High resolution observations of displacements caused by bainitic transformation

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High resolution measurements of the displacements caused by the formation of bainite in a steel have confirmed that the surface relief due to each platelet in any given sheaf is identical to that of the others in a sheaf, and that the relief of each subunit conforms with the general features of an invariant plane strain. The maximum observed shear strain was found to be about 0·26, which is consistent with the magnitude expected from the phenomenological theory of martensite crystallography. Clear evidence has also been obtained of the plastic deformation induced in the adjacent austenite by the growth of bainite. MST/3270

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Introduction

The Bain strain is a homogeneous pure deformation which can change the fcc crystal structure of austenite into the bcc (or bct) structure of ferrite. This pure deformation can be combined with a rigid body rotation to give a net lattice deformation which leaves a single line unrotated and undistorted (i.e. an invariant-line strain). In a situation where the transformation is constrained, such a low degree of fit between the parent and product lattices would lead to a great deal of strain as the product phase grows.

However, by adding a further inhomogeneous lattice-invariant deformation (shear or twinning), the combination of deformations appears macroscopically to be an invariant-plane strain.[1,2] An invariant-plane is one which is unrotated and undistorted, so that a greater degree of fit is achieved between the austenite and ferrite. This invariant-plane is also known as the habit plane, and the nature of the invariant-plane strain is illustrated in Fig. 1, where s and δ represent the shear and dilatational components of the strain.

As pointed out by Christian,[3] a planar interface traversing a crystal and having no long range elastic field produces an invariant-plane strain (IPS) shape change for all transformations. However, when the IPS shape change has a large shear component, it implies at the very least coherency at the parent/product interface, which must be glissile.

Another obvious implication of an IPS shape deformation is that the product phase must be in the form of a thin plate whose habit plane is the invariant-plane.[3,4] This is because the strain energy due to the IPS shape deformation is then minimised.[4] It is not therefore surprising that in all instances where austenite transforms to plate shaped ferrite, the growth of the ferrite causes an IPS shape change (*see* Table 1). This is not the case when the ferrite grows in any other shape.

Of the transformation products listed in Table 1, both Widmanstätten ferrite and martensite can be obtained in the form of plates which can be observed using an optical microscope. It has therefore been possible to clearly establish the nature of the shape deformation by measuring the deflection of scratches (or optical interference fringes) on a surface which was polished flat before transformation.

However, the microstructure of bainite consists of fine plates of ferrite, each of which is only 0·2 μm in thickness, which is below the limit of light microscopy. These plates are called 'subunits' because they grow in clusters known as sheaves. Within each sheaf the subunits are parallel and of identical crystallographic orientation and habit plane. The subunits are usually separated from each other either by the presence of carbides or residual austenite.[8]

This fine microstructure has made it difficult to establish the surface relief introduced as bainite grows. Ko and Cottrells' classic work[9] qualitatively established the nature of the surface relief. Subsequent light microscopy[10–12] indicated values of s of $\sim$0·13, which were found to be inconsistent with the larger values predicted by theory, which are in the range 0·22–0·28 (Refs. 12–14). These difficulties are associated with the fact that light microscopy can only reveal the average shape deformation of the whole sheaf rather than of an individual subunit. This was demonstrated by Sandvik,[6] who determined from the displacement of twin boundaries by subunits, using transmission electron microscopy (TEM), that the value of s is close to 0·22, which is consistent with theory.

A new method has recently become available for the quantitative high resolution measurement of surface topography. This involves the use of a scanning tunnelling microscope; the method has already been applied to martensitic transformations[15–17] but has considerable potential for the fine structure of bainite. In fact, the method has been applied to bainite,[18] but only to make morphological observations.

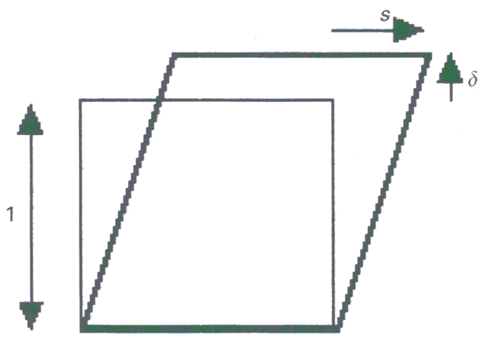
The purpose of the present work was to use an atomic force microscope (AFM), which is also capable of high resolution observation and measurement of surface topography, to determine directly the surface relief of subunits of bainite.

Table 1 Approximate values of s and δ for variety of transformation products in steels

Transformation	s	δ	Morphology	Ref.
Widmanstätten ferrite	0·36	0·03	Thin plates	5
Bainite	0·22	0·03	Thin plates	6–14
Martensite	0·24	0·03	Thin plates	7

Experimental technique

The alloy studied has the chemical composition Fe–0·24C–2·18Si–2·32Mn–1·05Ni (wt-%). The alloy is one from a separate investigation on high strength bainitic steels. The essential feature of this alloy is that it contains a relatively large silicon concentration, which suppresses the precipitation of cementite from austenite.[8] Consequently, carbide free upper bainite can be obtained which consists of just bainitic ferrite and retained austenite, allowing the observation of surface relief without interference from the secondary formation of carbides. The alloy also has

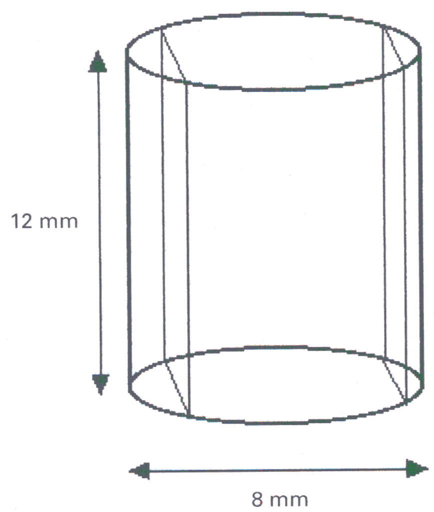


1 Invariant-plane strain with shear strain *s* and dilatational strain *δ*: dilatational strain is directed normal to habit plane

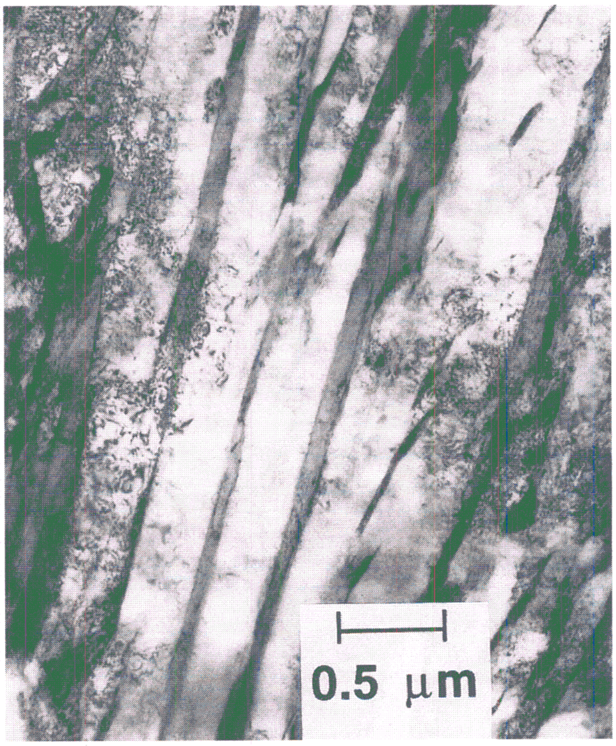
sufficient hardenability to avoid the formation of other high temperature transformation products during heat treatment.

A number of cylindrical samples (8 mm in diameter and 12 mm in length) were machined; 2 and 1 mm deep flats were then machined on opposite sides of the cylinder, as illustrated in Fig. 2. The overall size of the sample is suitable for heat treatment in a thermomechanical simulator; one of the flats is needed to securely place the sample in an AFM, and the other (2 mm deep) to make observations of surface relief. The latter was metallographically polished to a 0·5 μm finish, and cleaned thoroughly with high purity ethanol. All subsequent handling was with gloves, and any storage was in a dry dessicator.

All the heat treatments were carried out in an adapted Thermecmaster Z simulator. Using this, the polished sample was heated by induction in a high vacuum (4 Pa, with a Pt–PtRh (type *R*) thermocouple attached to the centre of one of the curved sides. After the austenitisation heat treatment (1200°C for 120 s), the sample was cooled rapidly to the required isothermal transformation temperature using helium gas. Helium was used in preference to nitrogen to achieve higher cooling rates and more accurate control. A number of isothermal transformation temperatures were tested to characterise the formation of bainite. The isothermal heat treatment finally selected for the surface relief experiments was 350°C (for 2000 s), on the grounds that it gave sufficient bainitic ferrite to enrich the austenite and prevent excessive martensitic transformation on cooling



2 Sample shape for heat treatment in thermomechanical simulator followed by examination in atomic force microscope (AFM)



3 Transmission electron micrograph illustrating general microstructure

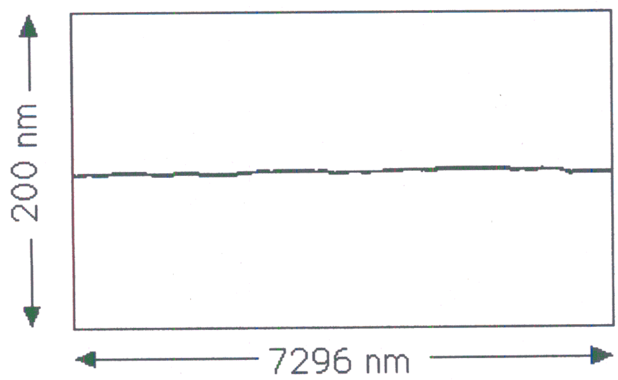
to ambient temperature. To avoid oxidation of the sample, the vacuum was reapplied immediately after the quench to 350°C, so that the formation of bainite occurred in a vacuum. The specimen was helium quenched to ambient temperature after the 350°C treatment.

The surface relief caused by the formation of bainitic ferrite was examined using a Seiko SPA300 AFM.

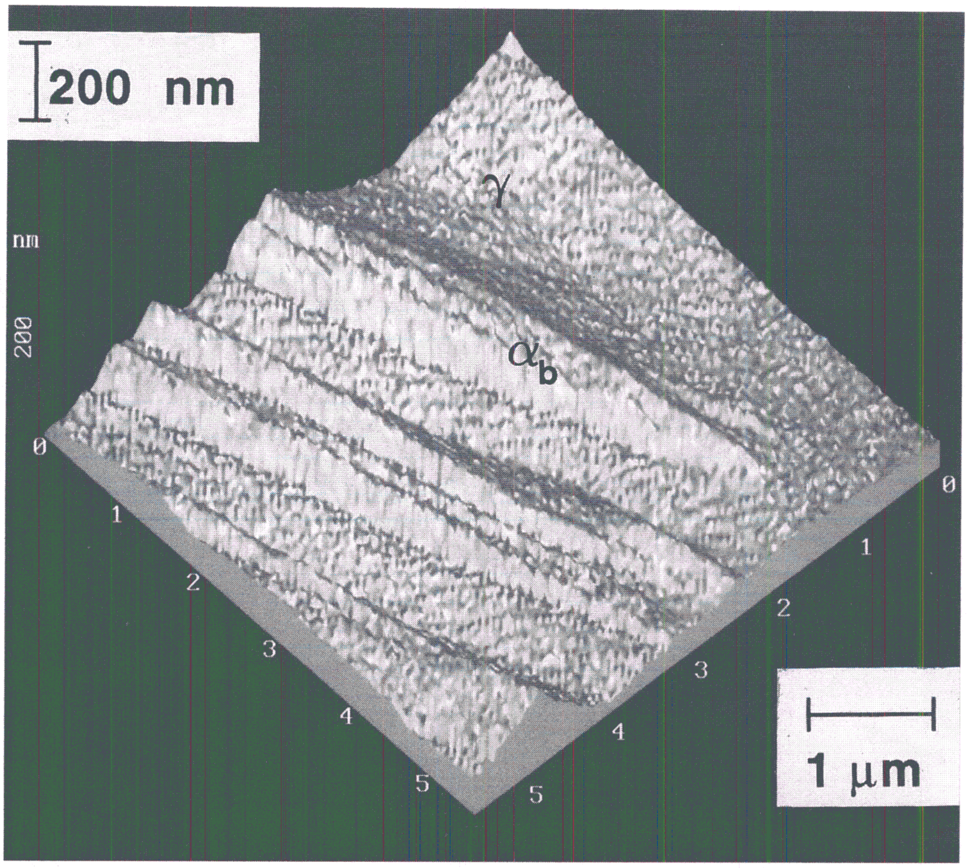
Results and discussion

The general microstructure, as observed using TEM, is illustrated in Fig. 3. It consists of a mixture of bainitic ferrite and retained austenite, although a small amount of extremely fine martensite cannot be ruled out. Since the scale of any martensite is much smaller than that of the bainite, it is not expected to interfere with the relief observations.

The surface relief of a sample that was not heat treated after metallographic preparation (i.e. polished) is shown in Fig. 4. This image is presented at a similar resolution to



4 AFM measurement of topography of untransformed metallographically polished sample



5 AFM image showing surface relief due to individual bainite subunits which all belong to tip of sheaf

surface relief images from transformed specimens to enable a comparison to be made. It will become obvious from these images that the magnitude of the transformation relief is far greater than the polishing relief evident in Fig. 4. It should also be noted that deliberate attempts were made throughout the work to make measurements on platelets which were clearly isolated, as in Fig. 5, which represents a region near the tip of a bainite sheaf. An AFM image from a more confusing region (as far as measurement and phase distinction are concerned) is presented and discussed below.

Measurements were attempted in positions where bainite subunits could be clearly seen in regions of austenite, both to obtain a reference surface for the relief measurements, and to clearly identify the bainitic ferrite from the morphology. An AFM image showing such a microstructure is presented in Fig. 5 which is a 'solid model'. The AFM actually scans the surface and records the surface contours at regular intervals. This information is valuable for the quantitative measurements, but for visualisation purposes, the information can be processed to give a perspective view as illustrated in Fig. 5. The individual subunits of bainite are at the tip of a single sheaf of bainite. They have identical displacements, and the displacements are well defined, unlike the smoothly curving regions of plastic accommodation in the adjacent austenite.

Figure 6*a* and *b* shows two typical linescans across a single bainite subunit. This particular measurement gave an apparent shear strain s_A of $\sim$0·18. This measurement is called an apparent shear because the inclination of the habit plane relative to the free surface is unknown. The apparent shear strain is always expected to be less than, or equal to, the actual shear strain *s*. The actual shear strain would be that measured only if the habit plane normal lies in the free surface. It is estimated that the error in the measurement of the shear is about $\pm$0·02, judging from the variation seen along the length of a single platelet.

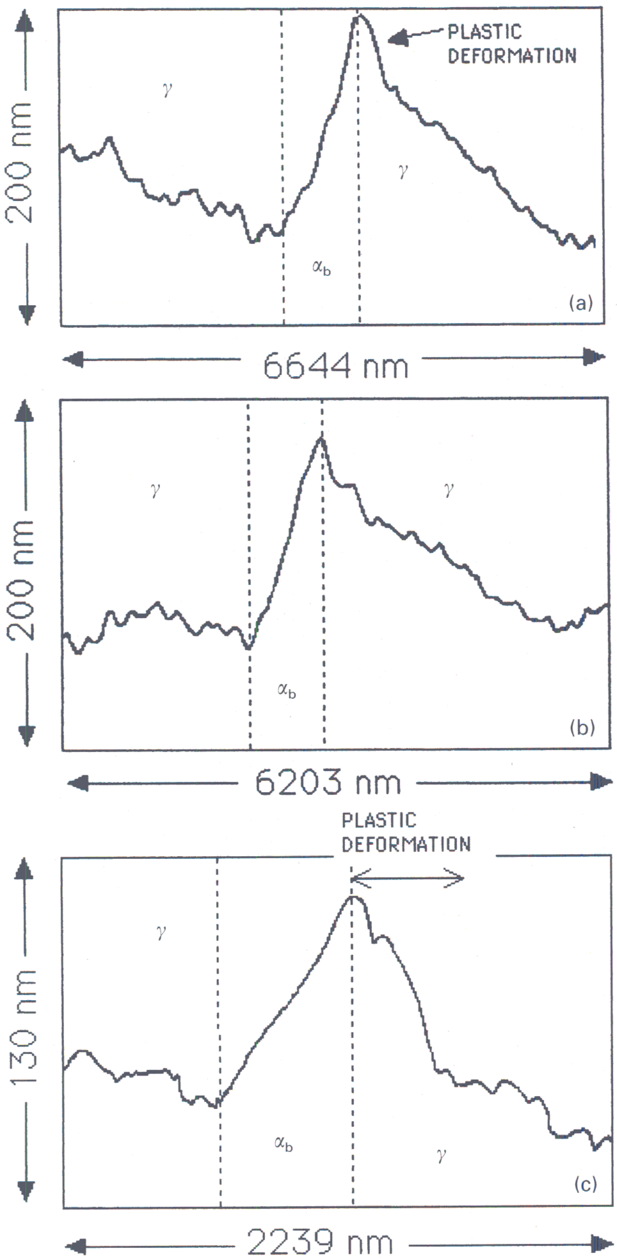
It is evident from Fig. 6*a* and *b* that there is some plastic deformation in the austenite adjacent to the bainite subunit. This is indicated by the fact that the scan is not perfectly linear away from the ferrite, but is initially steep (marked on Fig. 6*a*). It is also interesting to note that the scan within the ferrite is much more regular than within the austenite. This indicates that the plastic accommodation of the shape strain is mostly confined to the austenite, as would be expected theoretically.[19,20]

It should be emphasised that the effect of the dilatational component of the IPS is ignored in the measurements quoted in the present paper. It is not possible to separate *s* and δ because the detailed orientation of the habit plane relative to the free surface is not known. However, the error should be very small since $\delta \ll s$.

Figure 6*c* shows a linescan across another bainite subunit ($s_A = 0{\cdot}19$) where the effect of plastic accommodation is much more evident. The plastic deformation extends about the width of the bainite and appears to be asymmetrical. The reasons for the asymmetry could be complex, for example the location of other plates in the proximity could alter the deformation field, but a small degree of asymmetry is expected in any case because of the dilatational strain.[20]

Figure 7 gives further examples of the features discussed above; these serve to reinforce the earlier observations. In addition, Fig. 7*a* shows some periodic irregularities (marked by a vertical arrow) to the right of the bainitic ferrite region. It is speculated that these could be due to the formation of martensite, which grows in the form of packets of extremely fine platelets from the carbon-enriched austenite.

The measured values of s_A are summarised in Table 2, and plotted in Fig. 8, as a function of the test number. Also plotted in Fig. 8 is the range of theoretically estimated values of the shear strain *s* of bainite.[12–14] The present results are all consistent with theory, and they suggest that



a, *b* $s_A = 0{\cdot}18$; *c* $s_A = 0.19$

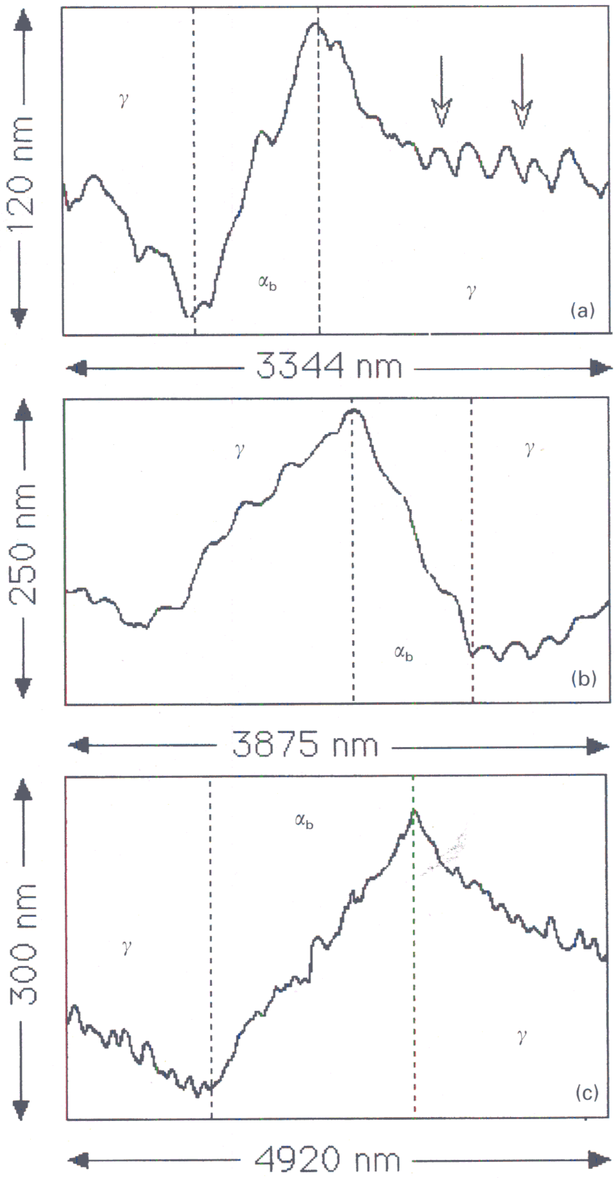
6 AFM scans showing surface relief due to individual bainite subunit: *a* and *b* are linescans from different regions of same platelet, and *c* is scan from another platelet (arrows in part *c* indicate plastically deformed austenite)

the shear strain is at least at the maximum observed value of $s_A = 0{\cdot}26$.

The work shows that the shape strain is in fact larger than was reported in earlier work,[10–12] consistent with

Table 2 Measured values of apparent shear component s_A of shape deformation due to bainite subunits (maximum value is $s_A = 0{\cdot}26$)

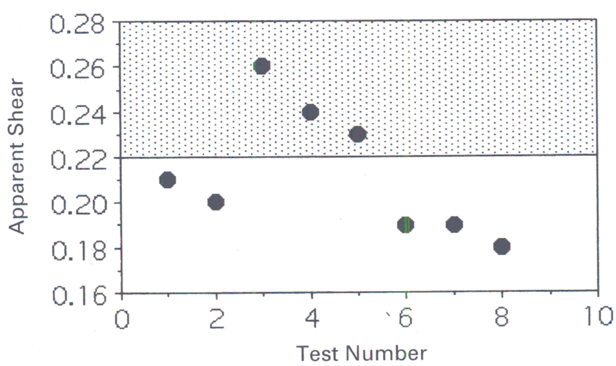
Sample	Measured s_A
1	0·21
2	0·20
3	0·26
4	0·24
5	0·23
6	0·19
7	0·19
8	0·18



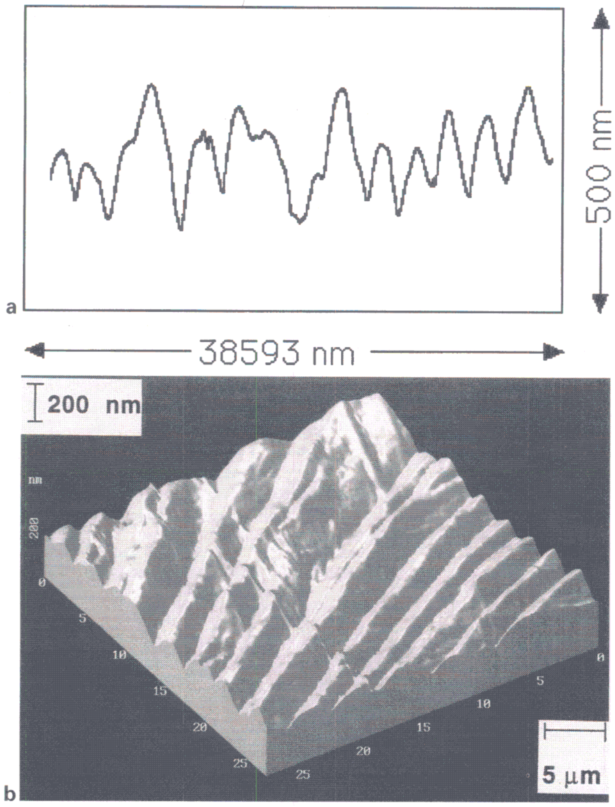
a $s_A = 0{\cdot}21$; *b* $s_A = 0{\cdot}26$; *c* $s_A = 0{\cdot}20$

7 AFM scans showing surface relief due to three different bainite subunits

Sandvik's TEM deduction. This is because the early data were derived using light microscopy, in which case they refer to the shape deformation averaged over a sheaf, rather



8 Measured values of *s* plotted against arbitrary test number: shaded region represents range of theoretically predicted values of *s* depending on various assumptions about mode of lattice-invariant shear, etc.



9 *a* AFM scan showing surface relief due to bainite sheaf consisting of parallel subunits and *b* AFM image showing bainite sheaf structure

than to an individual subunit. This is apparent from Fig. 9, which shows a linescan across a sheaf of parallel subunits. An averaged scan across the sheaf would clearly give a misleading value of the shear strain.

Conclusions

An atomic force microscope has been used to characterise the surface relief caused by the growth of individual subunits of bainite. These studies have been conducted using an alloy which contains only bainitic ferrite and austenite. The results indicate that the minimum value of the shear strain is about 0·26 with an estimated error of ±0·02. This is consistent with what is expected from the phenomenological theory of martensite crystallography. The shape change is not elastically accommodated, but there is significant plastic deformation caused in the adjacent austenite. This result is expected to be quite general for bainite in all steels, since the yield strength of austenite decreases with increasing temperature. It is interesting that much of the plastic accommodation is confined to the austenite rather than the ferrite.

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