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The effect of yttrium and erbium dispersoids on the deformation behaviour of titanium

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The deformation characteristics at 295 K and 575 K of polycrystalline Ti–Y and Ti–Er alloys containing 20 to 90 nm diameter dispersions were investigated by stress–strain measurements and by transmission electron microscopic observations of deformation substructures. The presence of the dispersoids increases the yield stress at both 295 K and 575 K, with the dispersion strengthening being more pronounced for the larger grain size and at the higher temperature. In dispersoid-free titanium the yield stress varies with grain size at both 295 K and 575 K in accordance with the Hall–Petch relation, but the yield stress of the Ti–Er alloys does not show a well-defined linear dependence on the inverse square root of grain size. The work hardening is less sensitive to grain size in the Ti–Er alloys than in pure titanium. The extent of twinning is significantly higher in the Ti–Er and Ti–Y alloys than in titanium at both temperatures. The influence of the dispersoids on deformation substructure and grain size related deformation behaviour is discussed.

1. Introduction

Small additions of yttrium and erbium to titanium result in a uniform dispersion of 20 to 90 nm diameter particles in the titanium matrix. The presence of such fine dispersoids significantly refines the microstructure and retards grain growth at elevated temperatures [1, 2]. In the present investigation, the effect of Er and Y dispersoids on the deformation behaviour of polycrystalline titanium was studied to determine the principal mechanisms by which the Er and Y additions influence the deformation characteristics. The dependence of the flow stress and the work hardening index on grain size, dispersed phase parameters, and deformation substructure was determined by tensile–property measurements and by transmission electron microscopic observation of deformation substructures.

2. Experimental

Electrolytic Ti (EL-60 Ti), Ti–0.3 Y, Ti–0.05 Er, and Ti–0.1 Er alloys were prepared by the non-consumable electrode vacuum arc-melting method and hot-rolled to a 3.2-mm final thickness. The rolled sheets were subjected to various recrystallization–annealing treatments from 775 to 1025 K to produce grain sizes from 9 to 90 μm in diameter. Strip tensile specimens were machined from the annealed sheets and deformed in tension at 295 K and 575 K to various strains; the high-temperature deformation was performed in a purified, dry argon atmosphere. The development of dislocation substructures in the deformed specimens was studied by transmission electron microscopic examination of thin foils prepared from specimens deformed to various strains.

3. Results and discussion

The dispersoids in the Ti–0.1 Er alloy were incoherent with an average particle diameter of 70 nm. In the case of the Ti–0.3 Y alloy, the dispersion consisted of coherent particles ($<$ 10 nm diameter) and incoherent particles with an average diameter of 75 nm. The effect of Er dispersoids on the stress–strain characteristics of polycrystalline titanium of two different grain sizes at room

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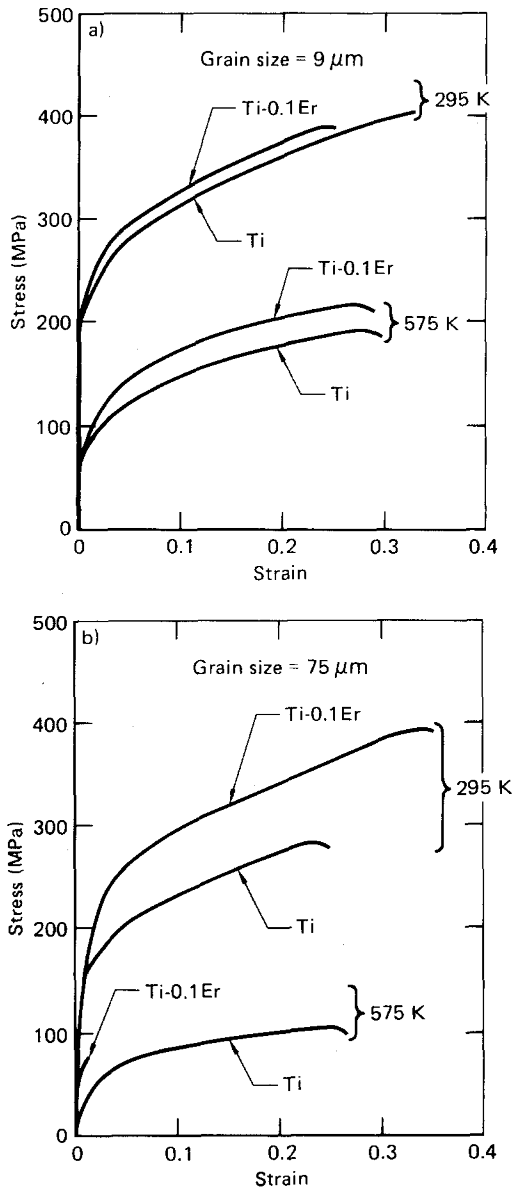

Figure 1 The effect of erbium on the stress–strain behaviour of titanium at 295 K and 575 K; (a) 9 μm grain size and (b) 75 μm grain size.

temperature and at 575 K is shown in Figs. 1a and b. The presence of the dispersoids increases the yield stress at both 295 K and 575 K, with the dispersion strengthening effect being more pronounced for the larger grain size and at the higher temperature. The grain size dependence of yield stress in Ti and Ti–Er at the two test temperatures is shown in Fig. 2. Pure Ti exhibits the Hall–Petch linear relationship between yield stress and the inverse square root of grain size at both temperatures, but the Ti–Er alloys do not show a well-defined dependence. The strengthening effect of

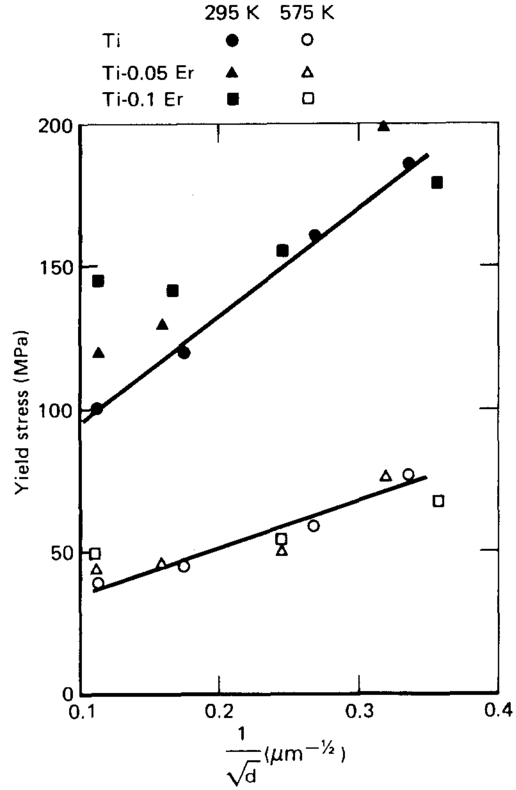

Figure 2 Grain size dependence of yield stress in Ti and Ti–Er at 295 K and 575 K.

the Er dispersoids decreases as the grain size decreases because of the increasing dominance of grain-boundary strengthening. The linear relationship between the yield stress at 0.2% offset and the inverse square root of grain size in El60-Ti suggests that the initial plastic flow occurs by a simple Hall–Petch mechanism in which dislocations pile up against grain boundaries and create stress concentrations that nucleate dislocation sources in neighbouring grains. In the Ti–Er alloys, however, the plastic deformation occurs by a combination of particle-bypassing and slip-transmittal across grain boundaries.

The rare earth dispersoids modify the slip behaviour by their influence on dislocation sources as well as by acting as barriers to dislocation motion. Dislocations overcome the dispersed-phase particles by the Orowan bypass [3] and Hirsch cross-slip [4] mechanisms, resulting in greater densities of localized dislocation channels during the initial stages of deformation. The electron micrograph in Fig. 3 shows the dislocation–dispersoid interaction resulting in the formation of Orowan loops and prismatic loops. The debris formed by the localized slip in the vicinity of the

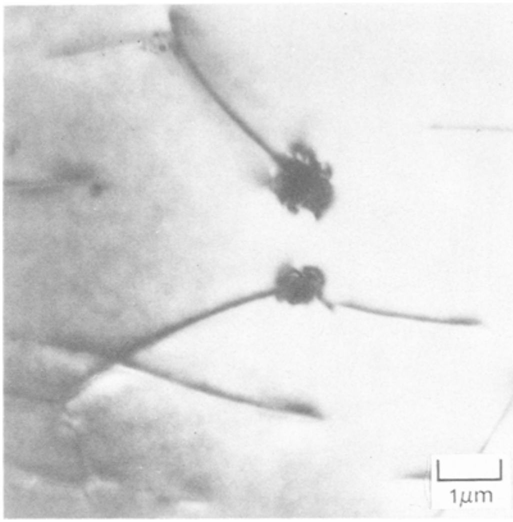

Figure 3 Electron micrograph showing the interaction between dislocations and dispersoids resulting in the formation of Orowan loops and prismatic loops.

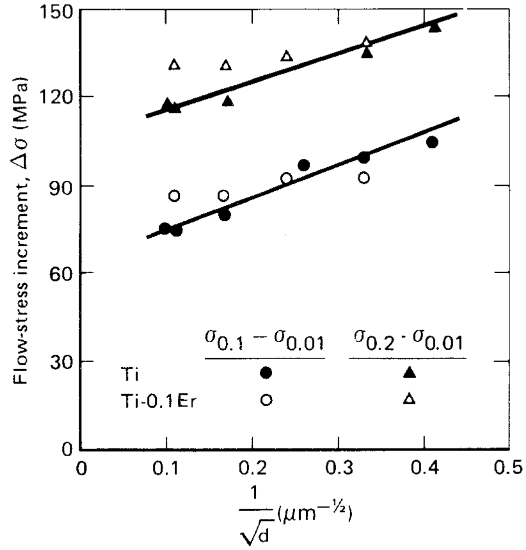

Figure 4 Grain size dependence of work hardening in Ti and Ti–0.1 Er.

dispersoids restricts the movement of dislocations, and thus a greater density of localized channels of dislocations is formed during the initial states of deformation. the mean free path of the travel of the dislocations is considerably smaller than the grain size in the presence of dispersoids, and thus the Ti–Er alloys do not show a well-defined Hall–Petch relationship. A calculation of the particle by-pass stress from the modified Orowan equation [5], yields a value much higher than the experimentally observed strength increment, even in the largest grain size alloy. This discrepancy is probably due to the scavenging of oxygen by Er, which reduces the matrix flow-stress contribution. Thus, the dispersion strengthening is counterbalanced to some extent by the softening caused by a reduction in the interstitial oxygen concentration, and the theoretically predicted dispersion strengthening is not observed experimentally.

The grain size dependence of work hardening in Ti and Ti–0.1 Er is shown in Fig. 4, in which the flow-stress increments between 10% and 1% strain, and between 20% and 1% strain, are plotted as functions of the inverse square root of grain size at room temperature. The work hardening is less sensitive to grain size in the Ti–Er alloy than in pure titanium, which again demonstrates the significant role of the particles in modifying the grain size related deformation behaviour.

The differences in the grain size dependence of work hardening in Ti and Ti–Er alloys can be explained on the basis of the observed dislocation structures. In Ti–Er, the presence of dispersoids leads to the formation of a well-defined dislocation braid structure early in the deformation process. The braid spacing is determined by the spacing of the dispersoids, and as the deformation increases, the slip length and consequently the flow stress are increasingly influenced by the dislocation braid spacing rather than the grain size. In pure titanium, however, a well-defined dislocation braid structure is formed only at large strains, and the slip length varies with strain and grain size to high strain [6].

Typical dislocation cell structures observed after large amounts of deformation in Ti and Ti–0.1 Er are shown in Figs. 5 and 6. The dislocation structure in Ti–0.1 Er alloy deformed to 40% strain at 295 K (Figs. 5a and b) consists of well-defined cells with the cell walls aligned parallel to $\langle\bar{1}2\bar{1}0\rangle$ directions. The average spacing of the cells in the $\langle\bar{1}2\bar{1}0\rangle$ directions is $\approx 0.5\,\mu m$; dispersoids are located both at the cell boundaries and within the cells. Dispersoid-free titanium that is similarly strained develops a coarser cell structure with an average cell spacing $\approx 1\,\mu m$. Furthermore, misorientations across cell walls are smaller in Ti–0.1 Er than in dispersoid-free titanium. At the higher temperature, a decreased planarity of slip, accompanied by recovery, results in larger dislocation cells (Figs. 6a and b).

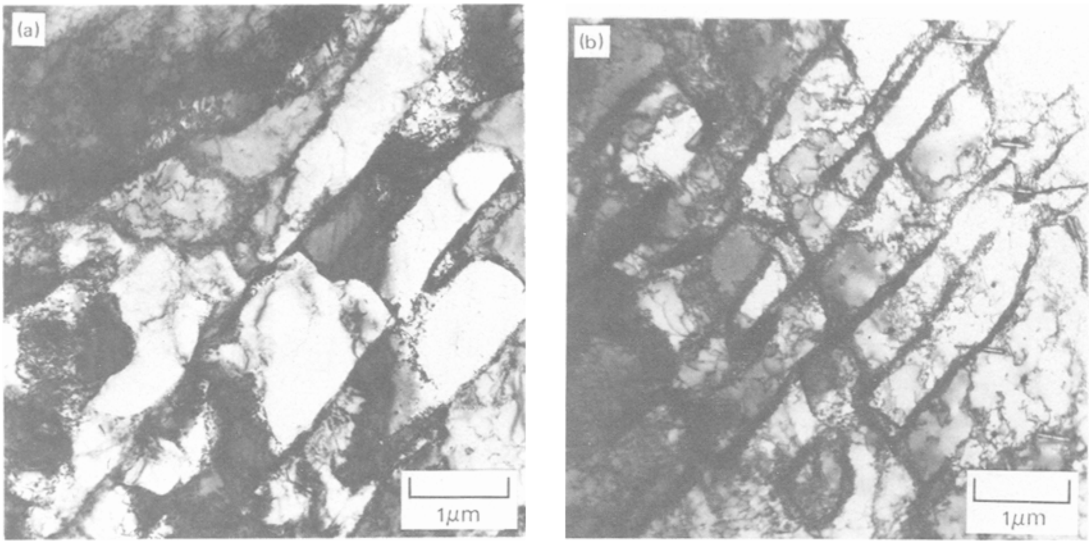

Figure 5 Dislocation cell structure observed after deformation to fracture at 295 K in (a) Ti and (b) Ti–0.1 Er.

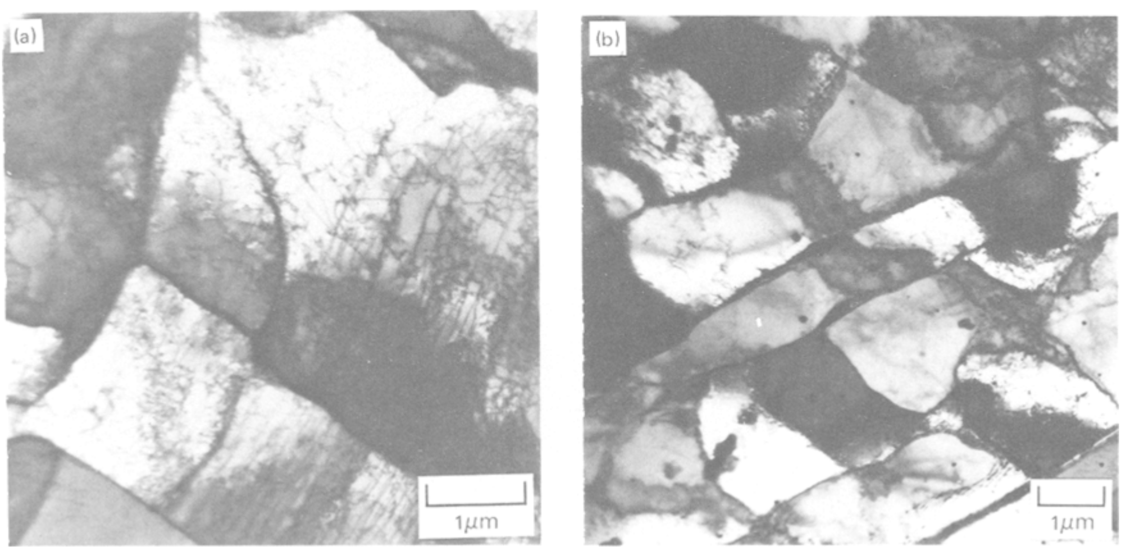

Figure 6 Dislocation cell structure observed after deformation to fracture at 575 K in (a) Ti and (b) Ti–0.1 Er.

Because twinning is a significant deformation mode in titanium [7–9], the influence of Er and Y dispersoids on twinning was examined at 295 K and 575 K. The extent of twinning was significantly higher in the Ti–Er and Ti–Y alloys than in Ti at both temperatures. In contrast to the predominance of dislocation cell structures observed in Ti at high temperature, the Ti–Er and Ti–Y alloys exhibit considerable twinning at 575 K, as is shown in Fig. 7, and twinning is more pronounced in large grain size specimens.

4. Conclusions

(1) The yield stress at 295 K and 575 K of titanium is increased by Er and Y additions, with the dispersion strengthening being greater at the higher temperature and for larger grain sizes.

(2) While titanium exhibits a Hall–Petch dependence of yield stress on grain size, the Er- and Y-containing alloys do not show a well-defined Hall–Petch relationship.

(3) The work hardening of Ti–Er alloys is less dependent on grain size than in pure titanium.

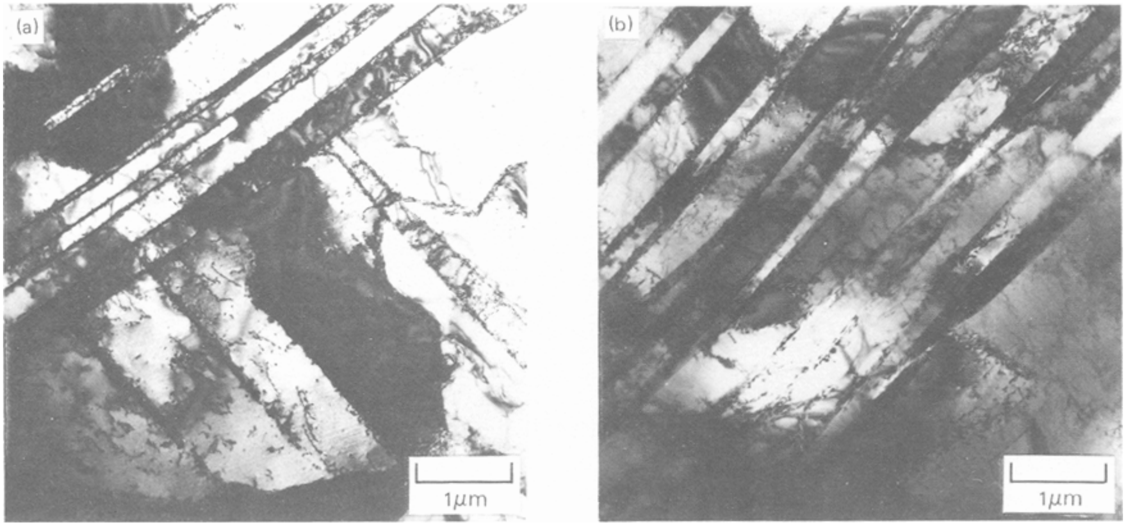

Figure 7 Twin activity in (a) Ti–0.05 Er and (b) Ti–0.3 Y after deformation to fracture at 575 K.

(4) Twinning in titanium is increased at 295 K and 575 K by Er and Y additions.

Acknowledgement

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References

1. B. B. RATH, J. E. O'NEAL and R. J. LEDERICH, Procedings of the Electron Microscopy Society of America, edited by C. J. Arceneaux (Claitors Publ. Div., Baton Rouge, La, 1974) p. 522.
2. B. B. RATH, R. J. LEDERICH and J. E. O'NEAL, "Grain Boundaries in Engineering Materials", edited by J. L. Walter, J. H. Westbrook and D. A. Woodford (Claitors Publ. Div., Baton Rouge, La, 1975) p. 39.
3. E. OROWAN, Symposium on Internal Stress in Metals and Alloys (Institute of Metals, London, 1948) p. 45.
4. P. B. HIRSCH and F. J. HUMPHREYS, *Proc. Roy. Soc. (London)* **A318** (1970) 45.
5. M. F. ASHBY, "Physics of Strength and Plasticity", edited by A. S. Argon (M.I.T. Press, Cambridge, Ma, 1969) p. 113.
6. R. J. LEDERICH, S. M. L. SASTRY, J. E. O'NEAL and B. B. RATH, *Mater. Sci. Eng.* **33** (1978) 183.
7. F. D. ROSI, C. A. DUBE and B. H. ALEXANDER, *Trans. AIME* **197** (1963) 257.
8. A. M. GARDE and R. E. REED-HILL, *Met Trans.* **2** (1971) 2885.
9. N. E. PATON and W. A. BACKOFEN, *ibid.* **1** (1970) 2839.

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