

## EFFECT OF STRESS & STRAIN ON FORMATION OF BAINITE IN STEELS

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### Abstract

The growth of bainite causes coordinated displacements of atoms during the motion of its glissile transformation interface. The response of such a transformation to elastic or plastic strain in the austenite is discussed in the context of steel processing. There are some remarkable effects not found in ordinary thermomechanically processed steels.

### THERMOMECHANICAL PROCESSING

The success of thermomechanically processed steels is because of their fine and relatively homogeneous microstructures, consisting of 10  $\mu$  m sized equiaxed ferrite grains. These grains grow by a *reconstructive* mechanism in which all the atoms (including iron) diffuse, so that their growth is not impeded by crystallographic discontinuities such as austenite grain boundaries. This is not the case when the ferrite crystal structure is generated by a deformation of the corresponding austenite structure, because the ensuing coordinated displacements cannot in general be sustained across grain boundaries. Ferrite which grows by this *displacive* mechanism will be in the form of thin plates because this minimises the strain energy associated with the displacements [1]. The inability of such ferrite to cross grain boundaries, and its shape, both lead to a greater refinement of the microstructure. The mean free slip distance through a plate is only about twice its thickness [2], which is typically less than a micrometer in size. The modern trend is, therefore, to process lean steels into fine microstructures of ferrite plates [3]. Examples include the accelerated cooled thermomechanically processed steels with bainitic microstructures [3].

The microstructure of the (final) austenite, just before it begins to transform, is the main outcome of thermomechanical processing. Fine austenite grains inevitably evolve into fine ferrite grains owing to the increased number density of grain boundary nucleation sites. Defects other than grain boundaries, or residual stress in the final austenite can also influence the development of the subsequent transformation microstructure. The purpose here is to present an understanding of the effects of stress and plastic strain on the bainite transformation.

## BAINITE: ORIGIN OF THE STRESS POTENTIAL

The familiar mechanisms of plastic deformation are slip, mechanical twinning and diffusion-induced creep. For very small plastic strains, the first two of these deformation modes are conservative – *i.e.* they preserve an atomic correspondence between the deformed and undeformed parts of the crystal so that the crystal contains a memory of its original shape. All of these deformation modes are *lattice-invariant* because although they cause a change in the shape, the crystal structure remains as it was prior to deformation.

Phase transformations, by contrast, refer to changes in the crystal structure. However, like lattice-invariant deformations, they can also be associated with changes in the shape of the transformed regions. The transformation can have a pronounced interaction with an externally applied stress only if it is accompanied by a shape change (a strain).

We define therefore, the term *stress potential* which describes the ability of a transformation to react to an externally applied stress. In steels, austenite transforms into many varieties of ferrite, which can be classified into two broad categories according to the mechanism of transformation (Fig. 1). Reconstructive transformations occur with the uncoordinated diffusion of all of the atoms, including iron [3,4]. They need not be accompanied by a change in chemical composition, but it is worth emphasising that reconstructive transformations cannot be sustained without the diffusion of all atoms. For example, the freezing of water to ice is a reconstructive transformation where there is mass transport but no change in composition. Similarly, austenite in pure iron can transform into ferrite, either by a reconstructive mechanism involving diffusion, or by a martensitic mechanism. The latter case represents displacive transformations, in which the pattern of atomic arrangement is altered by deformation. There is no diffusion of iron or substitutional solutes during displacive transformations in iron alloys.

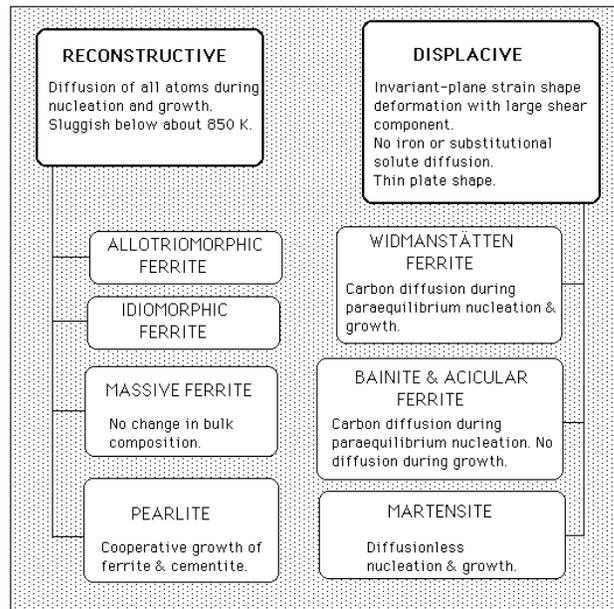


Fig. 1: Classification of the transformation products of austenite [3].

Both kinds of reactions lead to changes in shape (Fig. 2a). The simplest of these occurs during

reconstructive transformations in which the densities of the parent and product phases are different. Hence, when austenite transforms into grain boundary allotriomorphic ferrite, there is a uniform expansion, and vice-versa.

Much more interesting deformations are associated with displacive transformations, where the shape change is in general described as an *invariant-plane strain* with a large shear component. Such a deformation leaves the plane of contact between the parent and product phases undistorted and unrotated. This plane is called the *habit plane*. The strain energy associated with a constrained invariant-plane strain (IPS) is minimised when the product phase has a thin-plate shape. This is why Widmanstätten ferrite, bainite, acicular ferrite and martensite in steels grow in the form of plates. The detailed characteristics of the shape change due to transformation are emphasised in Table 1. The latter also includes some data for mechanical and annealing twins in ferrite and austenite respectively. These are not, of course, transformation products, so that they do not cause any volume change; however, their growth mechanisms are analogous to those of displacive and reconstructive transformations respectively. The shear component of the IPS for the mechanical twin is seen to be substantially larger than all of the transformations listed. Indeed, it is only in iron–beryllium alloys ( $\text{Fe}_3\text{Be}$ ) that martensitic transformation has a shear strain which is as large as that for mechanical twins in ferrite [5]. These empirical observations show that in general, the shear associated with transformations in steels tends to be smaller than that caused by mechanical twinning, but nevertheless, much larger than the dilatational strains.

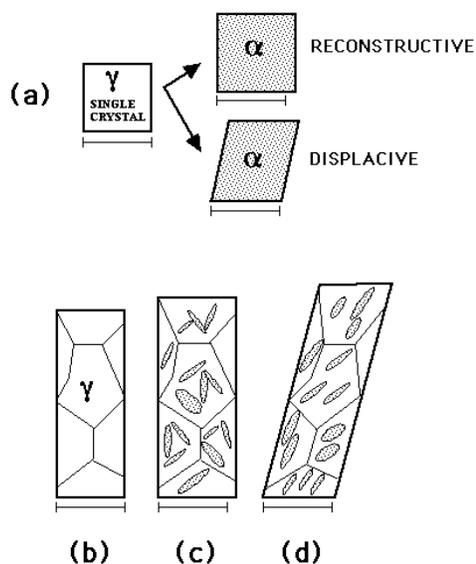


Fig. 2: Shape changes accompanying unconstrained transformations. Horizontal scale bars are all of the same length. (a) Two kinds of deformations as a crystal of austenite transforms to one of ferrite. (b) Polycrystalline austenite. (c) Polycrystalline austenite partially displaced into random arrangement of ferrite plates. (d) Polycrystalline austenite partially displaced into organised array of ferrite plates.

Table I – Shape change due to unconstrained transformation. An invariant–plane strain here implies a large shear component as well as a dilatational strain normal to the habit plane.  $s$  and  $\delta$  refer to the shear and dilatational strains respectively.

Transformation	Shape Change	$s$	$\delta$	Morphology
Allotriomorphic Ferrite	Volume change	0.00	0.02	Irregular
Idiomorphic Ferrite	Volume change	0.00	0.02	Equiaxed, faceted
Pearlite	Volume change	0.00	0.03	Spherical colonies
Widmanstätten ferrite	Invariant–Plane Strain	0.36	0.03	Thin plates
Bainite	Invariant–Plane Strain	0.22	0.03	Thin plates
Acicular Ferrite	Invariant–Plane Strain	0.22	0.03	Thin plates
Martensite	Invariant–Plane Strain	0.24	0.03	Thin plates
Cementite Plates	Invariant–Plane Strain ?	0.21 ?	0.16 ?	Thin plates
Mechanical Twins ( $\alpha$ )	Invariant–Plane Strain	$1/\sqrt{2}$	0.00	Thin plates
Annealing Twins ( $\gamma$ )		0.00	0.00	Faceted

#### *Origin of Shear During Bainitic Transformation*

The shear strains listed in Table I are of particular importance. They are large in comparison with the dilatational strains, they interact equally with either a tensile or a compressive uniaxial stress, and dominate all aspects of the microstructure of Widmanstätten ferrite, bainite and martensite. What then is the origin of these shears? There is a well–proven and elegant theory to answer this question [6,7].

The Bain strain is a homogeneous pure deformation which can change the face–centered cubic crystal structure of austenite into the body–centered cubic (or tetragonal) 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 [6,7]. 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*.

This is illustrated further in Fig. 3 [8,9] in order to emphasize the details of the deformations involved. The Bain strain converts the structure of the parent phase into that of the product phase. When combined with an appropriate rigid body rotation, the net homogeneous lattice deformation  $\mathbf{RB}$  is an invariant–line strain (step  $a$  to  $c$  in Fig. 3). However, the observed shape deformation is an invariant–plane strain  $\mathbf{P}_1$  (step  $a$  to  $b$  in Fig. 3), but this gives the wrong crystal structure. If, however, a second homogeneous shear  $\mathbf{P}_2$  is combined with  $\mathbf{P}_1$  (step  $b$  to  $c$ ), then the correct structure is obtained but the wrong shape since

$$\mathbf{P}_1\mathbf{P}_2 = \mathbf{RB}$$

These discrepancies are all resolved if the shape changing effect of  $\mathbf{P}_2$  is canceled macroscopically by an inhomogeneous lattice-invariant deformation, which may be twinning (step  $c$  to  $d$ ) or slip (step  $c$  to  $e$ ), as illustrated in Fig. 3.

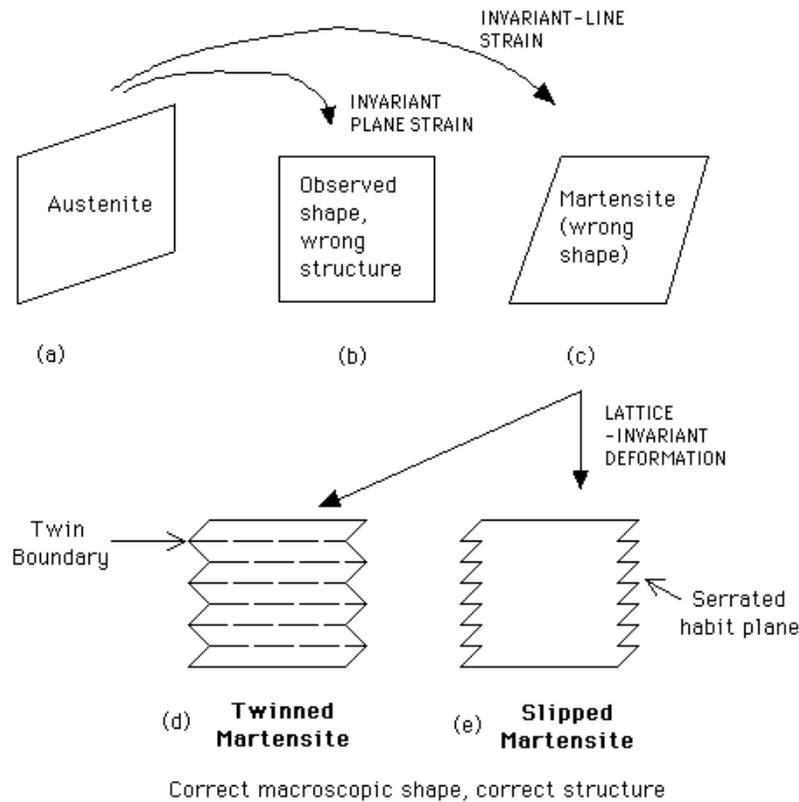


Fig. 3: Schematic illustration of the phenomenological theory of martensite crystallography

## STRESS-AFFECTED TRANSFORMATION

### *Mechanical Driving Force*

The interaction with an externally applied stress can influence the free energy change driving the transformation, the details being dependent on the mechanism of transformation. For reconstructive transformations, it is only the hydrostatic component of stress that can interact with the shape change (due to density change). The response to an arbitrary stress is therefore not expected to be as large as that for displacive transformations in steels. In that case, the shear component of the IPS is relatively large and so is the corresponding mechanical driving force component [10].

We now focus on displacive transformations, using a method which is based on Patel and Cohen's work [11]. In the above discussion, the total driving force was partitioned implicitly into a "mechanical driving force" and the more familiar chemical driving force [11,12]. This is based on the reasoning that the movement of a glissile interface is a combined deformation and transformation process. The work done by the external stress may be added to the chemical free energy change in order to obtain the net free energy difference.

The mechanical driving force is assumed to be given as the work done ( $\Delta G_{MECH}$ ) by the

external stress in producing the macroscopic shape deformation:

$$\Delta G_{MECH} = \sigma_N \delta + \tau s \quad (1)$$

where  $\sigma_N$  is the normal stress on the habit plane and  $\tau$  is the component of the shear stress on the habit plane which is parallel to the direction along which the shear displacements of the shape deformation occur (Fig. 4a).

The strains  $\delta$  and  $s$  are the dilatational and shear components of the shape deformation. Some typical values of the mechanical driving force terms are given in Table 2. Given a free choice of some 12 to 24 crystallographic variants of the transformation product in each grain of austenite, the work done by the shear stress is always expected to be positive, whereas that due to the dilatational component depends on the sign of  $\sigma_N$ . For steels, this latter component is relatively small. Any observed consequences of stress must therefore reflect the dominant role of the shear component unless the stress is purely hydrostatic.

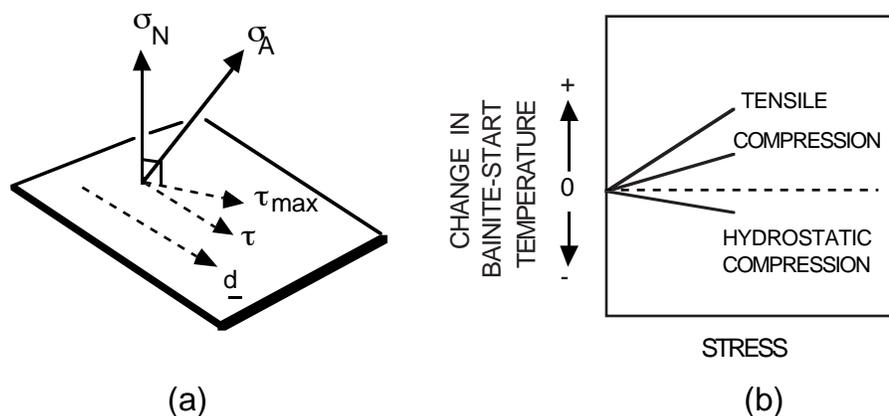


Fig. 4: (a) Resolution of the applied stress  $\sigma_A$ . The normal stress  $\sigma_N$ , and the shear stress  $\tau$ , both act on the habit plane.  $\underline{d}$  is the direction along which the shear displacements of the shape deformation lie.  $\tau_{MAX}$  is the maximum shear stress on the habit plane, but  $\tau$  is given by resolving  $\tau_{MAX}$  along  $\underline{d}$ . Note that  $\underline{d}$  differs slightly from the displacement vector of the invariant-plane strain, which includes a dilatational component in addition to the shear. (b) Variation of transformation-start temperature with the nature and magnitude of applied stress below yield.

Table II – Typical values of the mechanical driving force coefficients [13]. The stress state for the crack tip is multiaxial, but the coefficient is calculated by expressing the stress in terms of the von Mises equivalent tensile stress.

Nature of Stress	$\partial\Delta G/\partial\sigma$ J mol <sup>-1</sup> MPa <sup>-1</sup>
Uniaxial Tension	-0.86
Uniaxial Compression	-0.58
Elastic crack tip	-1.42

It follows from the Patel and Cohen analysis, that since the shear stress remains positive irrespective of whether the sample is pulled in tension or uniaxially compressed, and since the shear component of the shape change is large, a uniaxial stress will always cause an increase in the temperature for displacive transformations in steels. Hydrostatic stress, on the other hand, has no deviatoric components and consequently only interacts with the dilatational component of the shape change. Thus, hydrostatic compression is expected and found to lead to a decrease in the transformation temperature (Fig. 4b); some data [13] on the sensitivity of the transformation temperature to applied stress are presented in Table 3.

Table III – Sensitivity of transformation temperatures to applied stress [13].

Phase	Nature of Stress	Sensitivity / $\text{K MPa}^{-1}$
Martensite	Pressure	-0.06
Bainite	Pressure	-0.09
Eutectoid	Pressure	-0.011
Martensite	Tensile	+0.06

A reservation to the methodology described above is that shear stresses, unlike pressures, cannot strictly be considered as state variables so that their use in thermodynamic equations can be uncertain [12]. This is unimportant as long as irreversible processes, such as diffusion or the multiplication of dislocations, do not act to relieve any of the shear stresses during the interval in which the experiment is conducted. Thus, the state of the system should not change if the shear stress is altered and then restored to its original value.

A second complicating factor could arise if the stress influences the very nature of the transformation product, either by stimulating the formation of some metastable phase, or by decoupling groups of self-accommodating variants which would form in the absence of stress [12]. This could lead to a large modification of the chemical driving force term. As discussed later, there is some evidence to show that there are significant microstructural changes when bainite, acicular ferrite or martensite grow under the influence of an externally applied stress. Similar effects are expected for Widmanstätten ferrite (and carbides which occur at low temperatures) but they have not yet been confirmed experimentally.

Assuming that the interaction of the applied stress is with the macroscopic shape deformation, the stress must tend to favour the growth of those particular variants for which  $\Delta G_{MECH}$  is maximised. Hence, for a tensile stress, plates which have their habit planes inclined at approximately  $45^\circ$  to the tensile axis will tend to be favoured. The angle will not be exactly  $45^\circ$  because the displacement vector of the IPS is not quite parallel to the habit plane due to the volume change.

It is usually assumed that the applied stress interacts primarily with growth. Its influence on nucleation could lead to different variant selection [12]. Indeed, efforts at predicting the martensitic transformation texture from the crystallographic texture of the austenite are successful only if the stress is assumed to interact with the Bain strain rather than with the IPS shape deformation which is a feature of a macroscopic particle [14]. It is probable that only the Bain strain exists at the nucleation stage.

## PLASTIC STRAIN AFFECTED TRANSFORMATION

We have seen that the shape deformation caused by the formation of bainite can be described as an invariant-plane strain with a large shear (Table I); the invariant-plane is the habit plane of each bainite plate. A planar interface (such as the habit plane) traversing a crystal and having no long-range elastic field produces an invariant-plane strain shape change for *all* transformations [15]. However, when an IPS has a large shear component, it implies at the very least coherency at the parent/product interface, which must be glissile. The bainite-austenite interface is therefore glissile, moves without any diffusion, and like slip dislocations, will be impeded by obstacles to shear. We noted earlier that this is not the case for transformation products such as allotriomorphic ferrite where the barriers to interfacial motion (such as grain boundaries or forest dislocations) are overcome easily by diffusion mechanisms.

The phenomenological theory of martensite crystallography [6,7] has been remarkably successful in explaining the crystallography of Widmanstätten ferrite, bainite and martensite. There is, however, a complication not covered by this theory. The shear strains  $s$  are rather large and may not be accommodated elastically when the austenite is weak. This happens even with martensite [*e.g.* 16] but becomes particularly important as the transformation temperature increases because the yield strength of the austenite then decreases. Consequently, the austenite adjacent to the plate is plastically deformed by the transformation as illustrated in Fig. 5, which is an atomic-force microscope image of the displacements caused when a polished sample of austenite is allowed to transform to bainite. The IPS displacements contain a shear of about  $0.26 \pm 0.02$ . Furthermore, the adjacent austenite has been permanently deformed, giving an overall tent-like appearance to the surface relief.

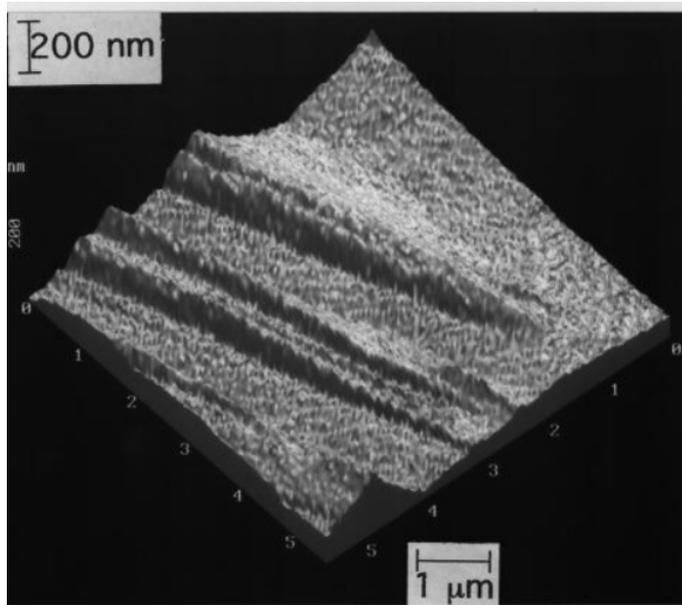


Fig. 5: Atomic force microscope image of displacements at a free surface due to bainite ( $\alpha_b$ ), and deformation in adjacent austenite ( $\gamma$ ) [17].

The transformation therefore does some plastic work on the austenite. The forest of dislocations thus generated eventually blocks the progress of the otherwise glissile transformation interface,

thus limiting the size of the bainite platelets [3]. A glissile interface can be thought of as an array of dislocations which can glide conservatively as the interface moves. Forests of accommodation dislocations would clearly block the motion of such an interface, an effect akin to work-hardening in deformation theory. It is for this reason that individual platelets of bainitic ferrite stop growing before they impinge with an austenite grain boundary. The transformation itself causes plastic deformation which eventually stifles it.

This effect can be demonstrated directly by experiments in which the austenite is deformed prior to transformation. Fig. 6 shows that the size of each of the platelets of bainitic ferrite is obviously reduced as the plastic strain in the austenite is increased. Plastic strain is also thought to increase the nucleation rate of bainite although this has yet to be confirmed quantitatively. It is possible therefore, that at small strains the overall rate of transformation is increased due to the larger number density of nucleation sites (and the small extent of growth retardation). However, at large strains the effect of the forest dislocations is undoubtedly so large that the increased nucleation rate cannot compensate for the greatly reduced size of the bainite platelets. The overall extent of transformation is then greatly reduced – this is known as *mechanical stabilization*, a phenomenon quite familiar in martensitic transformations.

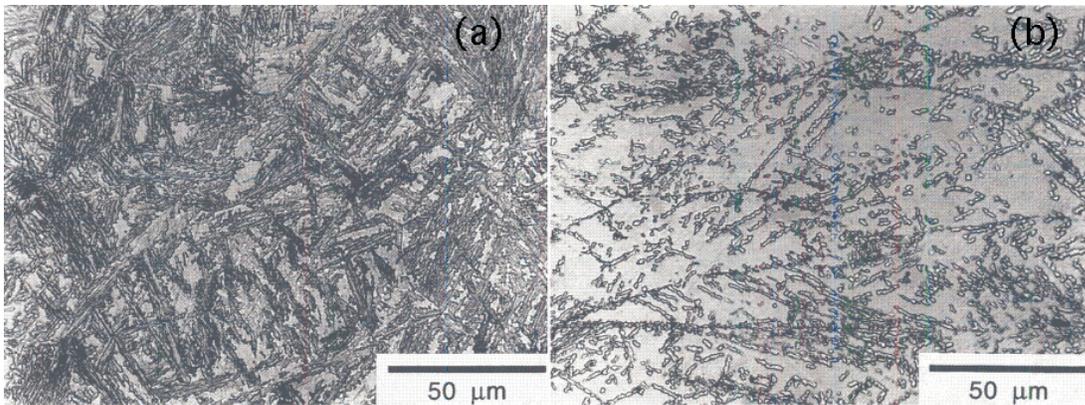


Fig. 6: Mechanical stabilization of bainite using optical microscopy. (a) The microstructure when undeformed austenite is transformed to bainite; the hardness is 470 HV. (b) The microstructure when plastically deformed austenite is transformed to bainite under identical conditions as (a); the hardness is much higher at 650 HV, reflecting the much lower quantity of bainite. [18]

Fig. 7 shows some data reported by Davenport [19], which have been interpreted on the basis of mechanical stabilisation [3]. In these experiments where samples of austenite were hot-rolled before continuous cooling transformation to bainite. The transformation was suppressed to significantly lower temperatures when the austenite was relatively heavily deformed. This highlights the fact that the thermomechanical processing of bainite is quite different from that of the usual steels where allotriomorphic ferrite or pearlite form; in such cases, transformation is always accelerated by deformation.

To summarise, plastic strain in the austenite prior to transformation has the effect of blocking the growth of individual platelets of bainite. There will therefore be a refinement of the

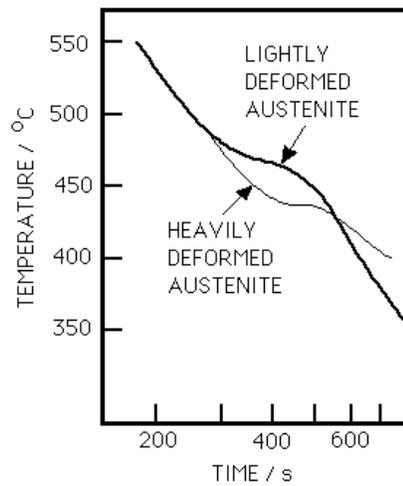


Fig. 7: Plots of temperature versus time, for samples undergoing bainitic transformation during cooling. The deviations from linearity indicate the onset of transformation. The reaction is retarded in the austenite deformed to a greater degree before transformation, due to mechanical stabilisation. Data from Davenport [19].

microstructure due to the stifling of growth. This is quite unlike the effect on allotriomorphic ferrite where every single experiment indicates that the transformation is in all respects accelerated in plastically strained austenite. *For displacive transformations, deformation does not always accelerate transformation.*

## MICROSTRUCTURAL EFFECTS

### *Elastically Strained Austenite*

An applied stress is expected to favour the development of crystallographic variants which comply with that stress. This is analogous with the deformation of single-crystals, where the preferred slip system is that on which the resolved shear stress is largest. Assuming that variant selection is similarly controlled by the interaction of the applied stress with the shape deformation, the stress should cause an alignment of plates at roughly  $45^\circ$  to the tensile axis. Such alignment has been observed in many experiments involving martensitic transformations [e.g. 9]. Good evidence is also available for bainite when the stress is relatively large [3].

There are more subtle effects of stress on microstructure, even in the absence of any obvious plate alignment, at stress levels as small as 45 MPa. Variant selection leads to the development of a less chaotic microstructure [3,10]. Each grain of austenite tends to transform to many different orientations of bainite, in the absence of an applied stress. Significantly fewer variants develop during stress-affected transformation, so that the selected variants grow unhindered and form thick packets of bainite plates. The sheaves are therefore found to be longer, and their number density per grain smaller when variant selection operates (Fig. 8).

### *Anisotropy of Transformation Plasticity: Polycrystalline Austenite*

In a polycrystalline specimen transforming freely, the bainite plates will form in all directions on average (Fig. 2b,c). Consequently, the shear components of the different plates should

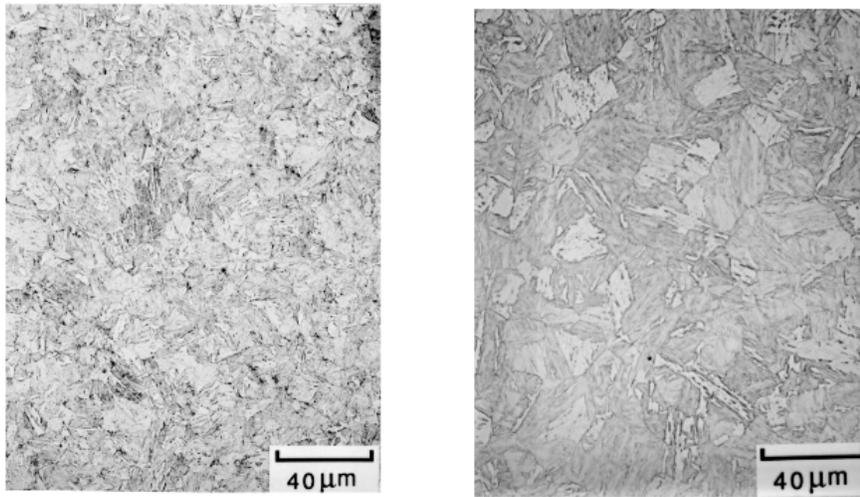


Fig. 8: Light micrographs of bainite in a power plant steel. (a) Transformed at zero stress. (b) Transformed under the influence of a tensile stress of 95 MPa. After [20].

on average cancel out in a fine grained polycrystalline sample containing plates in many orientations. However, if the stress favours the development of an organised microstructure as described above, then the shear component rapidly begins to dominate the transformation plasticity, which will be highly anisotropic (Fig. 2d)[21].

#### *Plastically Strained Austenite*

The transformation to bainite involves a coordinated motion of atoms. Such a motion cannot be sustained across grain boundaries, but small misorientations of the type caused by the presence of dislocation cells inside deformed austenite, can be tolerated. The bainitic ferrite then inherits the parent dislocation structure and there are corresponding changes in its crystallographic orientation and growth direction as well. This is illustrated in the optical micrograph shown in Fig. 9, where the bainite sheaves appear on a macroscopic scale to be curved [22]. They are in fact following the deformation-induced misorientations within the austenite grains.

This is again in contrast to reconstructive transformations such as allotriomorphic ferrite or pearlite; by their very nature, the new ferrite is *reconstructed* by the diffusion of atoms, so that neither the defects nor the cell-misorientations are found in the ferrite that grows anew from the austenite.

## SUMMARY

It has not been possible in this short review to do justice to the bainite transformation, for circumstances where it forms in less than perfect austenite. Nevertheless, it should be evident that the thermomechanical processing of bainitic steels has to be treated differently from that of conventional ferritic steels. The essential reason for the difference is the coordinated motion of atoms during the bainite transformation. The resulting displacements make the transformation far more sensitive to elastically or plastically deformed austenite and give a greater potential for microstructural control.

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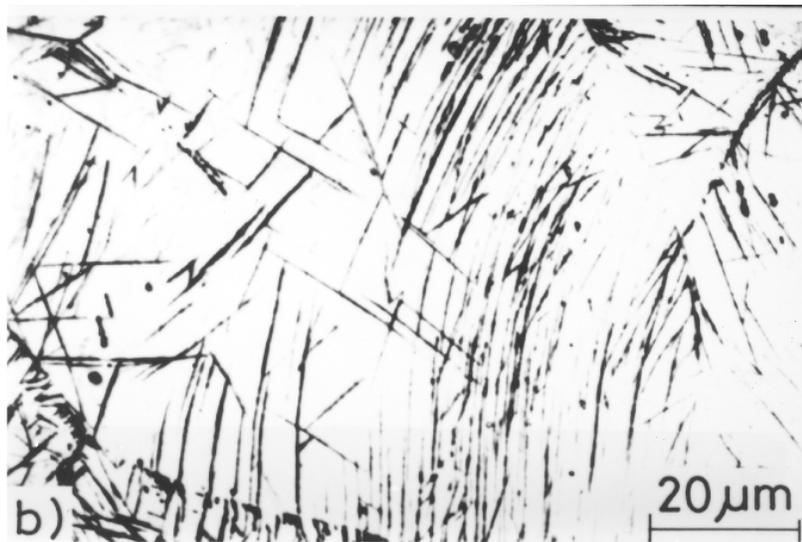


Fig. 9: “Curved” bainite grown in deformed austenite [22].

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