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Deformation twinning in alloys of low stacking-fault energy

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A High Voltage Electron Microscopy (HVEM) examination of quenched and strained alloys of low stacking-fault energy has indicated a high incidence of deformation twins at second phase particles. Deformation twins are produced at NbC particles during quenching of 20/25/Nb steel and following approximately 1% strain in the case of Ni_3Al particles in PE-16 alloy.

The association of twins and particles in these alloys is a direct result of the development of large stresses at the particle-matrix interface, due to differential contraction rates during quenching. At NbC particles of the order 10^3 nm diameter, the twinning stress is exceeded on quenching from the solution treatment temperature. In the case of PE-16 alloy the thermal stress set up at the Ni_3Al particle is smaller, necessitating a further external applied stress in order to nucleate such twins.

1. Introduction

Deformation twinning in fcc materials is normally observed at low temperatures following fast strain rates. Cook and Parsons [1] investigated the effect of chromium additions to a Fe-25Ni-C, alloy, which was known to twin [2]. Although the chromium addition raised the stacking fault energy (SFE, γ) of the basic alloy to approximately 25 ergs cm^{-2}, twins were always observed at strains greater than 0.10. Using the equations of Venables [3] these workers estimated the twinning stress to be 17.5 kg mm^{-2} for the 20/25/Nb stabilized steel investigated. Venables [3] has collected together the available twinning data for copper based alloys shown in Fig. 1. As suggested by Cook and Parsons [1], these data should be comparable with those of Fe:Ni:Cr alloys due to their similarities in lattice parameter and dislocation reactions.

In the present investigation, tensile tests at high (700° C) and low (25° C) temperatures have revealed load drops in 0.01 % C, 20/25 Nb steels. From previous work [4] the SFE of 20/25/Nb is known to be in the region 25 ergs cm^{-2}. These two facts suggest deformation twinning as an alternative to dislocation slip in 20/25/Nb steel under certain conditions.

It was therefore decided to investigate in detail the morphology of deformation twinning in 20/25/Nb steel and PE-16 alloys. High Voltage Electron Microscopy (HVEM) has made possible the examination of thick foils, with the consequent increased incidence of twin observation.

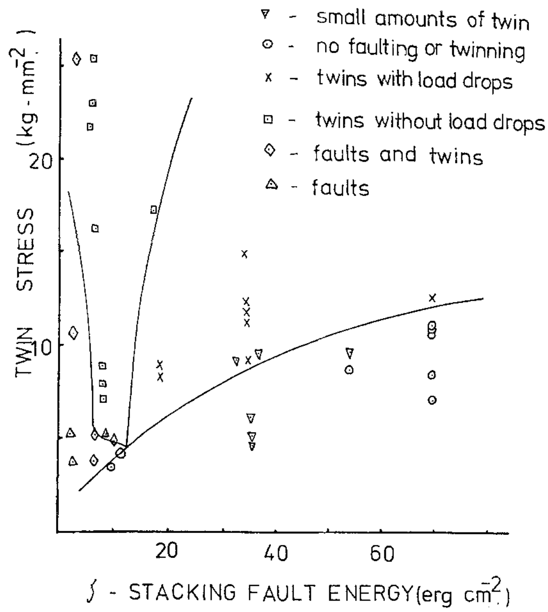

Figure 1 Twinning characteristics of copper based alloys.

2. Experimental

2.1. 20/25/Nb stabilised steel

This alloy was solution-treated at 1030° C for 1 h

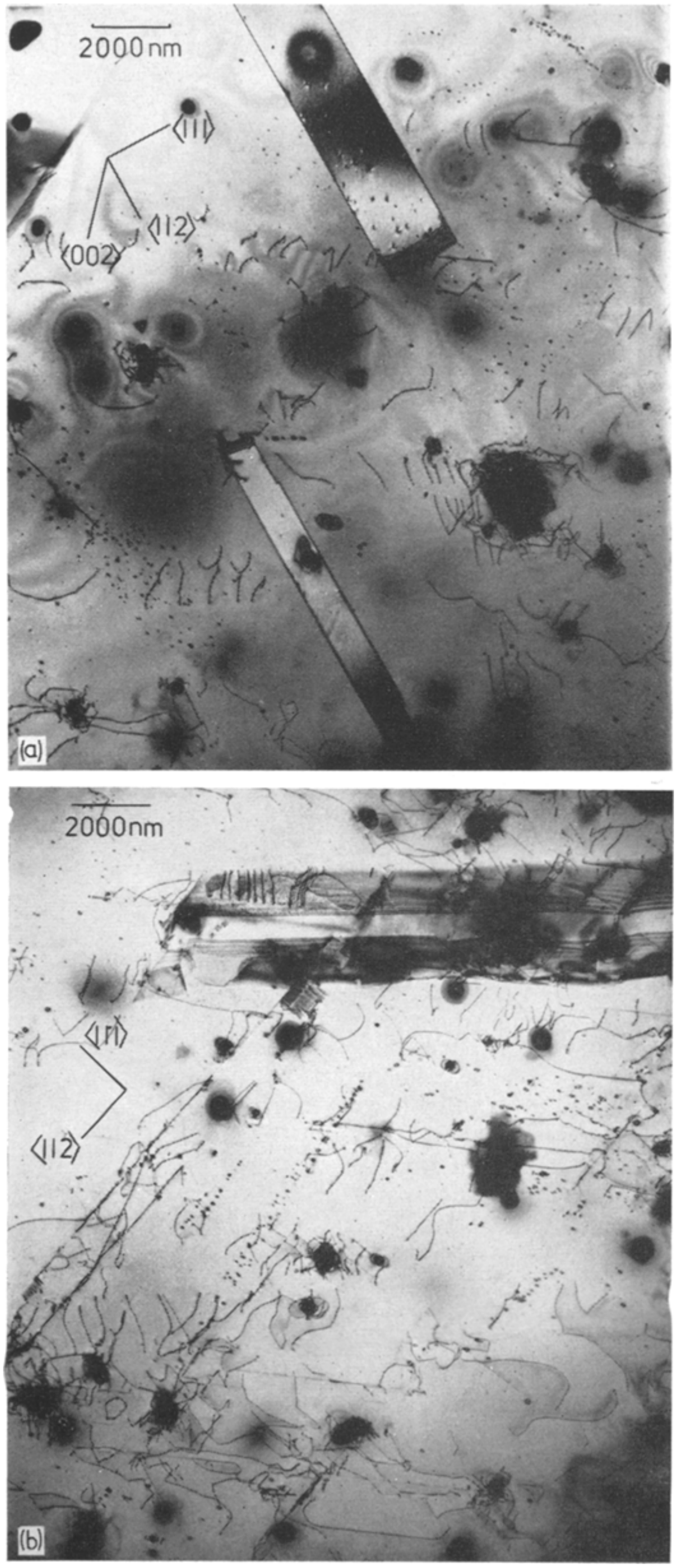

Figure 2 (a) Deformation twins and NbC particles. (b) Same as 2a.

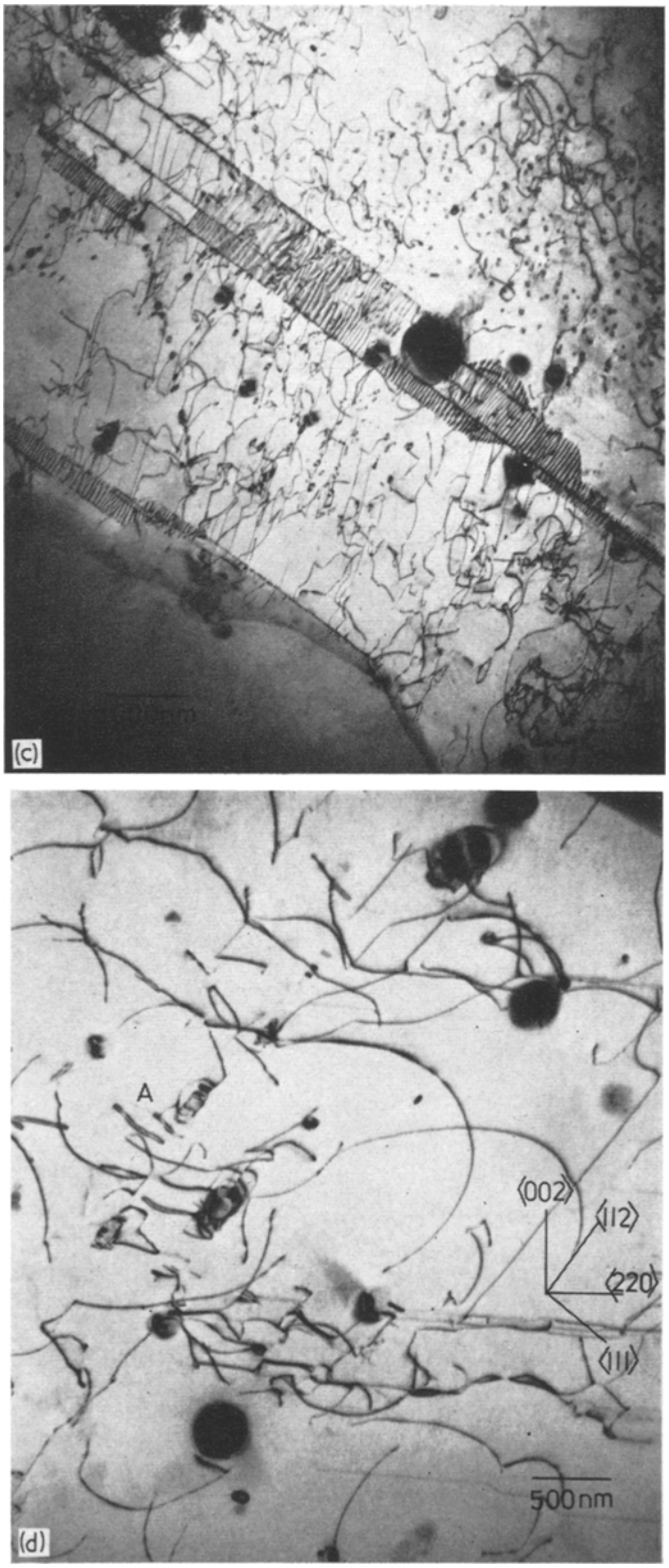

Figure 2 (c) Deformation twin at NbC particle. (d) Dislocation loop formation at NbC following quenching.

followed by water-quenching.

2.2. PE-16 alloy

Solution-treated at 1080°C for 1 h followed by water-quenching. Subsequently strained 4% in tension at room temperature.

Analyses and foil preparation of these alloys has been published elsewhere [5, 6].

3. Results

Typical micrographs enclosing several different grains are shown in Fig. 2a, b, c and d. Several twins and dislocation loops are shown associated with undissolved NbC particles. A selected area diffraction pattern enclosing a twin-matrix boundary is shown and indexed in Fig. 3. In the majority of cases, the examined grains show a ⟨110⟩ orientation with respect to the electron beam. All twin lamallae are therefore examined with their coherent boundaries at right angles to the foil surface.

An example of twin formation in PE-16 is shown in Fig. 4 together with the selected area diffraction pattern.

At certain electron beam orientations, many of the incoherent twin boundaries show uncharacteristic dislocation contrast. This is illustrated in Fig. 5a and b. The dislocation contrast changes abruptly on moving from matrix to twin, i.e. at point A, Fig. 5. In addition several of the dislocations show a periodic contrast at the dislocation core, i.e. along dislocation *X-Y*, Fig. 5b.

4. Discussion

The observed association between deformation twins and particles can be accounted for as follows.

The difference in coefficient of thermal expansion of NbC particle and matrix over a temperature range of approximately 0 to 1000°C is sufficient to set up stresses greater than the local yield stress at the particle-matrix interface with subsequent formation of dislocation loops. Taking the mean coefficient of thermal expansion of NbC as $6.85 \times 10^{-6}/°C$ and 20/25/Nb as $10.0 \times 10^{-6}/°C$ over the temperature range 0 to 1000°C, the differential contraction (ΔL) between lattice and particle is

$$\Delta L = h(\alpha_m - \alpha_p)T \qquad (1)$$

where h = mean particle diameter; α_m = mean coefficient of thermal expansion of matrix; α_p = mean coefficient of thermal expansion of particle and T = temperature change.

For a temperature change 1000°C, ΔL is

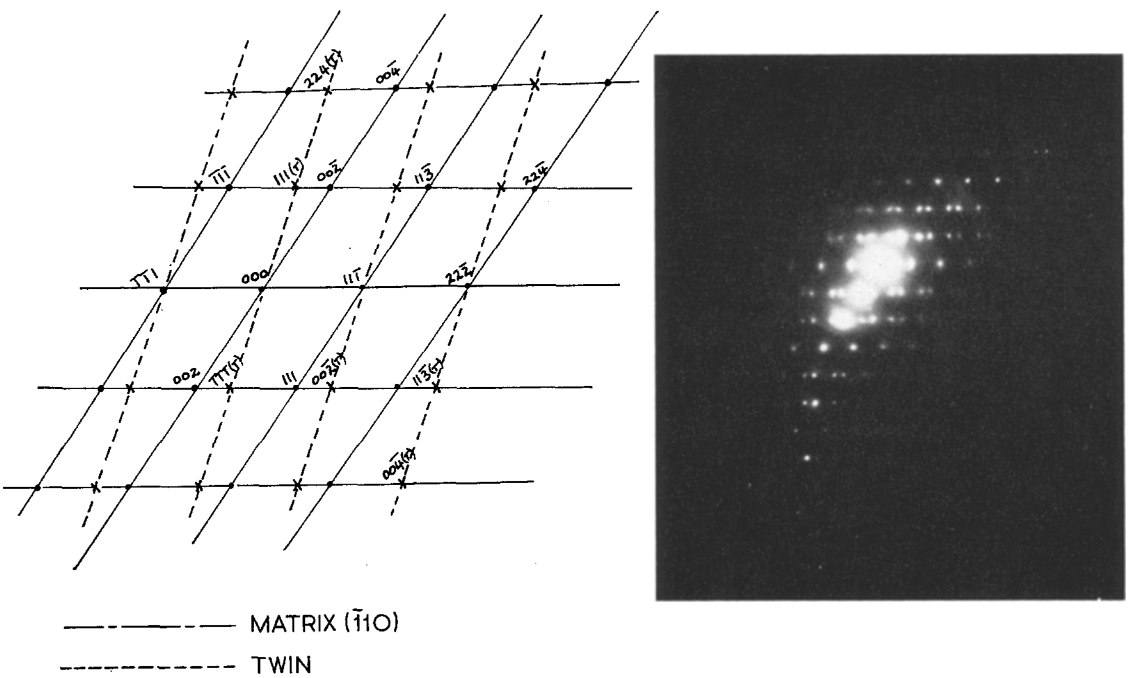

Figure 3 Typical selected area diffraction pattern from twin-matrix in a ⟨110⟩ type foil. Indexing diagram for a (110) pattern with a ⟨111⟩ twin axis.

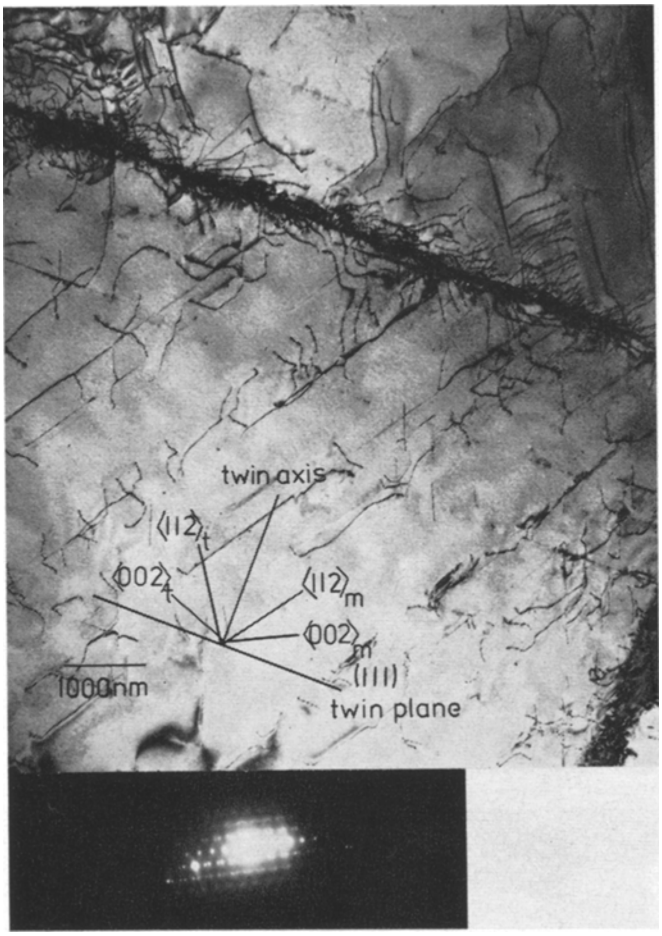

Figure 4 Deformation twin boundary and ⟨110⟩ selected area diffraction pattern in PE-16 alloy.

approximately 32Å for a 10^3 nm particle. The number of loops (n) punched out from such a particle will be of the order

$$n = \frac{\Delta L}{\mathbf{b}} \tag{2}$$

$\mathbf{b}$ = Burgers vector loop, giving $n \simeq$ 12-14 for a $\frac{1}{2}[110]$ loop.

Examination of many hundreds of particles has failed to show such a number of punched out loops. Quite often two or three loops are found (i.e. Fig. 2d, point A), but in several instances micro-twins are associated with particles.

The mechanism of formation of deformation twins from $\frac{1}{2}[110]$ loops has been suggested by Venables [7]. The twin source is shown illustrated in Fig. 6. Following quenching, the high stresses set up in the neighbourhood of the particle punch out loops and activate lattice dislocations near by. In the situation envisaged by Venables [7] shown in Fig. 6, twin formation occurs following the formation of one loop, providing the stored energy remaining is sufficient to activate twinning. The elastic strain (ϵ) stored in the region of the particle is approximately $\epsilon = 3 \times 10^{-3}$ which means a stress of the order ϵE (E = Youngs Modulus) i.e. approximately 30 kg mm^{-2}. The critical resolved shear stress in the twinning direction τ_T has been given in terms of the stacking fault energy γ and shear modulus μ [7].

$$n\,\tau_T = \frac{\gamma}{\mathbf{b}} + \frac{\mu\mathbf{b}}{2a_0} \tag{3}$$

$\mathbf{b}$ = Burgers vector, a_0 = radius twinning loop.

For the case of the present model involving twinning at particles we have $2a_0 \simeq 10^{-4}$ cm; $\mu = 10^{12}$ dyn cm^{-2}; $\mathbf{b} = a/2\ [110] \simeq 2.5$Å and

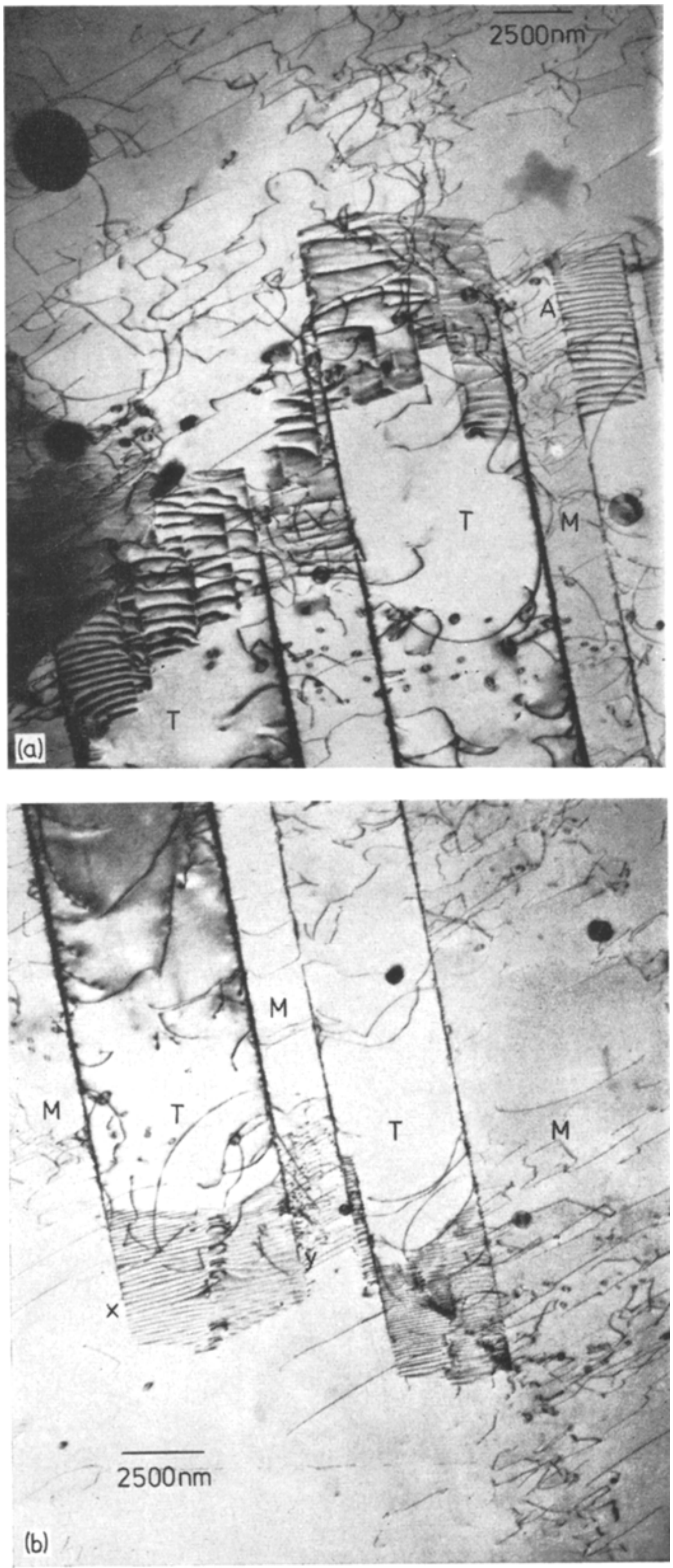

Figure 5 (a) Typical dislocation contrast at incoherent twin boundaries in 20/25/Nb alloy. (b) Same as 5a. Opposite end of twins shown in 5a.

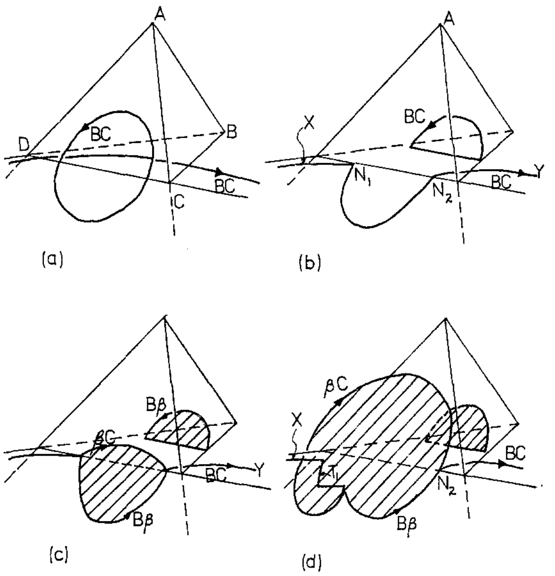

Figure 6 The twin source in neutron-irradiation metals.
(a) A dislocation of the primary slip system BC on plane *a*, held up by a prismatic loop BC on plane *b*.
(b) The slip dislocation intersects the loop by moving to the left and displaces the two halves of the loop by a Burgers vector BC. The slip dislocation is now continuous with the open half of the loop.
(c) When the stress on *b* is high enough, the dissociation BC $\rightarrow$ Bβ + βC occurs.
The twinning dislocation βC spirals round the pole *X* and creates a jog at N_1. The new twin source T_1 is formed by movements of this jog around the loop, and it can dissociate again producing a second layer of twin. The pole *Y* may be inoperative because of the presence of the sessile loop Bβ. The growth of the twin is analogous to that described in Fig. 6.

$\gamma = 25$ ergs cm^{-2}, giving $n \tau_T \simeq 18$ kgm mm^{-2}.

Clearly, the elastic stored energy at the particle is sufficient to nucleate the deformation twin following formation of one prismatic loop.

The rate of twin growth is exceedingly complex, but Venables [3] has estimated this for a series of low stacking-fault energy copper alloys. Assuming a similar argument for the 20/25/Nb alloys, then the speed of twin propagation for the twin morphology shown in Figs. 2 to 5 is approximately 0.6 × speed of sound. The alternative mode of stored energy release near the particle, involves nucleation and movement of a train of prismatic loops. Such a process will involve diffusion mechanisms which are inevitably slower than twin propagation.

As indicated in Equation 3, twin formation can only occur at NbC particles of the order 10^3 nm diameter following a temperature transient of 1000°C. At the high accelerating voltages employed in the present experiment, many such particles are entirely enclosed within the metal foil, enabling the observation of particle-twin reactions. These reactions are not observed at lower voltages employing thinner foils.

In many cases, twin propagation develops large stresses near the growing twin tip. These stresses are relieved by the formation of emissiary dislocations as shown in Fig. 5a and b. Several of these emissiary dislocations are dissociated at some points along their length (i.e. along dislocation *X*-*Y*, Fig. 5b). Such emissiary dislocations may therefore form at twin-matrix steps or by the accumulation of several twin-matrix boundary dislocations during twin growth [8, 9].

Formation of deformation twins can lead to the modification of mechanical properties. Twins can act as sites for recrystallization, precipitation and fracture [1].

5. Conclusions

1. Deformation twins have been found associated with NbC and Ni_3Al particles in 20/25/Nb and PE-16 alloys.
2. Stresses set up at NbC particles during thermal quenching are in excess of those required to nucleate twins.
3. Deformation twinning can rapidly relieve such high stresses at NbC and Ni_3Al particles.
4. The developed twins may then modify the mechanical properties due to inhomogeneity of structure.

Acknowledgement

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