8 *Stress and Strain Effects*

A displacive transformation can justifiably be regarded as a mode of deformation of the parent phase, with the additional characteristic that the crystallographic structure of that phase is altered in the deformed region (Table 8.1). For this reason, the permanent strain is called *transformation plasticity*. A phase transformation can be triggered by cooling below a certain transformation-start temperature, by the application of a stress in appropriate circumstances or by a combination of these factors. In the latter case, where the chemical driving force and stress act in concert, transformation plasticity can be obtained at stresses which are much smaller than the yield strength of the stable parent phase.

Table 8.1 Characteristics of different modes of deformation.

	Slip deformation	Mechanical twinning	Displacive transformation	Reconstructive transformation
Permanent deformation?	Yes	Yes	Yes	Yes
Invariant-plane strain with large shear?	Yes	Yes	Yes	No
Crystallographic orientation altered?	No	Yes	Yes	Yes
Lattice change?	No	No	Yes	Yes
Density change?	No	No	Yes	Yes

Just as a combination of a plane and a direction constitutes a deformation system for slip or twinning, the habit plane and displacement vector of an invariant-plane strain describe the deformation system for transformation plasticity. There will in general be 24 of these systems per austenite grain and they may operate simultaneously with varying contributions. Unlike ordinary slip, the different variants of transformation cannot intersect except in special circumstances where intervariant transformations are possible. The ordinary notion of work hardening does not therefore apply. Work hardening nevertheless manifests itself via an increase in the stability of the austenite as it becomes more finely divided.

Given the large number of transformation variants available per grain, the Taylor criterion leads to the conclusion that transformation plasticity can cause or accommodate any externally imposed, arbitrary shape change assuming that there is sufficient austenite available to cope with the imposed strain. It follows that polycrystalline samples can remain intact at grain boundaries when transformation plasticity is the sole mode of deformation. Furthermore, the transformation plasticity can cause anisotropic changes in shape even in polycrystalline samples transformed without applied stress if the parent phase is crystallographically textured.

8.1 The Mechanical Driving Force

Given that displacive transformations in steels cause large strains, it is natural to expect an interaction between any applied stress and the progress of the transformation, in a manner which is related uniquely to the transformation mechanism. The total driving force can be partitioned into a *mechanical* and the more usual *chemical* components (Patel and Cohen, 1953; Delaey and Warlimont, 1975; Christian, 1982). The physical reasoning behind this idea is 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 total free energy difference. The mechanical driving force is assumed to be given by the work done (ΔG_{MECH}) by the external stress system in producing the macroscopic shape deformation:

$$\Delta G_{MECH} = \sigma_N \zeta + \tau s \tag{8.1}$$

where σ_N is the normal stress on the habit plane and τ 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. 8.1). The strains ζ and s have previously been defined as the dilatational and shear components of the shape deformation. Given a free choice of some 12 to 24 crystallographic variants of the transformation product in a 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 σ_N. For steels where this latter component is small, the observed stress effects reflect the dominant role of the shear component. The exception is when τ is small or zero, as would be the case when the applied stress is a hydrostatic pressure.

It follows from the equation 8.1, 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

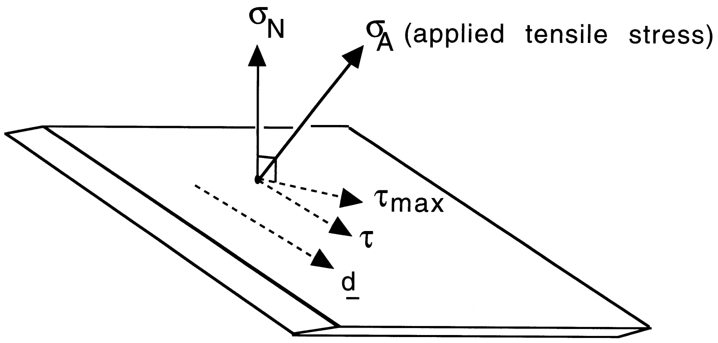

Fig. 8.1 Resolution of the applied stress σ_A. The normal stress σ_N, and the shear stress τ, both act on the habit plane. The vector $\underline{d}$ is the direction along which the shear displacements of the shape deformation lie. τ_{MAX} is the maximum shear stress on the habit plane, but τ is given by resolving τ_{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.

uniaxial stress will always cause an increase in the transformation 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 to lead to a decrease in the transformation temperature (Fig. 8.2).

Shear stresses, unlike pressures, cannot strictly be considered as state variables so their use in thermodynamic equations is uncertain (Christian, 1982). This difficulty is unimportant provided irreversible processes such as diffusion or dislocation motion do not act to relieve the shear stresses during the time scale of the experiment. In practice, this means that in the absence of transformation, the state of the system should not be altered if the shear stress is changed 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 a metastable phase or by decoupling groups of self-accommodating variants which would form in the absence of stress (Christian, 1982). This would lead to a modification of the chemical driving force term, and as discussed later, there is some evidence to show that there are significant microstructural changes when bainite grows under the influence of an externally applied stress.

Assuming that the interaction of the applied stress is with the macroscopic shape deformation, the stress must favour the growth of those variants for which ΔG_{MECH} is maximised. Hence, for a tensile stress, plates which have their habit planes inclined at approximately 45° to the tensile axis will tend to

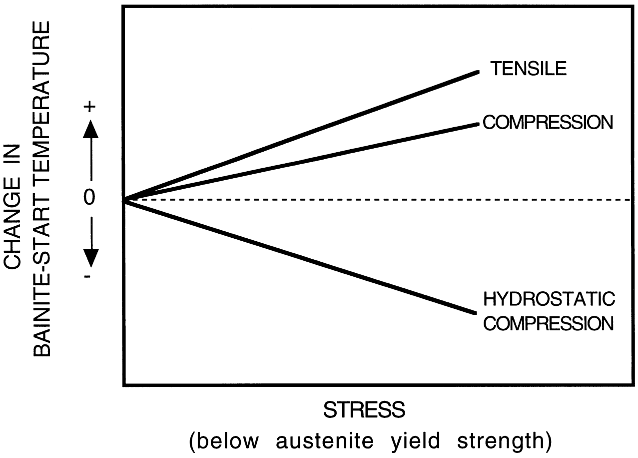

Fig. 8.2 An indication of how the bainite-start temperature should vary as a function of the nature and magnitude of the applied stress.

be favoured.[†] This does assume that the applied stress interacts solely with the growth process whereas its interaction with nucleation events could lead to a different criterion for variant selection (Christian, 1982). Indeed, efforts at predicting the martensitic transformation texture from the crystallographic texture of the parent austenite, are apparently more successful if it is assumed that variant selection depends on the Bain strain rather than on the macroscopic shape deformation (Ray and Jonas, 1990). The IPS deformation is unlikely to have developed at the nucleation stage, where the particle might be too small to sustain a lattice-invariant deformation. The Bain strain is essential to accomplish the lattice change, so the texture prediction work suggests that variant selection may depend on the interaction of the applied stress with the nucleation process.

8.2 The B_d Temperature

The highest temperature at which martensite forms during the cooling of austenite is the M_S temperature. This can be increased by the application of a suitable stress (Patel and Cohen, 1953). The maximum temperature at which martensite grows under the influence of stress is called the M_d temperature.

[†] The angle will not be exactly 45° because the displacement vector of the IPS is not quite parallel to the habit plane whenever ζ is finite.

There are no similar experiments for bainite but it is possible to piece together evidence to show that the behaviour is similar to that of martensite.

The transformation stresses associated with the growth of lower bainite are known to stimulate upper bainite at temperatures just above B_S, proving that stress can indeed raise the bainite-start temperature (Goodenow *et al.*, 1969). It should in principle be possible to define a B_d temperature. Thus, Drozdov *et al.* (1962) found that no amount of deformation causes the austenite to transform to bainite when the temperature is sufficiently greater than B_S, i.e.$T > B_d$.

The expected stress effects are illustrated in Fig. 8.3, which is based on similar ideas for martensitic transformations. The net driving force available for transformation, ΔG, is given by:

$$\Delta G = \Delta G_{CHEM} - \Delta G_{MECH} \tag{8.2}$$

It is assumed that the critical value needed to trigger bainitic transformation at zero stress (i.e. $\Delta G_{CHEM}\{B_S\}$) remains constant over the temperature range of

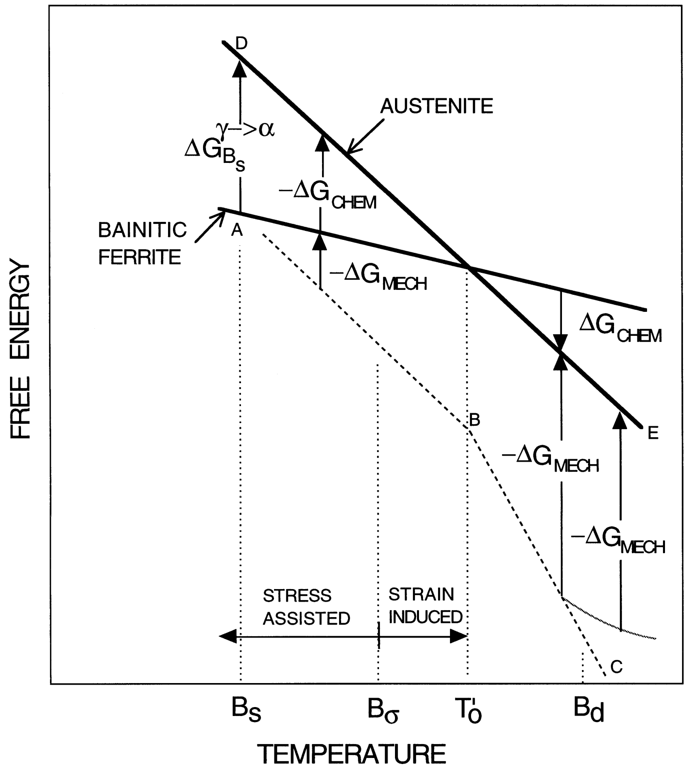

Fig. 8.3 Illustration of the stress-modified B_S, B_σ and B_d temperatures.

interest. The application of a tensile or compressive stress assists the transformation by boosting the overall driving force ΔG with the term $-\Delta G_{MECH}$, so that the B_S temperature rises continuously with the magnitude of the applied stress.

Consider a temperature B_σ corresponding to an applied stress σ, where σ becomes greater than the yield strength of the austenite. It is difficult to justify a thermodynamic analysis when the austenite undergoes plastic deformation prior to transformation. The dislocations and other defects generated during plastic deformation will nevertheless influence the progress of transformation. Following the terminology established for martensitic transformations, the region below B_σ is said to represent *stress-assisted* transformation, whereas *strain-induced* transformation describes the regime where the yield stress of the parent phase is exceeded.

The B_S temperature continues to increase as the stress is raised beyond the yield stress of the austenite. When the T_0' temperature is reached, the chemical driving force *opposes* transformation so that the mechanical component has to be larger than $\Delta G_{CHEM}\{B_S\}$. The yield strength of austenite is smaller at high temperatures so a point is reached where the austenite can no longer support a stress large enough to stimulate bainitic transformation; that temperature is B_d (Fig. 8.3).

8.3 General Observations

8.3.1 *Externally Applied Stress*

There are many independent observations which suggest that stress has a large effect on the progress of transformation (Fig. 8.4). Deformation during the thermomechanical processing of steels accelerates the rate of the bainite reaction.[†] There rate of reaction also increases with the rate of deformation (Drozdov *et al.*, 1962; Mutui *et al.*, 1977). A tensile stress during transformation even stimulates bainite beyond that expected from the T_0 condition (Cottrell, 1945).

8.3.2 *Internally Generated Stress*

The stress influencing transformation need not be applied externally. Internal stresses generated by other transformations also have an effect. Early studies

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[†] Cottrell, 1945; Jepson and Thompson, 1949; Drozdov *et al.*, 1962; Duckworth, 1966; Dubrov, 1969; Freiwillig *et al.*, 1976; Mutui *et al.*, 1977; Umemoto *et al.*, 1986a; Tsuzaki *et al.*, 1989; Yang *et al.*, 1995, 1996; Larn and Yang, 2000.

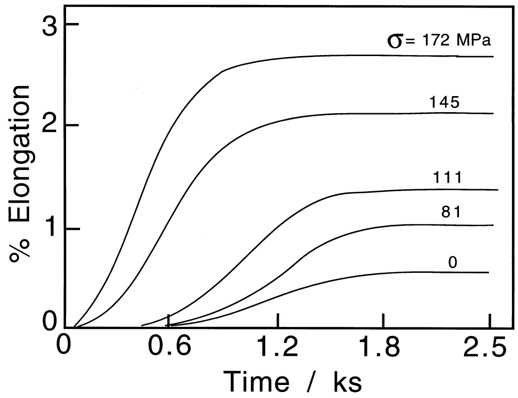

Fig. 8.4 The overall kinetics of bainitic transformation as a function of an externally applied tensile stress. Assuming that the degree of transformation is related to the elongation, the data show an increase in the rate of reaction as a function of the magnitude of the applied stress (after Umemoto *et al.*, 1986a).

indicated an acceleration in the rate at which upper bainite forms in specimens which are first transformed partially at a lower temperature (Lange and Mathieu, 1938; Jellinghaus, 1952). Martensite is the first phase to form on cooling a steel below the M_S temperature, but after the initial burst of transformation and a suitable incubation period, the austenite undergoes accelerated decomposition to bainite (Howard and Cohen, 1948). This is because it is deformed by the martensitic transformation. Supporting evidence is found in magnetometric studies, which have revealed that isothermal reaction below the M_S temperature leads first to the formation of the usual athermal martensite, followed by a small amount of isothermal martensite, an accelerated decomposition to bainite (Ericsson *et al.*, 1976). Similar results have been obtained by Radcliffe and Rollason (1959) and it has been shown that the upper bainite reaction is accelerated in the presence of lower bainite (Fig. 8.5).

A revealing observation is that both the nucleation and growth rates of bainite are accelerated by the proximity of a free surface (Ko, 1953; Hawkins and Barford, 1972). This is because the shape change can be accommodated better at a free surface where the constraint is reduced.

8.4 Plastic Deformation and Mechanical Stabilisation

It has been emphasised that displacive transformations involve the coordinated movement of atoms and that such movements cannot be sustained against strong defects such as grain boundaries. Thus, martensite plates, which form by a displacive mechanism, cannot cross austenite grain bound-

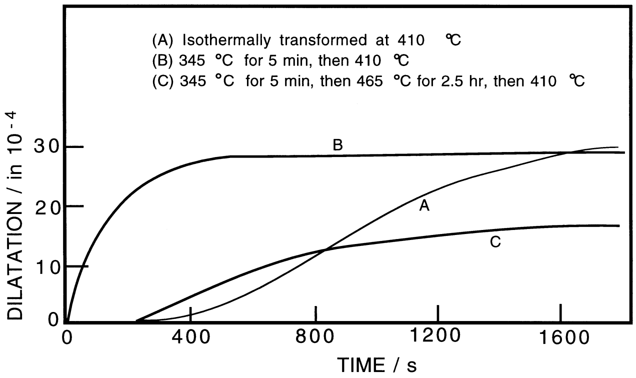

Fig. 8.5 The influence of internal stresses on the rate of transformation at 410 °C, in a Fe–0.31C–0.3Si–0.76Mn–3.07Ni–1.22Cr–0.49Mo wt% alloy. Curve A represents isothermal transformation to upper bainite; curve B is for a sample which was first partially transformed to lower bainite and then to upper bainite, showing an acceleration of reaction rate at 410 °C due to the internal stresses generated by the presence of lower bainite; curve C shows how annealing above the B_S temperature removes the stresses, and their accelerating influence on transformation kinetics (Goodenow *et al.*, 1969).

aries. Smaller defects such as isolated dislocations hinder the progress of such transformations, but can often be incorporated into the martensite lattice. However, severe deformation of austenite prior to its transformation hinders the growth of martensite, causing a reduction in the fraction of transformation in spite of an increased number density of nucleation sites. This applies to all martensitic transformations, irrespective of material; apart from steels, the phenomenon is, for example, known to occur for martensitic transformations in lithium (Maier *et al.*, 1997), in brass (Spielfeld, 1999) and during mechanical twinning (Christian and Mahajan, 1995).

This retardation of transformation by plastic deformation is called *mechanical stabilisation* and can be explained in terms of the structure of the transformation interface. Displacive transformations occur by the advance of glissile interfaces which can be rendered sessile when they encounter dislocation debris. Thus, whereas an appropriate stress can stimulate displacive transformation in the same way that it enables normal deformation, mechanical stabilisation actually retards the decomposition of the austenite (Bhadeshia, 1999).

Most of the work on mechanical stabilisation effects has been on martensitic transformations with few studies on bainite. Some early experimental data based on hot-rolling experiments can be interpreted to show that bainitic transformation is retarded in deformed austenite, as illustrated in Fig. 8.6.

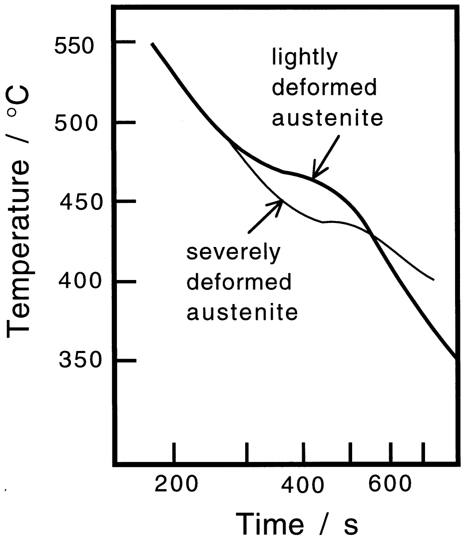

Fig. 8.6 Plots of temperature versus time for samples undergoing bainitic transformation during cooling. The deviations from linearity indicate the onset of transformation. The reaction is retarded in the austenite deformed to a greater degree before transformation, indicative of mechanical stabilisation (data from Davenport, 1977).

Direct evidence comes from the work of Tsuzaki *et al.* (1989) who found that although deformed austenite transformed faster, the net volume fraction of bainite decreased when compared with undeformed austenite, Fig. 8.7. This effect did not occur at higher temperatures, presumably because the amount of bainite that can form is then reduced. Stabilisation therefore only manifests itself when the 'easy' regions of austenite are exhausted, i.e. those regions left unaffected by the imposed deformation which is inevitably inhomogeneous. The nonuniformity of stabilisation is reflected in the microstructure. The bainite tends to align along specific directions within individual austenite grains (Fig. 8.8).

As is often the case with martensite in ausformed alloys, bainite plates sometimes follows a curved path. This is because of the deformation-induced lattice curvature present in the parent austenite grains prior to transformation (Fig. 8.8).

In recent work it has been demonstrated using metallography that the bainite transformation can be mechanically stabilised in a manner identical to the mechanical stabilisation of martensite in steels (Fig. 8.9). The mechanism appears to be that the growth of bainite is retarded by the deformation debris in the austenite. Heterogeneous nucleation becomes more frequent as defects

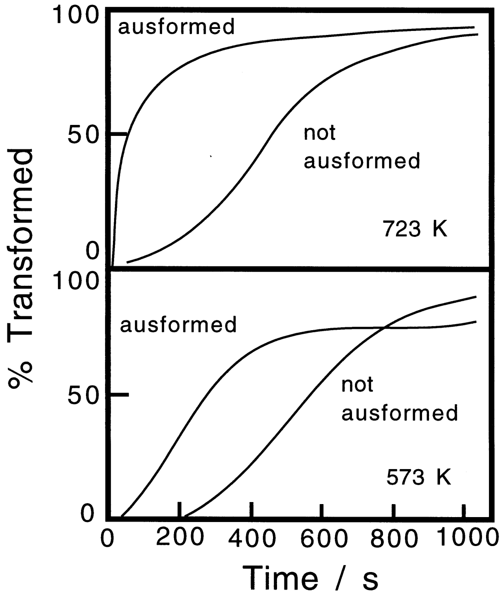

Fig. 8.7 The effect of ausforming on the kinetics of the bainite reaction in a Fe–0.59C–2.01Si–1.02Mn wt% alloy.

are introduced into the austenite, but their growth by a displacive mechanism is stifled as the interface encounters forests of dislocations. Heavily deformed austenite therefore transforms to a smaller quantity of bainite than undeformed austenite, and any bainite that forms is more refined.

Mechanical stabilisation is evident in quantitative experiments (Singh and Bhadeshia, 1996; Larn and Yang, 2000). There are two intriguing features illustrated in Fig. 8.10, first that transformation from deformed austenite leads to a smaller terminal fraction of bainite. Secondly, although the transformation rate is at first accelerated by deformation, it is eventually retarded relative to the undeformed sample. If this initial acceleration is explained by increasing the number density of nuclei then it is not possible to reach a smaller terminal fraction given that each nucleus transforms a fixed volume of austenite. On the other hand, if it is assumed that the smaller limiting fraction is due to the reduction in volume transformed per nucleus, then it is not possible to explain the initial acceleration. There are other complications described elsewhere, all of which can only be resolved by arguing that the austenite is inhomogeneously deformed (Singh, 1998). The lightly deformed regions transform more rapidly relative to undeformed austenite because of the increase in the defect density. The nucleation rate is larger in the heavily

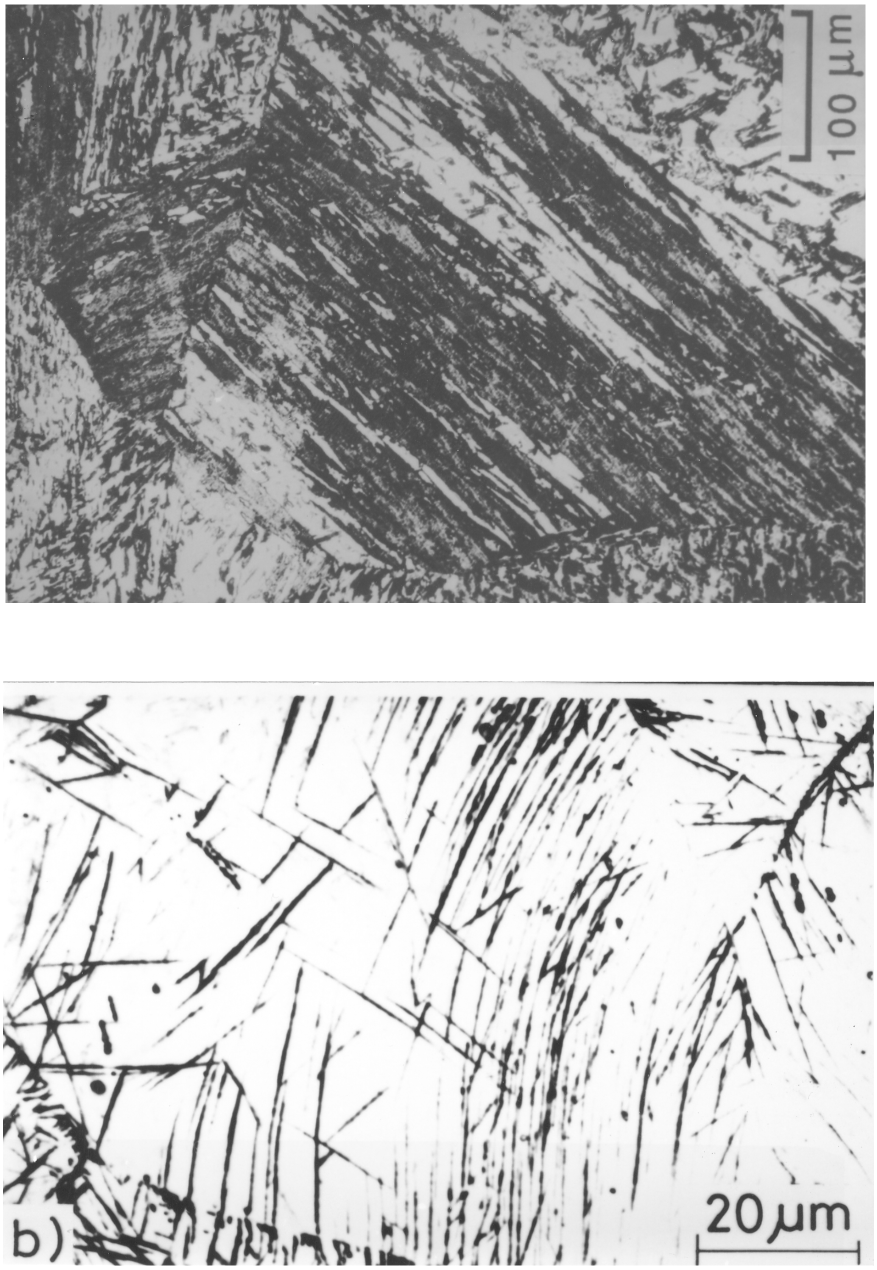

Fig. 8.8 Optical micrographs illustrating the microstructure of an ausformed bainitic steel: (a) alignment of sheaves of bainite in individual austenite grains; (b) curved bainite sheaves reflecting the deformation-induced misorientations within the austenite grains (Tsuzaki *et al.*, 1989).

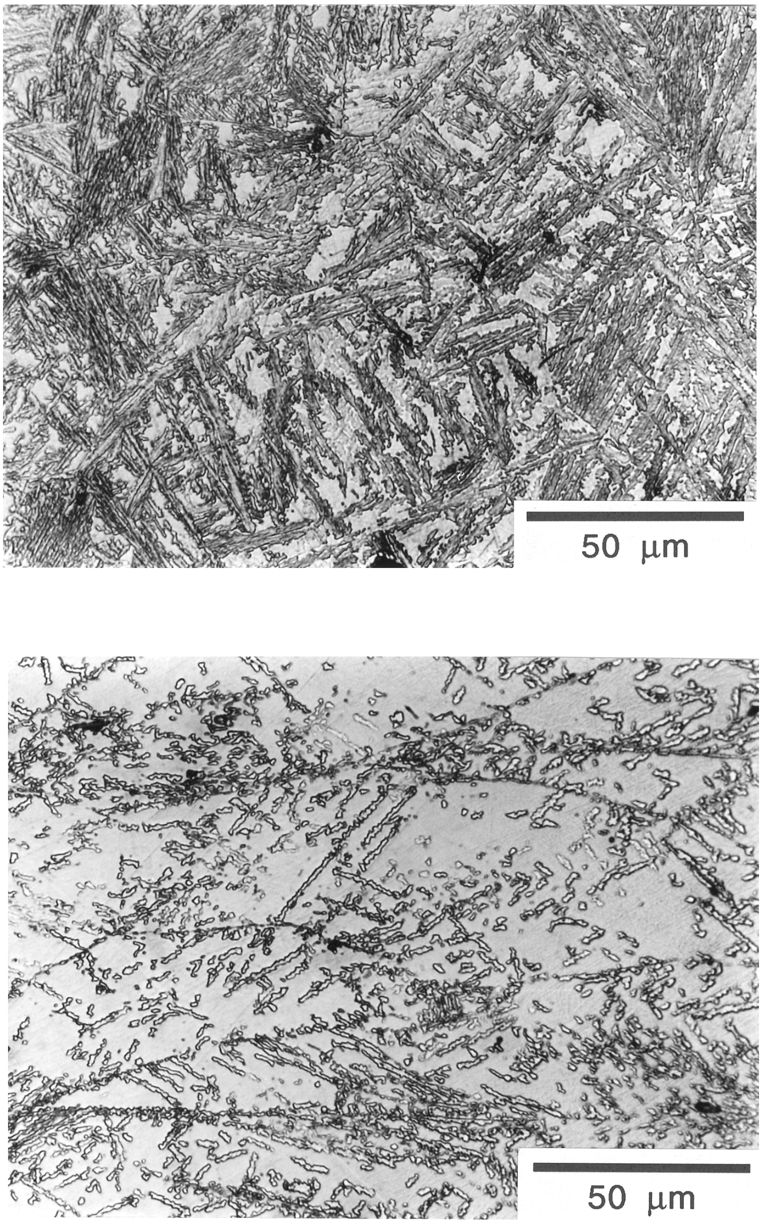

Fig. 8.9 Optical micrographs showing the large effect of mechanical stabilisation in refining the microstructure and in reducing the amount of bainite: (a) transformation from undeformed austenite; (b) transformation from plastically deformed austenite (Shipway and Bhadeshia, 1995).

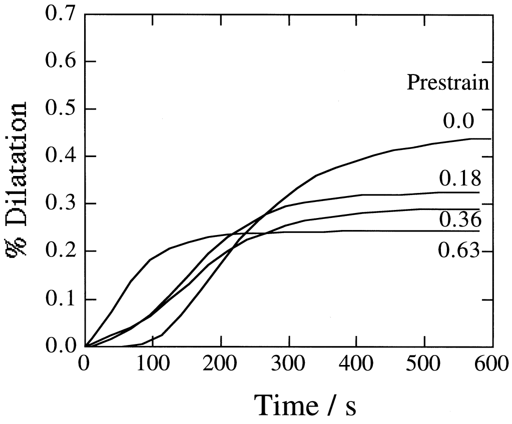

Fig. 8.10 Change in radial dilatation during isothermal transformation to bainite as a function of time and prestrain (the strain in the austenite prior to transformation); values of the prestrain are indicated next to individual curves. After Singh and Bhadeshia, 1996.

deformed regions but the overall rate of transformation is reduced because each nucleus then transforms to a smaller volume due to mechanical stabilisation of the interface.

These qualitative ideas need to be developed and backed by direct metallographic observations of the distribution of bainite sub-unit sizes, which should be bimodal. The importance of such work cannot be overemphasised given the increasing use of thermomechanical processing of bainitic steels.

Mechanical stabilisation has been found for all of the plate-shaped ferritic phases that occur in steels. This includes Widmanstätten ferrite (Shipway and Bhadeshia, 1997; Larn and Yang, 1999, 2000), bainite and martensite, all of which are displacive shear transformations. Mechanical stabilisation has been shown to occur in materials as diverse as lithium (Pichl and Krystian, 1999) and brass (Hornbogen, 1999). By contrast, reconstructive transformations are without exception accelerated if the parent phase is deformed prior to transformation. This is because of the increase in the number density of nucleation sites, and because the defects introduced by deformation are destroyed as the new phase grows, rather as in recrystallisation. The elimination of the defects contributes to the driving force for reconstructive transformation. In displacive transformations the defects are *inherited* by the growing phase and hence do not supplement the driving force. With these general observations it is possible to define a disarmingly simple criterion to distinguish these two mechanisms of transformation:

There is no mechanism by which plastic deformation can retard reconstructive transformation. Likewise, only displacive transformations can be mechanically stabilised.

8.4.1 Technological Implications of Mechanical Stabilisation

There are now many structural steels which have a bainitic microstructure but are manufactured using the same thermomechanical processing routes that have been applied so successfully to the ferrite–pearlite steels (Chapter 13). However, this has been done without the realisation that whereas the ferrite and pearlite reactions are accelerated by deforming the austenite, the bainite transformation can be retarded by the same action. The consequences of this for structural steels have simply not been explored.

It is possible to deduce evidence from the published literature of the consequences of mechanical stabilisation in commercial bainitic steels. Tsuji *et al.* (1999) found that the effect of forcing the bainite to grow in severely deformed austenite is to increase the quantity of untransformed austenite. This is precisely what is expected from mechanical stabilisation. Furthermore, they observed that although ferrite and pearlite are refined, their hardness does not increase greatly because they grow from deformed austenite. A much bigger increase in hardness was observed for the bainite. These observations are expected since ferrite and pearlite, both of which are reconstructive transformations, do not inherit the defect structure of the deformed austenite. The bainite on the other hand, acquires all the crystallographic errors present in the deformed austenite since its growth does not involve any diffusion.

8.5 The Effect on Microstructure

An applied stress will tend to favour the development of crystallographic variants which comply with that stress. This is analogous to the selective operation of a few of the available slip systems in a crystal under stress; it is the systems with the largest Schmid factors which are favoured. 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 the plates at roughly 45° to the tensile axis. This alignment has been observed in many experiments involving martensitic transformations (e.g. Bhadeshia, 1982a). The observations are more difficult for bainite, partly because of the rapid rate of reaction under the influence of stress. The experiments have to be conducted at high temperatures. Further transformation may occur as the sample cools to

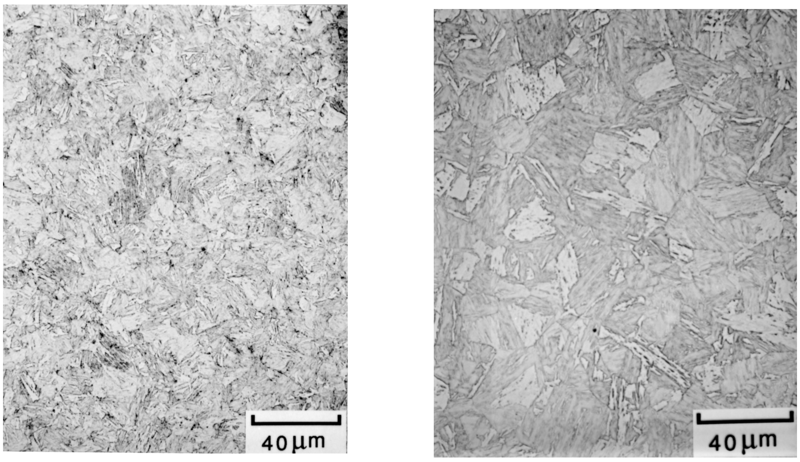

Fig. 8.11 Light micrographs of bainitic microstructures generated in a Fe–0.12C–0.27Si–0.84Mn–0.14Ni–1.48Mo–2.86Cr wt% alloy, by isothermal transformation at 400 °C under the influence of stress. (a) Zero stress; (b) 95 MPa.(after Bhadeshia *et al.*, 1991).

ambient temperature, confusing the interpretation of the microstructure. Nonetheless, good evidence for microstructural alignment has been reported for bainite platelets especially at relatively large stresses (Bhattacharyya and Kehl, 1955; Umemoto *et al.*, 1986a). All of these observations are based on polycrystalline samples, but that does not substantially alter the conclusions. There are so many variants available per austenite grain that there is a high probability of a plate orientation lying close to the optimum orientation with respect to the stress.

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 (Jepson and Thompson, 1949; Dubrov, 1969; Bhadeshia *et al.*, 1991). Without stress, each grain of austenite transforms into many different orientations of bainite. Fewer variants develop per austenite grain under the influence of stress, so that the selected orientations can grow unhindered and form thick packets of bainite plates. The sheaves then are longer and their number density per grain smaller when variant selection operates (Fig. 8.11).

A further effect on microstructure is when the austenite has been plastically deformed prior to transformation. Heterogeneous nucleation then occurs not only at the original austenite grain boundaries, but apparently also intragranularly on slip bands or other deformation heterogeneities (Dubrov, 1969).

8.6 The Effect of Hydrostatic Pressure

There is general agreement that the application of hydrostatic pressure causes a retardation of the bainite reaction (Jellinghaus and Friedewold, 1960; Radcliffe *et al.*, 1963; Nilan, 1967). The effect on the time–temperature–transformation diagram is illustrated in Fig. 8.12. The observed retardation is not in itself a feature unique to bainite. All transformations which are accompanied by a reduction in density are expected to be retarded by hydrostatic pressure, which opposes a volume expansion. The effect of hydrostatic pressure is two fold: it reduces the diffusion coefficients by decreasing the available free volume (although the details remain to be established), and it influences the free energy change for transformation. If ΔG_m is the molar Gibbs free energy change for a reaction, then since

$$\left(\frac{\partial G}{\partial P}\right)_T = V$$

it follows that:

$$\Delta G_m\{P\} - \Delta G_m\{1\} = \int_1^P \Delta V_m \, dP' \tag{8.3}$$

where ΔV_m is the change in molar volume on transformation, V is the volume and P is the pressure. The way in which the free energy change for transformation is influenced by the pressure determines how the transformation temperature changes as a function of pressure. An alternative way of expressing this relationship is the Clausius–Clapeyron equation, whence the change in transformation temperature is given by

$$dT^{\gamma\alpha} = T^{\gamma\alpha} \Delta V_m / \Delta H^{\gamma\alpha} \tag{8.4}$$

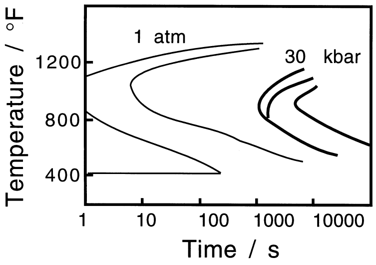

Fig. 8.12 Isothermal transformation diagrams of Fe–0.82C wt% at 1 atmosphere and at 30 kbar pressure (after Nilan, 1967).

where $\Delta H^{\gamma\alpha}$ is the enthalpy change on transformation at the transition temperature $T^{\gamma\alpha}$. The equation is approximate in that its derivation depends on the assumption that the enthalpy change does not vary with temperature. With typical values of all the parameters, the variation in transition temperature with pressure should be approximately $-0.01\,\mathrm{K\,MPa^{-1}}$ (Denis *et al.*, 1985).

Radcliffe *et al.* also found that the bainite transformation could be suppressed completely by the application of hydrostatic pressure ($\simeq 15$ kbar) but Nilan, using similar steels could obtain conventional bainite at the maximum pressures he used (34 kbar). Why these experiments are contradictory is not clear, but Nilan concluded that the transformations at high pressures do not differ substantially from those at ambient pressure.

8.7 Mechanical Stability of Retained Austenite

In steels where the precipitation of carbides during the bainite reaction is slow, the residual austenite becomes enriched in carbon and a large proportion is retained on cooling to ambient temperature. The austenite, if it decomposes under the influence of stress, can be detrimental to the steel concerned since the

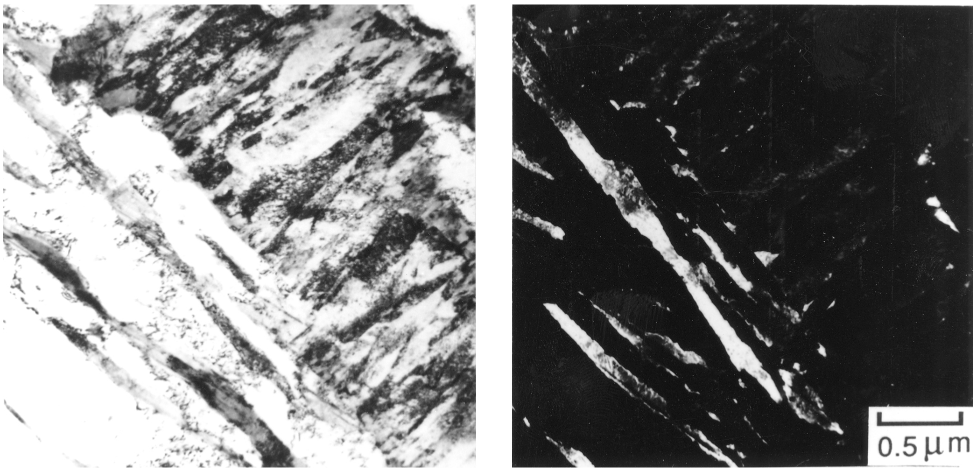

Fig. 8.13 Electron micrographs illustrating the effect of applied stress (850 MPa) on a sample which initially had a microstructure of bainitic ferrite and retained austenite (Bhadeshia and Edmonds, 1983a). The larger regions of austenite transform to martensite but the films are preserved. (a) Bright field image showing a large region of stress-induced martensite; (b) corresponding austenite dark field image. The sample was tempered prior to the application of stress in order to distinguish the martensite which forms during cooling from the bainite transformation temperature, from that which is induced by stress.

resulting high-carbon, untempered martensite is expected to be brittle. There is ample evidence that the austenite retained to ambient temperature after isothermal formation of bainitic ferrite, especially the larger blocky austenite, can decompose to martensite even at relatively small stresses, Fig. 8.13 (Horn and Ritchie, 1978; Kar *et al.*, 1979; Bhadeshia and Edmonds, 1983a,b; George *et al.*, 1985; Tsukatani *et al.*, 1991). The mechanical stability of retained austenite is therefore important in obtaining good toughness in bainitic steels.

Miihkinen and Edmonds (1987b) have shown that for high silicon steels in which the bainite reaction is allowed to proceed until it stops, the mechanical stability of the retained austenite decreases as the isothermal transformation temperature is increased. The mechanical stability was defined as the ratio of retained austenite content after 2% plastic deformation in a tensile test, to the original content. Given that the bainite reaction in such steels ceases when the carbon concentration of the residual austenite x_γ approaches x_{T_0}, and that x_{T_0} increases with decreasing temperature, the austenite on the basis of its composition is theoretically expected to be more stable as the bainite formation temperature is reduced (Bhadeshia and Edmonds, 1983a,b). Furthermore, if the T_0 curve can be shifted to higher carbon concentrations by modifying the substitutional solute content then the stability of the austenite is expected to increase, and this has also been confirmed experimentally.

8.8 Transformation under Constraint: Residual Stresses

Residual stresses are mostly introduced unintentionally during fabrication. They are of particular importance in welded structures where they have a detrimental effect. Jones and Alberry (1977a,b) conducted an elegant series of experiments to illustrate the interaction between transformations and residual stress. Using bainitic, martensitic and stable austenitic steels, they demonstrated that transformation plasticity during the cooling of a uniaxially constrained sample from the austenite phase field, acts to relieve the build up of thermal stress as the sample cools. By contrast, the non-transforming austenitic steel exhibited a continuous increase in residual stress with decreasing temperature, consistent with the degree of thermal contraction. On the other hand, with the steels which transformed to bainite or martensite, the transformation strain compensated for the thermal contraction strains. Significant residual stresses developed only after transformation was completed, and the specimens approached ambient temperature (Fig 8.14).

The interpretation of experimental data of the kind illustrated in Fig. 8.14 is difficult. The view that the volume change during transformation gives the major contribution to transformation plasticity is almost certainly incorrect for displacive transformations such as bainite. The shape change due to transformation has a shear which is much larger than the volume strain.

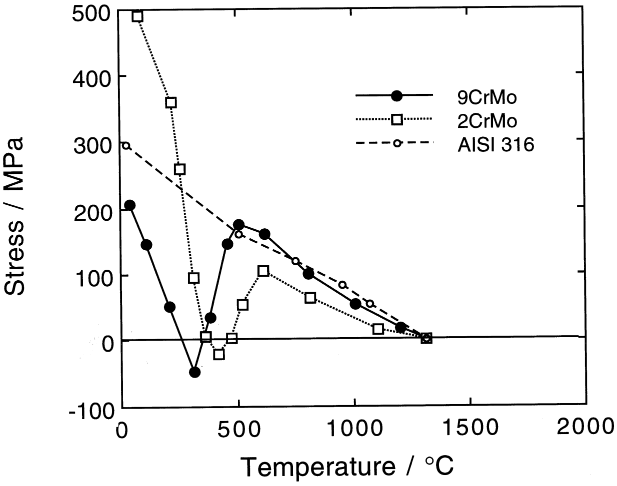

Fig. 8.14 Development of stress as a function of temperature as a constrained sample is cooled from the austenite phase field, for a martensitic (9CrMo), bainitic (2CrMo) and austenitic steel (AISI 316). After Alberry and Jones.

Admittedly, this shear component should on average cancel out in a fine grained polycrystalline sample containing plates in many orientations. However, the very nature of the stress effect is to favour the formation of selected variants in which case, the shear component rapidly begins to dominate the transformation plasticity.

8.9 Anisotropic Strain due to Transformation Plasticity

During their attempts to study the isothermal transformation of austenite using dilatometry, Davenport and Bain (1930) had noticed already that 'the volume change (due to transformation) is not necessarily uniformly reflected in linear change in all dimensions'. They even found that the thickness of flat disc specimens actually decreased as the volume increased! These results were stated without interpretation but it is now clear that in polycrystalline samples which are crystallographically textured, anisotropic transformation plasticity can be detected even in the absence of an applied stress (Bhadeshia *et al.*, 1991). When an unstressed polycrystalline sample of austenite is transformed, the shear components of the individual shape deformations of the large number of variants which form along any dimension should tend to cancel out on a macroscopic scale. Similarly, the dilatational component of the IPS shape deformation should average leaving an isotropic volume expansion. If the sample is not random, i.e. it is crystallographically textured, then the possibility of the individual shape deformations cancelling out over large distances is correspondingly reduced. Transformation will then lead to a large anisotropy in the strains even in the absence of an applied stress (Fig. 8.15).

8.10 Stress-Affected Carbide Precipitation

The idea that cementite at low temperatures precipitates by a displacive mechanism with only the partitioning of carbon is not unnatural – this mechanism has been demonstrated for the precipitation of vanadium hydride (Bowles *et al.*, 1977). The evidence for cementite has been discussed in Chapter 3. Although the shape deformation associated with precipitation has yet to be measured, it is believed to be an invariant-plane strain with a shear of 0.211 parallel to the habit plane and a dilatational strain of 0.157 normal to the habit plane (Taylor *et al.*, 1989b).

The effect of the shape change can be revealed by the precipitation microstructure when it is generated under the influence of an externally applied

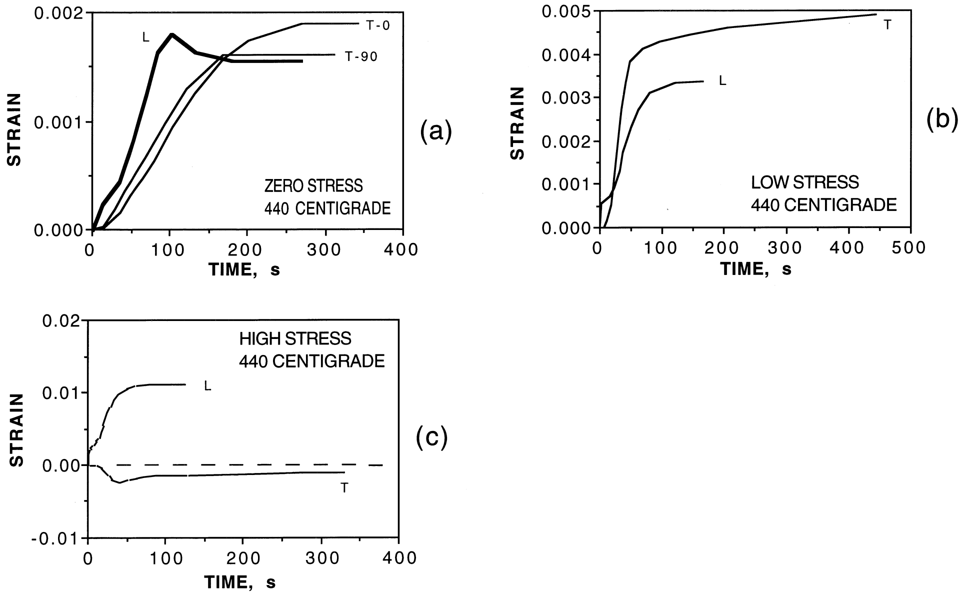

Fig. 8.15 Dilatometric curves showing the dimensional changes during transformation to bainite in a cylindrical sample. T-0 and T-90 refer to the strains monitored along orthogonal transverse directions, and L to the strain along the longitudinal direction: (a) transformation in the absence of an applied stress; (b) transformation under the influence of a tensile stress of about 45 MPa; (c) 90 MPa (Bhadeshia *et al.*, 1991).

stress (Matsuzaki *et al.*, 1992; Stewart *et al.*, 1994). When martensite is tempered in a stress-free condition, the carbides precipitate in several crystallographic variants in any given plate, in the so-called Widmanstätten pattern. When the tempering is carried out under a uniaxial stress, the variant which presumably complies best with the external stress begins to dominate the microstructure. Eventually, when the stress is large enough, it is only a dominant crystallographic variant is found in any plate of martensite (Fig. 8.16).

The response of the carbide microstructure to the applied stress is precisely that expected from the interaction with the transformation strain. In experiments reported in the literature, the mechanical driving force is similar in magnitude to the chemical driving force for the precipitation of cementite. Thus, $\Delta G_{MECH} \simeq 730\,\mathrm{J\,mol^{-1}}$ for a stress of 500 MPa, and $1380\,\mathrm{J\,mol^{-1}}$ for 950 MPa. This compares with $\Delta G_{CHEM} \simeq 1300\,\mathrm{J\,mol^{-1}}$ at the tempering temperature.

Figure 8.17 shows that the chemical driving force is sensitive to the carbon concentration of the martensite and to the tempering temperature. It follows that the effect of stress on the development of the carbide microstructure (and the tendency to precipitate a single variant) will be most prominent at low carbon concentrations or at high tempering temperatures. We note that the stress need not be applied externally; it is equally valid to consider the influence of internal stresses due to transformation from austenite.

Lower bainite forms at higher temperatures than martensite so ΔG_{CHEM} for carbide precipitation is smaller; any partitioning of carbon into austenite reduces ΔG_{CHEM} further. Therefore, lower bainite plates are more likely to contain only a single carbide variant than martensite, as observed experimentally.

8.11 Summary

There is little doubt that the bainite reaction is influenced by externally applied stress, and by stresses generated internally due to transformation or heat-treatment. This interaction with stress appears to be related to the displacive mechanism of transformation with its invariant-plane strain shape deformation with its large shear component. Stress-assisted transformation can lead to anisotropic dimensional changes whose magnitudes and senses are impossible to explain on the assumption of a reconstructive transformation mechanism. Transformation plasticity is readily detected during the growth of bainite under the influence of stress, the magnitude of the observed effect being excess of that expected from volume change criteria alone.

There is evidence that the response of bainite to stress is similar to that of martensite. The bainite-start temperature is raised by the application of a tensile stress, lowered by hydrostatic compression, and there exists a B_d tempera-

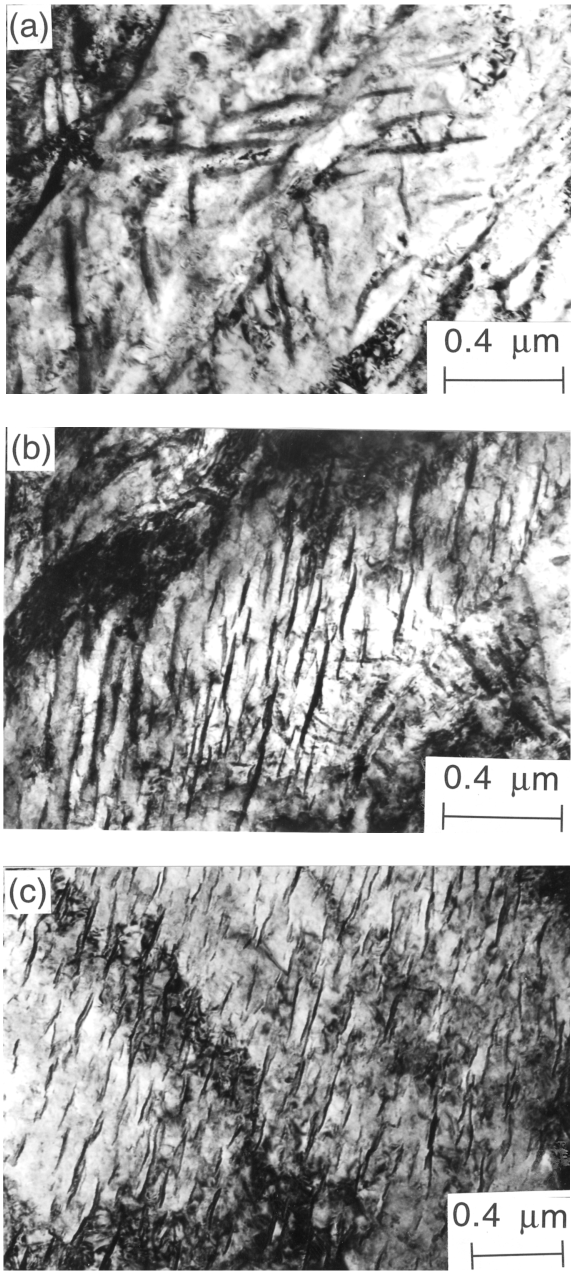

Fig. 8.16 The microstructure of martensite which is tempered under the influence of a uniaxial stress. The number of cementite variants in any given martensite plate decreases at the stress is increased: (a) zero stress; (b) 500 MPa; (c) 950 MPa. The stress is in all cases below the macroscopic yield strength of the sample at the tempering temperature. After Stewart *et al*., 1994.

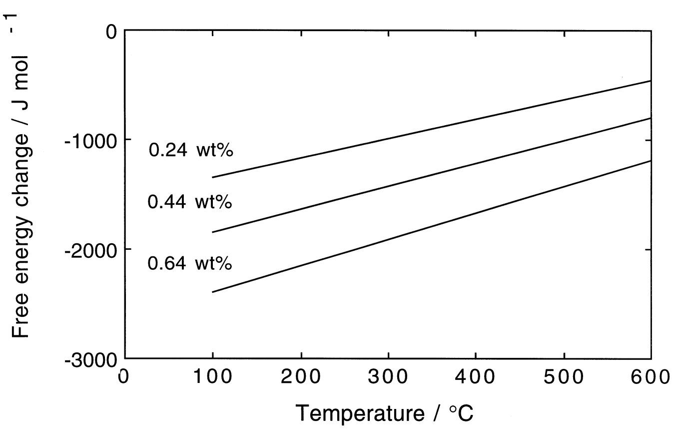

Fig. 8.17 Free energy change accompanying the precipitation of cementite from supersaturated ferrite, as a function of the carbon concentration and temperature (Stewart *et al.*, 1994).

ture beyond which the reaction will not occur whatever the magnitude of the stress. The microstructure of bainite responds to the applied stress, with clear evidence that the growth of certain crystallographic variants is favoured over others. The number of different sheaves per austenite grain decreases, causing the formation of large packets of parallel sheaves; this microstructure may be detrimental to toughness. Further work remains to be done in order to establish the criteria determining variant selection during stress-influenced transformation. Bainite also shows characteristics similar to those associated with the mechanical stabilisation of martensite, when the austenite is deformed prior to the growth of bainite.

Transformation to bainite accelerates under the influence of stress; whether this is predominantly due to enhanced nucleation or growth remains to be resolved. The extent to which the rate of reaction is accelerated increases with the rate of application of stress. On the other hand, heavy deformation of austenite prior to transformation causes mechanical stabilisation, another phenomenon associated uniquely with displacive transformations.

The primary effect on microstructure during transformation under stress is that of variant selection, which at low stresses reduces the number of sheaves per austenite grain. Variant selection does not lead to an obviously aligned microstructure until larger stresses are applied, in which case the sheaves probably form on habit plane variants which are most parallel to the planes of maximum shear stress. Although deviations from the random microstructures that form in the absence of applied stress are often not easily detectable, they reveal themselves unambiguously in the form of anisotropic dimensional changes during transformation.

The response of bainitic transformation to stress is therefore essentially similar to that of martensite, although there are some exceptions because of the higher transformation temperatures. Irreversible processes such as plastic deformation by lattice dislocations or the partitioning of carbon, are routine with bainite. This rules out the possibility of reversing the motion of the interface by reversing the stress, making phenomena like shape memory effects or rubber elasticity are extremely unlikely with bainite.

There is no doubt at all that the growth of bainite is stifled when the austenite is severely plastically deformed prior to transformation. The transformation can therefore be mechanically stabilised. This feature is impossible to explain except by a displacive transformation mechanism in which the interface motion is glissile.