Martensitic Transformation

Martensitic transformations are diffusionless. The change in crystal structure is achieved by a homogeneous deformation of the parent phase. To minimize the strain energy the martensite forms as thin plates on particular crystallographic planes known as the habit planes. The consequences of this mechanism can be seen macroscopically because the shape of the transformed region changes, the strain being a combination of shear (~ 0.25) parallel to, and a dilatational strain (~ 0.03) normal to the habit plane (Fig. 1). Given the nature of the homogeneous deformation, it should be possible to predict many of the aspects of the crystallography, morphology, transformation strains, and dynamic characteristics of martensite. We shall see that this is indeed the case, making martensitic transformation amenable to rigorous theoretical treatment.

Martensite can grow at temperatures close to absolute zero and at speeds in excess of $1000\,\mathrm{m\,s^{-1}}$. The transformation interface must then have a structure that is glissile, i.e., one that does not require diffusion. Only coherent or semicoherent interfaces can be glissile. A semicoherent interface will contain dislocations that periodically correct the misfit at the interface. For martensite, there can exist only one array of dislocations; multiple arrays can interfere, leading to the formation of jogs that require climb, rendering the interface sessile. This simple logic implies that for martensite the semicoherent interface must contain one line that remains undistorted and unrotated by the transformation strain; it is this *invariant line* that defines the line vector of the interfacial disloc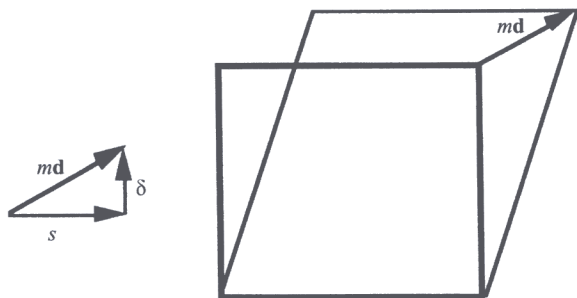ations. This single set can accommodate the misfit because there is no misfit parallel to the invariant line. The Burgers vector of the interface dislocations must in general lie out of the plane of the interface so that they can glide as the boundary advances.

Thus, martensitic transformations can occur only in systems where the parent and product lattices can be related by a transformation strain that leaves at least one line invariant.

The change from the f.c.c. (face-centered cubic austenite) to b.c.c. (body-centered cubic martensite) crystal structure in iron is sufficiently general to represent all martensitic transformations. Bain in 1924 proposed that the change in structure could be achieved by the simple homogeneous deformation illustrated in Fig. 2. The crystal structure of austenite can also be represented by a body-centered tetragonal (b.c.t.) unit cell, as shown in Fig. 2(b). It is then easy to see how the b.c.t. cell of austenite may be deformed to produce the required b.c.c. cell. Thus, the "Bain strain" consists of a contraction of the parent lattice along the a_3 axis and identical expansions along the a_1 and a_2 axes.

The Bain strain implies the following orientation relationship between the parent and product lattices:

$$[001]_{\mathrm{fcc}} \parallel [001]_{\mathrm{bcc}} \quad [1\bar{1}0]_{\mathrm{fcc}} \parallel [100]_{\mathrm{bcc}} \quad [110]_{\mathrm{fcc}} \parallel [010]_{\mathrm{bcc}}$$

but, in fact, the experimentally observed orientation

Figure 1
Schematic representation of the shape deformation accompanying martensitic transformation. The habit plane normal is vertical, δ and ε represent the dilatational and shear strains, respectively, and *m*d is the net displacement vector. Such a deformation is known as an invariant-plane strain because it leaves the habit plane undistorted and unrotated. The square has unit height.
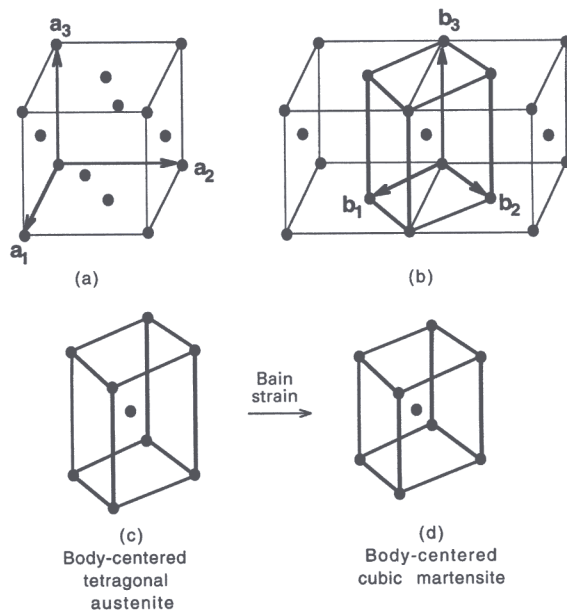

Figure 2
(a) Conventional f.c.c. unit cell. (b) Relationship between f.c.c. and b.c.t. cells of austenite (some of the lattice points have been omitted to avoid confusion). (c, d) Bain strain deforming the austenite lattice into a b.c.c. martensite lattice.
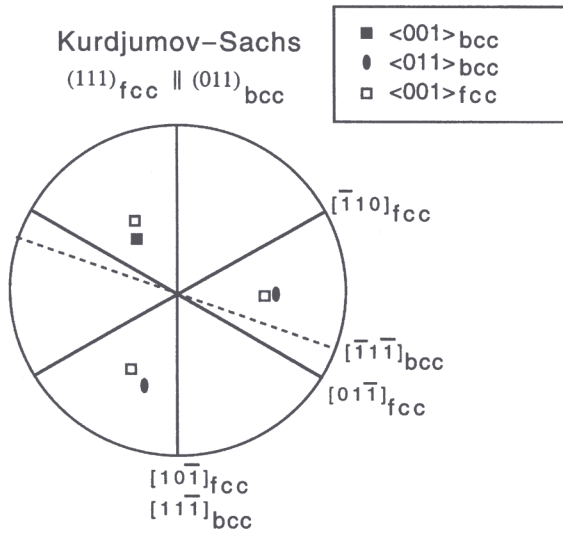
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Kurdjumov–Sachs

$(111)_{fcc} \parallel (011)_{bcc}$

■	$<001>_{bcc}$
●	$<011>_{bcc}$
□	$<001>_{fcc}$

Figure 3
Stereographic representation of the Kurdjumov–Sachs orientation relationship. The stereogram is centered on $(111)_{fcc} \parallel (011)_{bcc}$. The unmarked neighboring pairs of poles would superpose exactly for the Bain orientation but they do not do so for the Kurdjumov–Sachs orientation.

relationships are irrational, e.g., close to the Kurdjumov–Sachs orientation:

$$\{111\}_{fcc} \parallel \{011\}_{bcc} \quad \langle 10\bar{1}\rangle_{fcc} \parallel \langle 11\bar{1}\rangle_{bcc}$$

The difference between these two orientations is shown on the stereographic projection in Fig. 3.

Temporarily neglecting the fact that the Bain orientation is inconsistent with experiments, we proceed to examine whether the Bain strain leaves at least one line invariant. After all, this is a necessary condition for martensitic transformation.

The austenite is represented in Figs. 4(a) and (b) as a sphere, which, as a result of the Bain strain, **B**, is deformed into an ellipsoid of revolution that represents the martensite. There are no lines that are left undistorted or unrotated by **B**. There are no lines in the $(001)_{fcc}$ plane which are undistorted. The lines wx and yz are undistorted but are rotated to the new positions $w'x'$ and $y'z'$. Such rotated lines are not invariant. However, the combined effect of the Bain strain, **B**, and the rigid body rotation, **R**, is indeed an invariant-line strain (ILS) because it brings yz and $y'z'$ into coincidence (Fig. 4(c)). This is the reason why the observed irrational orientation relationship differs from that implied by the Bain strain. The rotation required to convert **B** into an ILS precisely corrects the Bain orientation into that which is observed experimentally.

As can be seen from Fig. 4(c), there is no rotation which can make **B** into an invariant-plane strain (IPS) since this would require two nonparallel invariant lines. Thus, for the f.c.c. → b.c.c. transformation, austenite cannot be transformed into martensite by a homogeneous strain which is an IPS. And yet the observed shape deformation leaves the habit plane undistorted and unrotated, i.e., it is an IPS (Fig. 1).

The phenomenological theory of martensite crystallography solves this remaining problem (Fig. 5). The Bain strain converts the structure of the parent phase into that of the product phase. When combined
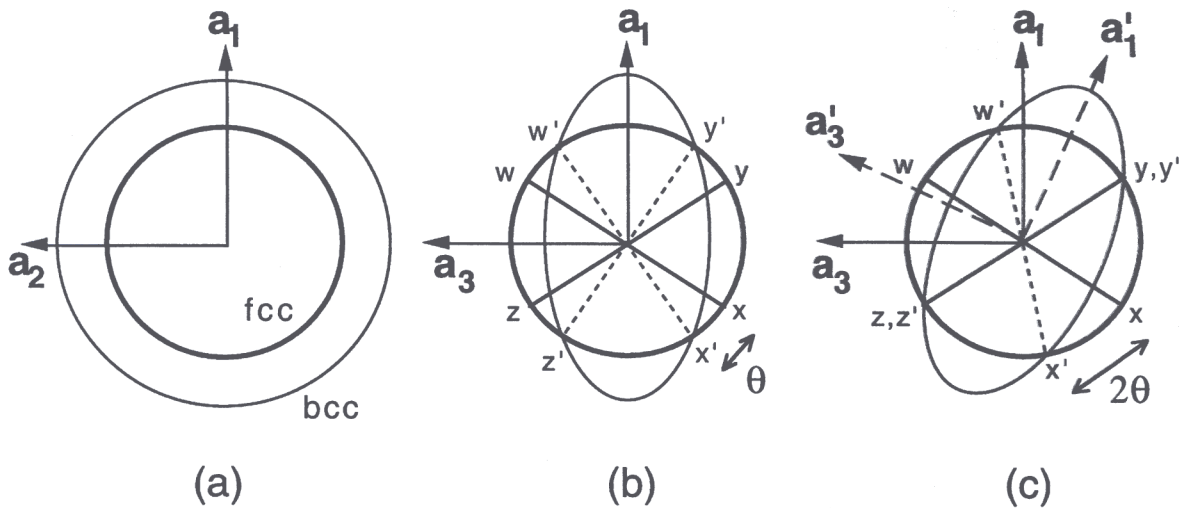

Figure 4
(a, b) Effect of the Bain strain on austenite, which when undeformed is represented as a sphere of diameter $wx = yz$ in three dimensions. The strain transforms it to an ellipsoid of revolution. (c) Invariant-line strain obtained by combining the Bain strain with a rigid body rotation through an angle θ.
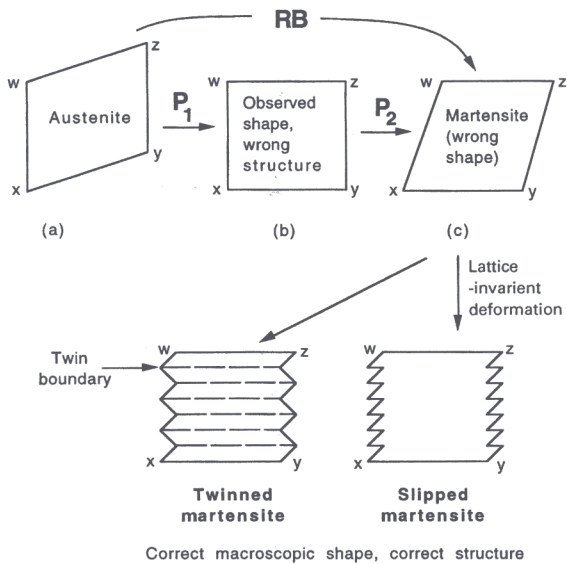
Figure 5
Phenomenological theory of martensite crystallography.

with an appropriate rigid body rotation, the net homogeneous lattice deformation, **RB**, is an ILS (step (a) to (c) in Fig. 5). However, the observed shape deformation is an IPS, **P**₁ (step (a) to (b) in Fig. 5), but this gives the wrong crystal structure. If a second homogeneous shear, **P**₂, is combined with **P**₁ (step (b) to (c)), then the correct structure is obtained but the wrong shape since

$$\mathbf{P_1 P_2 = RB}$$

These discrepancies are all resolved if the shape-changing effect of **P**₂ is cancelled macroscopically by an inhomogeneous lattice-invariant deformation, which may be slip or twinning, as illustrated in Fig. 5.

The theory explains all the observed features of the martensite crystallography. The orientation relationship is predicted by deducing the rotation needed to change the Bain strain into an ILS. The habit plane does not have rational indices because the amount of lattice-invariant deformation needed to recover the correct macroscopic shape is not usually rational. The theory predicts a substructure in plates of martensite (either twins or slip steps) as is observed experimentally. The transformation goes to all the trouble of ensuring that the shape deformation is macro-

scopically an IPS because this reduces the strain energy when compared with the case where the shape deformation might be an ILS.

The mechanism by which martensite nucleates must also be consistent with diffusionless transformation and the fact that the phase can form at remarkably low temperatures and high strain rates. The classical idea that nucleation happens when random phase and composition fluctuations reach a critical size is not reasonable in such circumstances. The probable mechanism involves the dissociation of three-dimensional arrays of dislocations. The faulted structure between the partials then represents the embryo, which is said to become the nucleus of martensite when the circumstances are right for the rapid growth of the embryo. In most cases this means that the chemical driving force for transformation must be large enough to allow the partials to propagate with a rate limited only by the usual barriers to dislocation motion. This theory is particularly useful because it predicts correctly that the activation energy for nucleation is directly proportional to the driving force for transformation. This behavior is not expected in classical nucleation by heterophase fluctuations.

There have been many proposals that nucleation involves some sort of lattice instability or a strain spinodal. These do not seem to be useful models except when the driving force for transformation is extremely large.

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