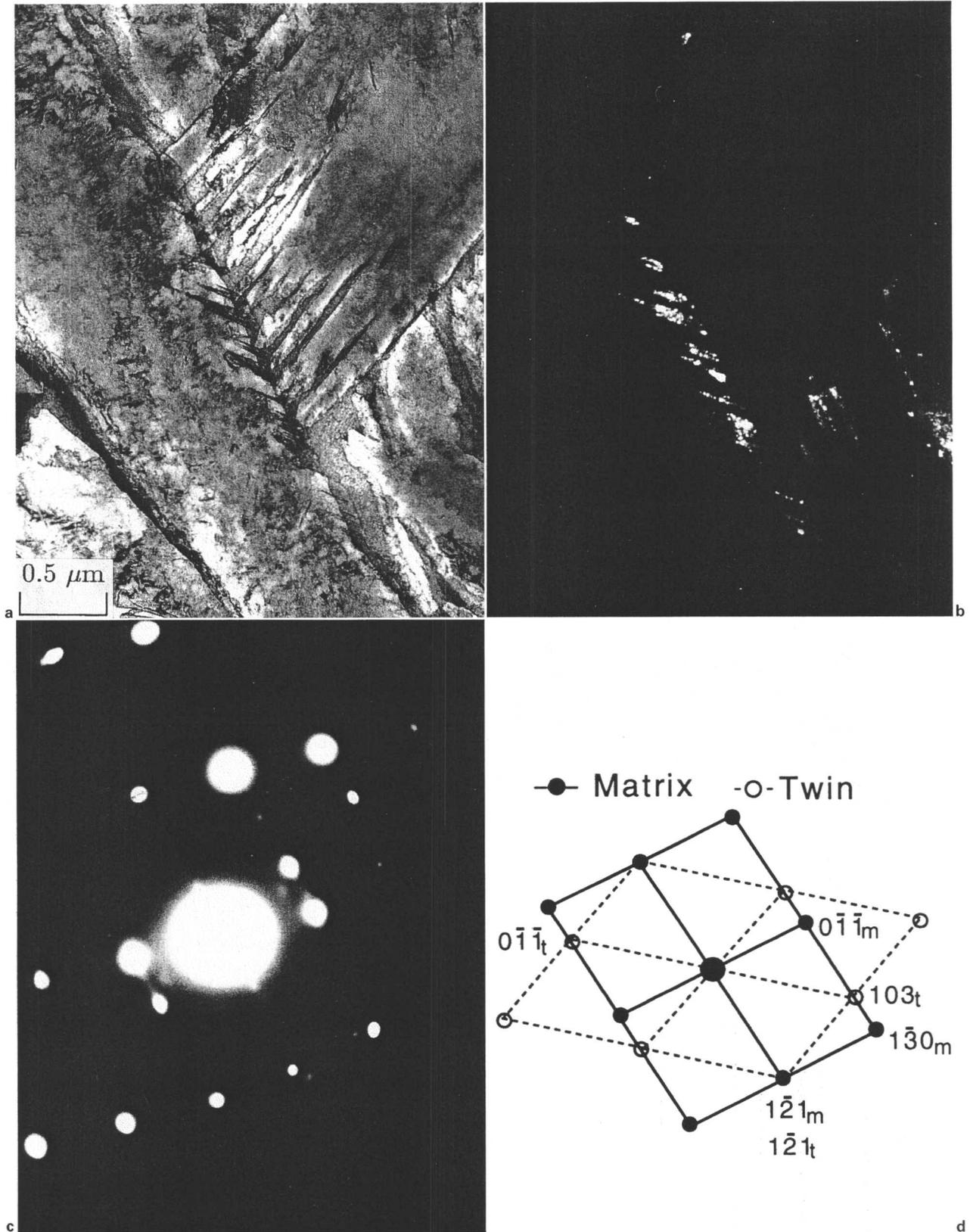
2 Bowing of transformation interface at strong pinning points, particularly prominent in regions identified by arrows



a

a bright field image; b dark field image of twins in one of martensite plates

3 Alloy B showing accommodation twinning at collision sites between inclined plates of martensite



a bright field image; b dark field image of twins in one of bainite plates; c diffraction pattern confirming twinning; d interpretation of pattern in c  
**4 Alloy D showing accommodation twinning at collision sites between inclined plates of bainite obtained by isothermal transformation**

observations<sup>9,10</sup> can be interpreted as supporting the idea that an increase of bainite plate thickness continues until the chemical driving force is exhausted by the accumulation of strain energy. The bainite is then in a different microstructural form called 'acicular ferrite'. Acicular ferrite

is bainite which nucleates intragranularly on non-metallic inclusions.<sup>8</sup> Thus, isolated platelets can be seen to emanate from point nucleation sites and are therefore free to increase in thickness even when their increase in length is restricted by other platelets growing from other inclusions. It is found

experimentally that the aspect ratio of acicular ferrite decreases as the transformation temperature is raised, or as the manganese or carbon concentration is increased.<sup>9,10</sup>

In the present work, bainite platelets were forced to grow in non-parallel formations by transforming the austenite under the influence of an applied compression stress. Figure 2a shows a phenomenon commonly found with martensite. Because the plates rapidly increase in length, impingement occurs and increase in thickness then continues, restricted by pinning points. The interface of plate A in Fig. 2a is clearly bowed between these points. The corresponding phenomenon for bainite platelets is illustrated in Fig. 2b. It is sometimes considered that bainite grows via a superledge mechanism<sup>11</sup> in which macroscopic ledges propagate as the habit plane. The smooth curved regions of the interface between the pinning points (Fig. 2b) provide evidence that the bainite/austenite interface moves continuously, rather than by a step mechanism involving the translation of superledges.

### ACCOMMODATION TWINNING

Bainite and martensite can be regarded as mechanisms for deformation which in addition lead to a change in crystal structure. However, the key feature of the deformation is the large shear component ( $\sim 0.22$ ) of the invariant plane strain shape change, which is much larger than the dilatational strain ( $\sim 0.03$ ).<sup>8</sup>

The small volume strain, which is directed normal to the habit plane, cannot induce any significant shear deformation in its vicinity. It is the shear which is directed parallel to the habit plane which has to be accommodated (elastically or plastically) by the surrounding material. This is why a martensite plate tapers to a sharp point, because the magnitude of the shear displacement scales with the plate thickness. The plate need not taper if the strain is plastically accommodated. This plastic deformation can take the form of accommodation twinning, so called in order to distinguish these mechanical twins from those sometimes found in martensite as transformation twins.<sup>12</sup>

This is apparent in the collision shown in Fig. 3 between inclined plates of martensite in a low alloy steel. Note that the resulting mechanical twins are confined to the abutting surface of plate B, and that as expected, the mechanical twins taper away from the collision site. Exactly the same phenomenon is shown for bainite in Fig. 4b, providing unusual direct evidence for the role of the shear strain of bainite, as the bainite forms deep inside the metal (Sandvik<sup>13</sup> previously demonstrated this by noting the displacement of narrow twins in austenite). The shear has of course on

many occasions been revealed in experiments where surface displacements are observed.<sup>8</sup>

### Summary

In appropriate circumstances, the austenite/bainite interface is capable of bowing between strong pinning points as the bainite platelet increases in thickness to an extent permitted by the available chemical driving force. The shear caused by the growth of bainite can induce accommodation (mechanical) twinning in adjacent plates.

### Acknowledgements

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### References

1. J. W. CHRISTIAN: *Acta Metall.*, 1958, **6**, 377-379.
2. G. V. KURDJUMOV and L. G. KHANDROS: *Dokl. Akad. Nauk SSSR*, 1949, **66**, 221.
3. G. B. OLSON and M. COHEN: *Scr. Metall.*, 1977, **11**, 345-347.
4. J. W. CHRISTIAN: in Proc. Int. Conf. 'ICOMAT '79', 220-234; 1979, Massachusetts, MIT Press.
5. R. J. SALZBRENNER and M. COHEN: *Acta Metall.*, 1979, **27**, 739-748.
6. L. C. CHANG and H. K. D. H. BHADSHIA: *Mater. Sci. Technol.*, to be published.
7. H. K. D. H. BHADSHIA and D. V. EDMONDS: *Metall. Trans.*, 1979, **10A**, 895-907.
8. H. K. D. H. BHADSHIA: 'Bainite in steels', 1-451; 1992, London, The Institute of Materials.
9. S. A. COURT and G. POLLARD: in 'Welding metallurgy of structural steels', (ed. J. Y. Koo), 335-347; 1987, Warrendale, PA, TMS-AIME.
10. Zhu Yau ZHANG: PhD thesis, University of Southampton, 1994.
11. H. I. AARONSON, C. LAIRD, and K. R. KINSMAN: 'Phase transformations', 313-396; 1970, Metals Park, OH, ASM.
12. H. K. D. H. BHADSHIA and D. V. EDMONDS: 'Phase transformations', (IV) 4-8; 1979, London, The Institute of Metallurgists.
13. B. J. P. SANDVIK: *Metall. Trans.*, 1982, **13A**, 777-787.