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Dislocations with ⟨100⟩ Burgers vectors in Ni_3Al

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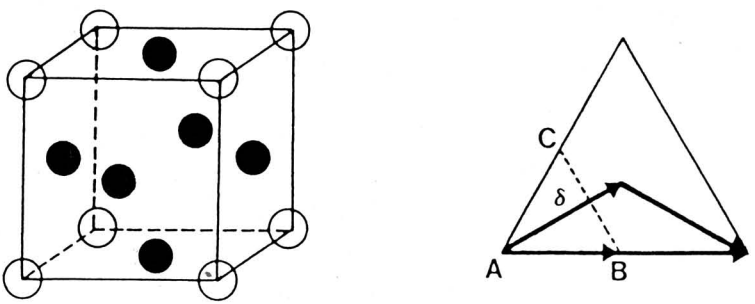
ABSTRACT

The occurrence of dislocations with ⟨100⟩ Burgers vectors, the shortest unit translation in the Ll_2 structure, is reported. The reasons which may explain why ⟨110⟩ Burgers vectors predominate during deformation are discussed on the basis of weak-beam observations. High Peierls forces on ⟨100⟩ dislocations, rather than anisotropy, are probably responsible for their rarity.

Although ⟨100⟩ is the shortest unit translation in the Ll_2 structure (fig. 1), the activation of sources of dislocations with perfect ⟨110⟩ Burgers vectors has been reported to result from deformation at any temperature. The dislocations may further dissociate into either two $\frac{1}{2}\langle 110\rangle$ or two $\frac{1}{3}\langle 112\rangle$ partials giving rise to an antiphase boundary (APB) or a superlattice intrinsic stacking fault (SISF) respectively. Except in Zr_3Al (Howe, Rainville and Schulson 1974) and possibly in Pt_3Al (Paidar, Pope and Yamaguchi 1981), the former dissociation mode prevails. This dissociation fails to explain the absence of a dislocation with ⟨100⟩ Burgers vector since, to a first-order approximation, its self-energy coincides with the minimum value of the total energy of a ⟨110⟩ dissociated dislocation (i.e. when the APB energy is zero and when the $\frac{1}{2}\langle 110\rangle$ partials are separated by a distance of the order of the size of the Frank network).

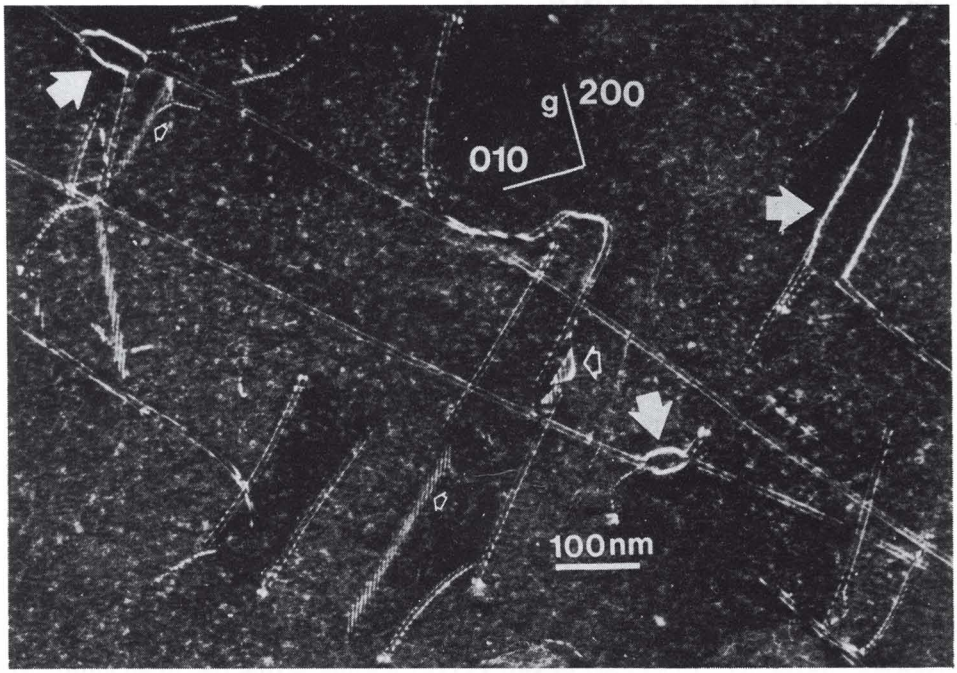
In this Letter, we report on the observation of minority dislocation segments with ⟨100⟩ Burgers vectors in polycrystalline Ni_3Al samples deformed at 650°C. The experimental procedure has been described in a previous paper (Veyssière, Guan and Rabier 1984). Each ⟨100⟩ segment appears at the intersection of two dislocations with orthogonal ⟨110⟩ Burgers vectors. It is worth mentioning that, whereas ⟨110⟩

Fig. 1



The unit cell of Ni_3Al (Ll_2) and the basic dissociation schemes for a ⟨110⟩ perfect dislocation in the Ll_2 structure. (Open circles represent aluminium atoms.)

Fig. 2



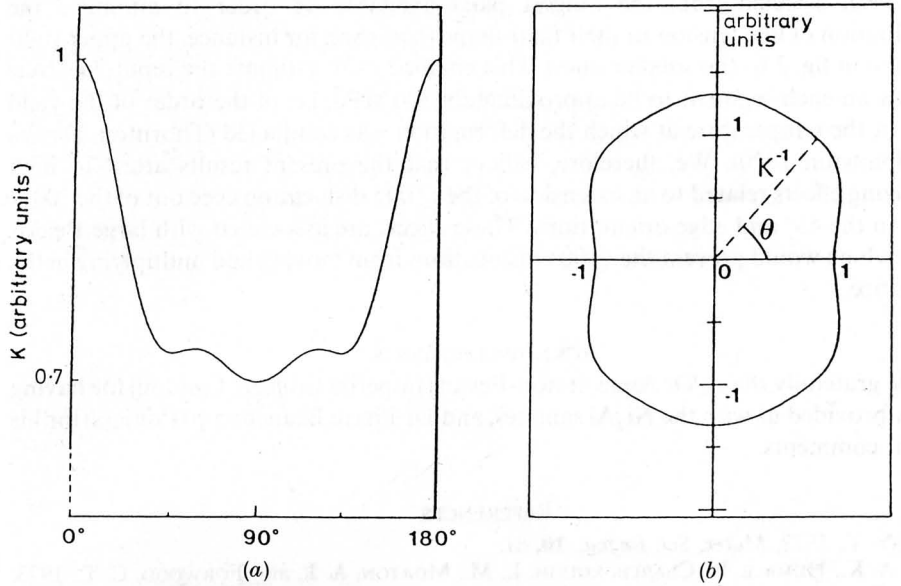
Network of dislocations after deformation at 650°C. Both families A ($\mathbf{b}=[110]$ and B ($\mathbf{b}=[1\bar{1}0]$) are dissociated. The partials are all imaged under $\mathbf{g}\cdot\mathbf{b}=1$ conditions. The junctions which are indicated by solid arrows have a [100] Burgers vector; their contrast is reinforced since $\mathbf{g}\cdot\mathbf{b}=2$ for these.

dislocation may extend over several tens of micrometres, these with $\langle 100\rangle$ Burgers vectors are much shorter, never exceeding 500 nm. In other words, despite being expected to exhibit the smallest self-energy in Ll_2 alloys, they can be formed but seem neither to multiply nor to propagate.

The dislocation microstructure shown in fig. 2 is a typical example of the deformation microstructure. It is composed of two families of dislocations, (A) lying in and (B) inclined to the foil plane with [110] and $[1\bar{1}0]$ Burgers vectors respectively. Their nearly screw segments are dissociated in the (001) plane whereas the segments with dominant edge character are dissociated by a mixture of glide and climb in a plane lying between $(3\bar{1}0)$ and $(2\bar{1}0)$, in excellent agreement with previous determinations (Veyssière at al. 1984, Veyssière 1984). SISFs are also present in the foil (open arrows), their reduced dimensions being consistent with the mechanism of formation suggested by Veyssière, Douin and Beauchamp (1985). Three pairs of dislocation segments with the same [100] Burgers vector are visible in fig. 2. Their formation results from the intersection without recombination of the $\langle 110\rangle$ dislocations according to the reaction $\frac{1}{2}[110]+\frac{1}{2}[1\bar{1}0]\rightarrow[100]$.

Since $\langle 100\rangle$ is a unit translation in the Ll_2 lattice, there need not be a dislocation segment at the edges of the two APBs with orthogonal displacement vectors. Accordingly, the equilibrium shape of the [100] junctions is controlled by their repulsive interaction only, since they are not linked by a faulted defect. As far as the geometry of these junctions is concerned, weak-beam stereomicroscopy indicates that the [100] segments are contained in the (001) plane. Each dislocation appears to align

Fig. 3



Variation of the dislocation energy factor K with respect to the angle θ between its line and its $\langle 100 \rangle$ Burgers vectors calculated using anisotropic elasticity. (*a*) Direct representation; (*b*) $K^{-1}(\theta)$ in polar coordinates.

to preferred crystallographic orientations which are either 45° mixed or, to a lesser extent, pure edge in character. We have failed to detect any dissociation of these segments under weak-beam conditions where the dislocation image width was approximately 1 nm.

The reasons for the anomalous presence of $\langle 100 \rangle$ rather than $\langle 110 \rangle$ Burgers vectors are not understood. Their existence could be influenced by anisotropy ($A = 3{\cdot}3$ in Ni_3Al, Ono and Stern 1979), which, for example, has been shown to determine the Burgers vectors of junctions in b.c.c. crystals (Chou 1972). Thus, the extent to which anisotropy influences the shape of $\langle 100 \rangle$ dislocations in Ni_3Al has been computed in the case of a straight and infinite dislocation with a variable character. We have run the relevant subroutines made available by Head, Humble, Clareborough, Morton and Forwood (1973), using the elastic constants measured between 77 and 650 K by Ono and Stern (1979). The results of the computations are presented in fig. 3. They are not affected significantly by temperature. They indicate that the edge dislocation is the most stable, with a secondary minimum occurring when the dislocation line makes an angle of 45° with the Burgers vector (fig. 3 (*a*)). The calculations agree with the present observations, especially regarding the absence of screw dislocations. The region of instability of the $\langle 100 \rangle$ dislocation is better illustrated in fig. 3 (*b*) where $K^{-1}(\theta)$ has been plotted radially (K is the dislocation energy factor) (Steeds, 1973): dislocation lines oriented within 25° of screw orientation correspond to the concave part of the curve, i.e. to regions where instability occurs. This enables us to explain the absence of screw $\langle 100 \rangle$ segments. However, the agreement is only partial since another concave region should be observed in order to justify some alignment of the dislocations along 45° and edge orientations; in other words, the relative variation of K is much too small ($\simeq 3\%$) in this angular range (fig. 3 (*a*)). However, the average width of the $\langle 100 \rangle$ dislocation

pairs has been estimated from significant examples taken throughout the foil. These have been selected from the longest pairs available, in order to minimize the contribution of line tension to their final shape (compare, for instance, the upper right junction in fig. 2 to the smaller ones). This enabled us to estimate the repulsive stress applied on each segment to be approximately 300 MPa, i.e. of the order of the yield stress at the temperature at which the deformation was conducted (Thornton, Davies and Johnston 1970). We, therefore, believe that the present results attest to high stabilizing effects related to an extension of the $\langle 100 \rangle$ dislocation core out of the $\{001\}$ plane in the 45° and edge orientations. These effects are associated with large Peierls forces which would prevent the $\langle 100 \rangle$ dislocations from moving and multiplying in the Ll_2 lattice.

Acknowledgments

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