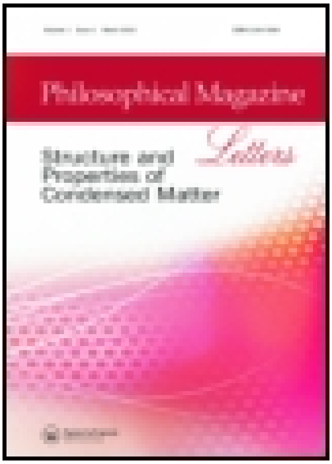

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The influence of second phase Ti_3Al on the deformation mechanisms in TiAl

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The influence of second phase Ti_3Al on the deformation mechanisms in TiAl

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ABSTRACT

Dislocations in samples of a heat-treated and quenched two phase Ti–46 at.% Al alloy have been characterized, and those with Burgers vectors, **b**, given by $\mathbf{b}=\frac{1}{2}\langle 110]$ and $\mathbf{b}=\frac{1}{2}\langle 112]$ have been observed. The microstructure of deformed samples is characterized mainly by dislocations with $\mathbf{b}=\frac{1}{2}\langle 110]$. These dislocations have high Peierls stresses in TiAl of nominal purity, because of the directionality of bonds between the Ti atoms. Therefore, the present observations have been interpreted on the basis that the phase Ti_3Al getters the interstitial elements from the TiAl, since the solubility of these elements in the former phase is significantly larger than in the latter. It is proposed that the removal of interstitials from TiAl in this way decreases the degree of directionality of bonding between the Ti atoms, and so reduces the anisotropy in Peierls stresses caused by these directional bonds. A relatively large number of twins have also been observed in the lamellae of TiAl, and this may be interpreted on the same basis, namely that the mobility of dislocations with $\mathbf{b}=\frac{1}{6}\langle 112]$, the twinning dislocations, is also increased when the concentration of interstitial elements is reduced.

§ 1. INTRODUCTION

In a recent paper (Kawabata, Tadano and Izumi 1988), the effect of purity and second phase on the ductility of TiAl (γ) has been considered. It was concluded by these authors that the ductility is improved by the use of high purity materials and the presence of second phase lamellae of Ti_3Al (α_2). The actual mechanism of ductility enhancement in TiAl itself was not determined unambiguously. It is suggested here that since Ti_3Al has a significantly larger solubility for interstitials than TiAl (Kaufman, Konitzer, Shull and Fraser 1986), the presence of Ti_3Al may act in a way to getter internally interstitials from the TiAl. In this case, it is the effect of a reduction in the interstitial content which causes the ductility increases observed by Kawabata *et al.* (1988). Vasudevan, Court, Kurath and Fraser (1989) have shown that when interstitials are gettered internally by alloying with Er, dislocations with Burgers vector, **b**, given by $\mathbf{b}=\frac{1}{2}\langle 110]$ dominate the deformation microstructure. It should be noted that in the presence of an incidental concentration of interstitial impurities, these dislocations are

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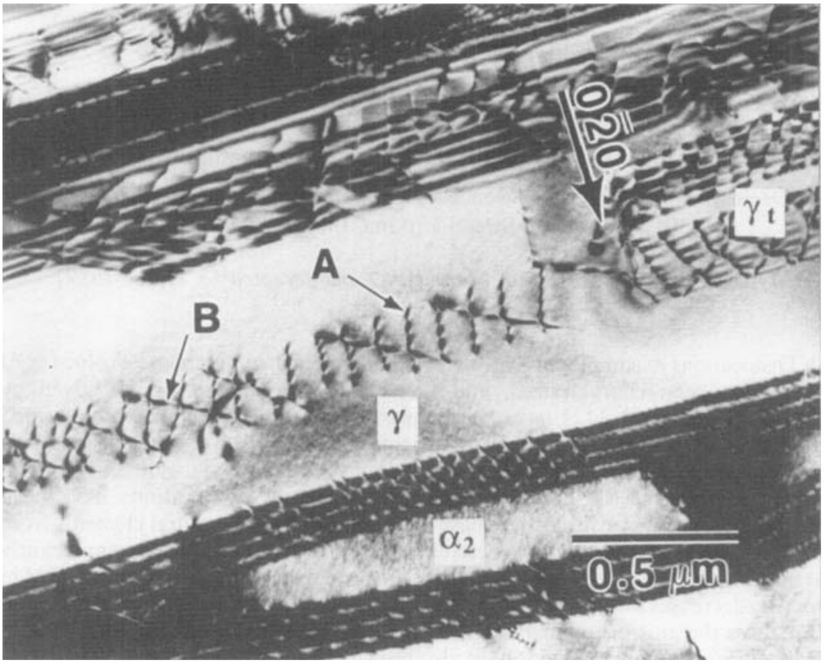
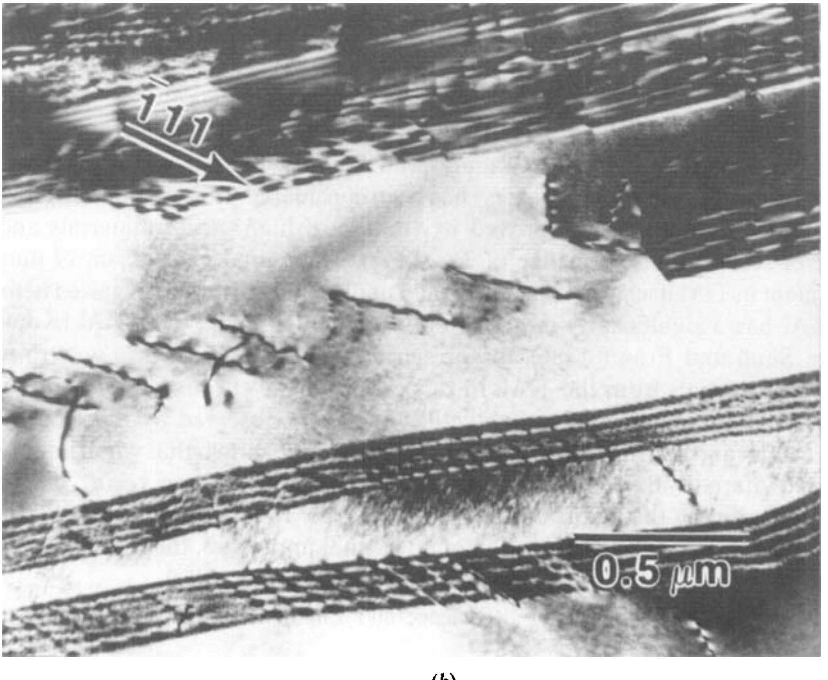
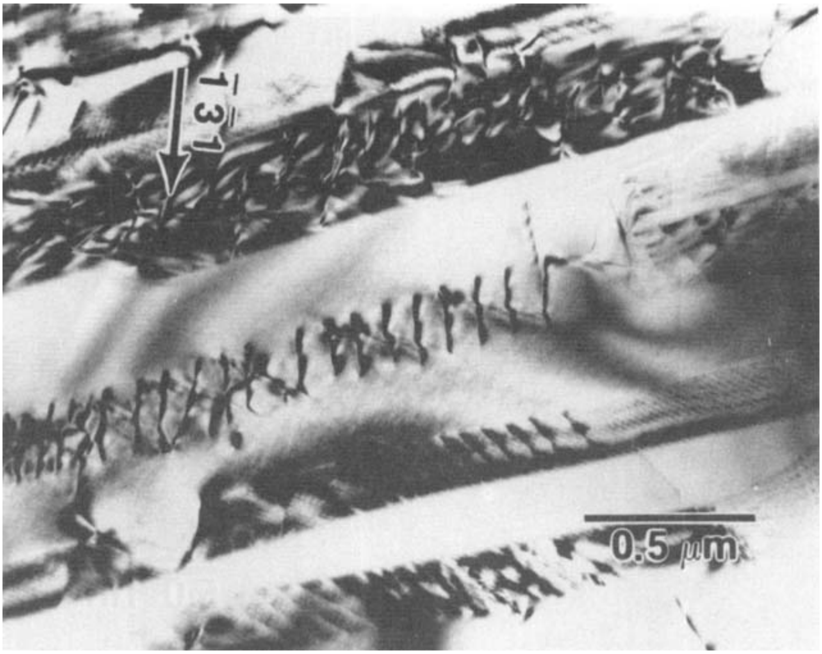
Fig. 1



(a)



(b)



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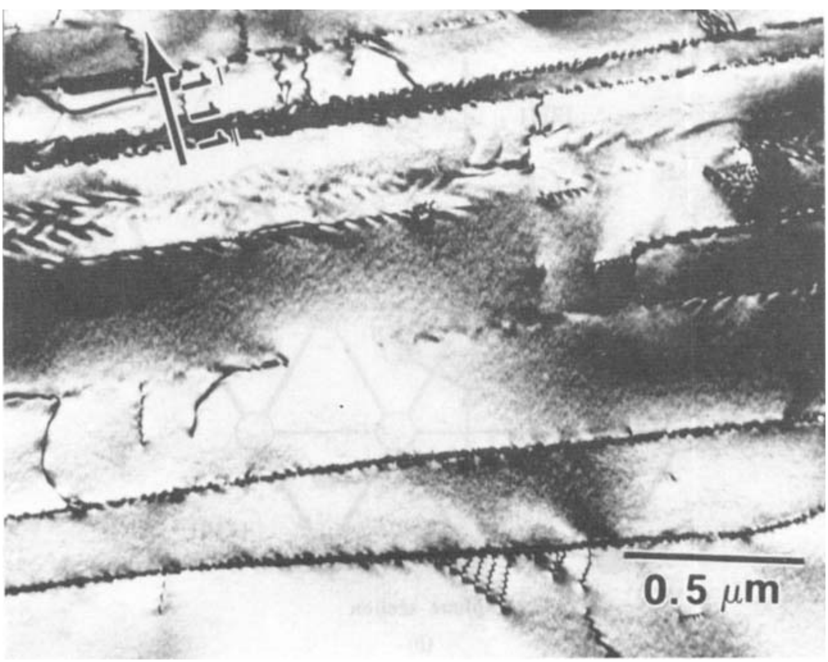
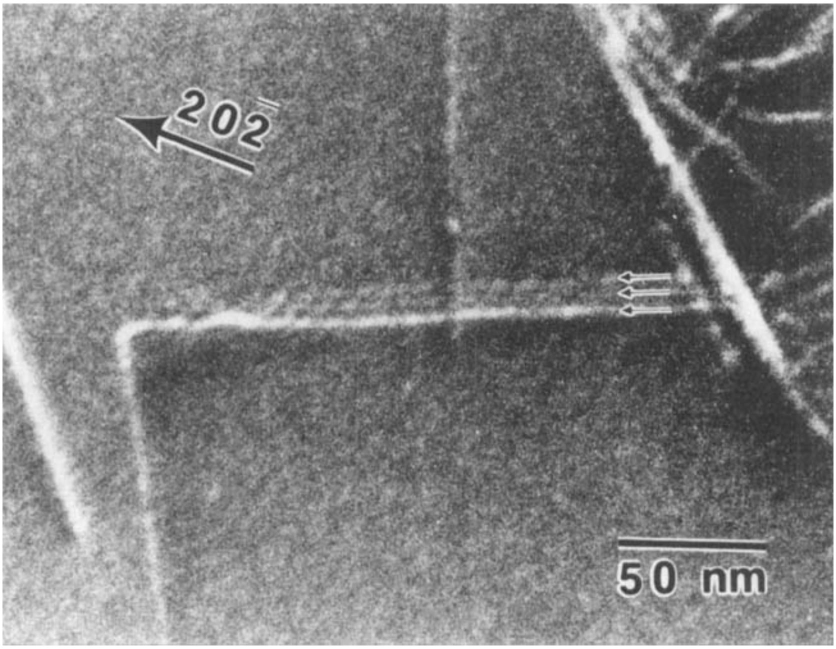


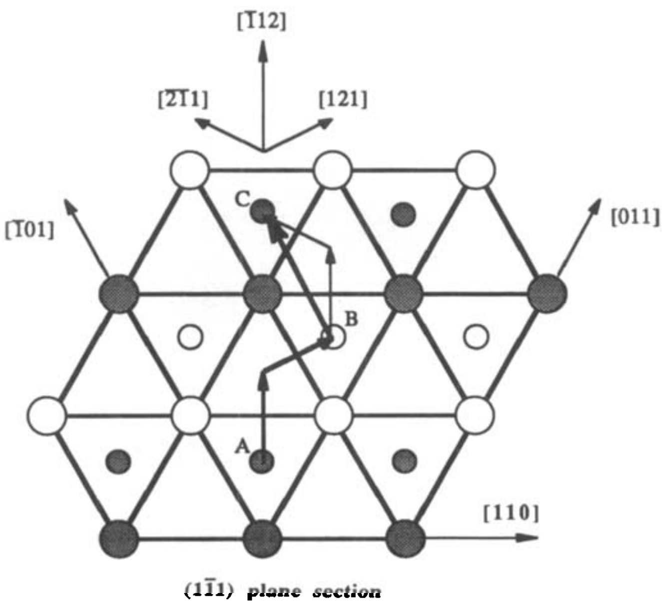
(d)

Bright-field electron micrographs taken from a sample of Ti–46 at.% Al heat-treated according to the schedule given in § 2. Images recorded with (*a*) diffracting vector $\mathbf{g}=0\bar{2}0$, beam direction $\mathbf{B}$ near [101]; (*b*) $\mathbf{g}=\bar{1}\bar{1}1$, $\mathbf{B}$ near [101]; (*c*) $\mathbf{g}=\bar{1}\bar{3}1$, $\mathbf{B}$ near [101], and (*d*) $\mathbf{g}=\bar{1}1\bar{1}$, $\mathbf{B}$ near [121]. Reference is made in the text to the various phases indicated in the figure and also to the dislocation segments marked A and B. See text for explanation.

Fig. 2



(*a*)



(*b*)

(*a*) Weak-beam dark-field electron micrograph taken from the same area as in fig. 2, showing the dissociated nature of a dislocation segment marked B in that figure. Image recorded with $\mathbf{g}=20\bar{2}$, beam direction, $\mathbf{B}$ near [101]. (*b*) Schematic diagram showing the possible dissociations of the dislocation with $\mathbf{b}=\frac{1}{2}[\bar{1}12]$ shown in (*a*). The large and small circles denote atom positions on adjacent planes; the open and shaded circles denote Ti and Al atoms respectively.

considered to be sessile. The reasons for this have been addressed in a recent paper by Court, Vasudevan and Fraser (1989), where it has been concluded that dislocations with $\mathbf{b}=\frac{1}{2}\langle 110]$ are sessile at room temperature because of a large Peierls stress which results from the anisotropy of the charge density about Ti atoms in this compound (Greenberg, Anisimov, Gornostirev and Taluts 1988). This anisotropy of the Peierls stress has a less important effect on dislocations with $\mathbf{b}=\langle 101]$ (Court *et al.* 1989), and at room temperature strain is accomplished by limited glide of dislocations with these latter Burgers vectors (their activity being limited not only by covalency but also by the formation of extrinsically faulted dipoles (Hug, Loiseau and Lasalmonie 1986)), and twinning. If the presence of Ti_3Al acts in a similar manner as additions of Er, then it would be expected that dislocations with $\mathbf{b}=\frac{1}{2}\langle 110]$ would be more glissile in two-phase $(\alpha_2+\gamma)$ mixtures than in single phase TiAl. The present paper shows that this is indeed the case, and therefore aids in developing an understanding of the effect of purity on the mechanisms of plastic deformation of TiAl.

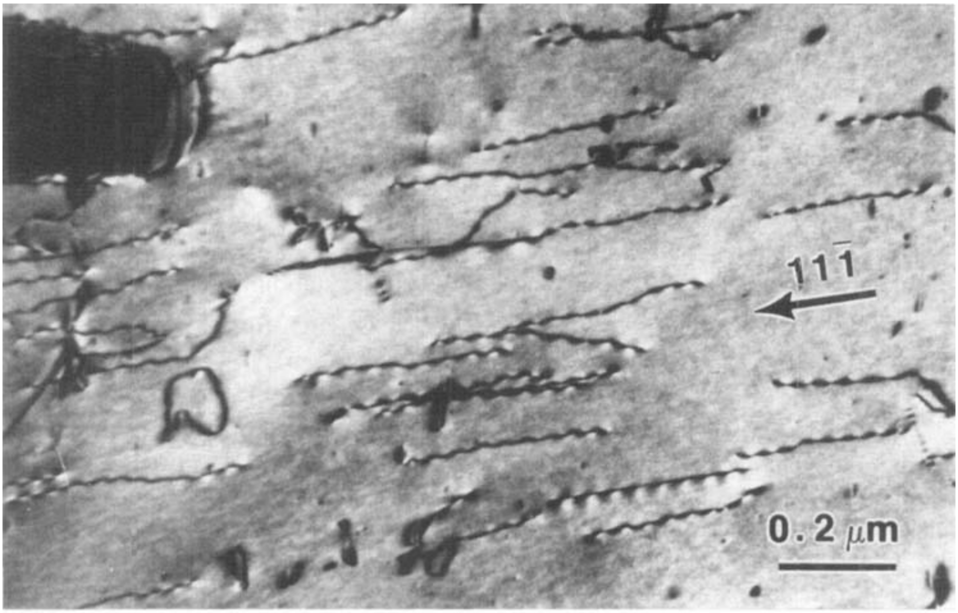
§ 2. Experimental procedure

Master ingots of Ti–46 at.% Al were prepared by non-consumable electrode arc-melting under an argon atmosphere, with 99·9% Ti and 99·999% Al (with respect to metallic elements) used as starting materials. Pieces of the ingots were encapsulated in quartz and heat-treated as follows; each was homogenized for 10 h at 1100°C, heated to 1225°C and held subsequently for 4 h before furnace cooling to 1135°C at a rate of $10°C\,h^{-1}$. After being heat-treated at this temperature for 10 h, the pieces were then furnace cooled to 1090°C at a rate of $10°C\,h^{-1}$, and held for 1 h before water quenching to room temperature. This heat-treatment developed a lamellar microstructure. Compression coupons were cut from this heat-treated material and deformed at room temperature to $\approx 3\%$ strain. Slices of both the heat-treated and deformed alloys were prepared for electron microscopy using standard techniques, and the thin foils were examined in a Philips CM12 transmission electron microscope operating at 120 kV.

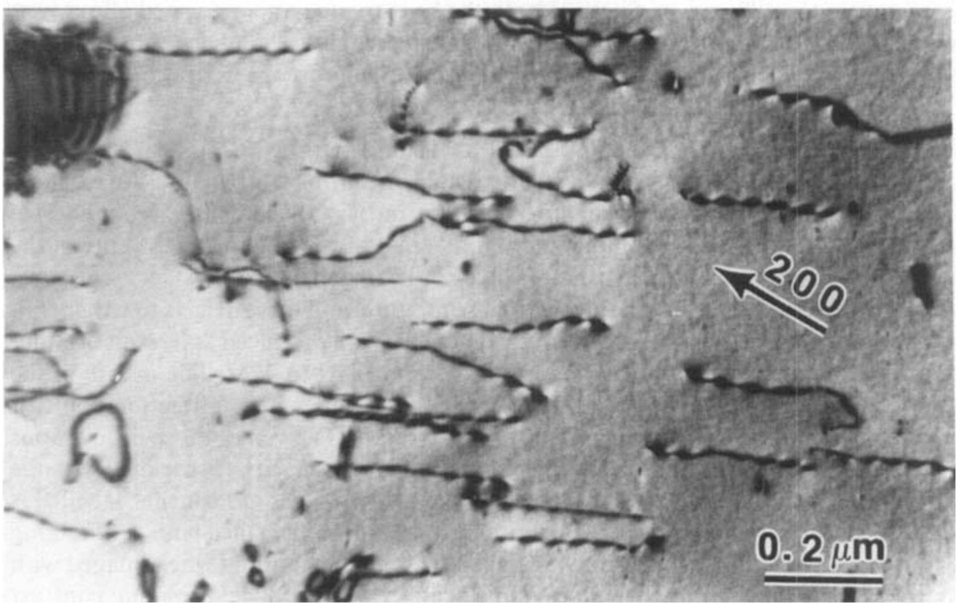
§ 3. Results and discussion

The heat-treatment listed above resulted in a lamellar microstructure. An example of part of this microstructure is shown in fig. 1 (*a*); most of the lamellae are of the TiAl phase, with an appropriate volume fraction of Ti_3Al also being present. Many of the lamellae of TiAl are twinned, an example being indicated in fig. 1 (*a*). In fact, twins are a common observation in TiAl in this two-phase mixture, and these defects together with the dislocations observed, mostly in the TiAl lamellae, are produced as a result of either quenching stresses or transformation of laths from Ti_3Al to TiAl. The nature of the dislocations present in these samples has been characterized by diffraction contrast experiments in the electron microscope, and an example is given of those dislocations lying in the centre of the field of view in fig. 1 and which appear to be associated with the twin marked γ_t in this figure. As can be seen there are two types of segments, labelled A and B in the figure. Those marked A are not in contrast when imaged with a diffracting vector $\mathbf{g}=\bar{1}11$, fig. 1 (*b*), whereas those marked B are not in contrast when imaged with $\mathbf{g}=\bar{1}\bar{3}1$, fig. 1 (*c*); images of both types of dislocation (A and B) yield residual contrast when formed with $\mathbf{g}=\bar{1}1\bar{1}$, fig. 1 (*d*). From this analysis, it is possible to determine the Burgers vectors of these dislocations as that for segments A being $\mathbf{b}=\frac{1}{2}[110]$, and for segments B, $\mathbf{b}=\frac{1}{2}[\bar{1}12]$; the factors of 1/2 in each case are selected since, although they may be dissociated, these dislocations appear in bright-field images to be perfect

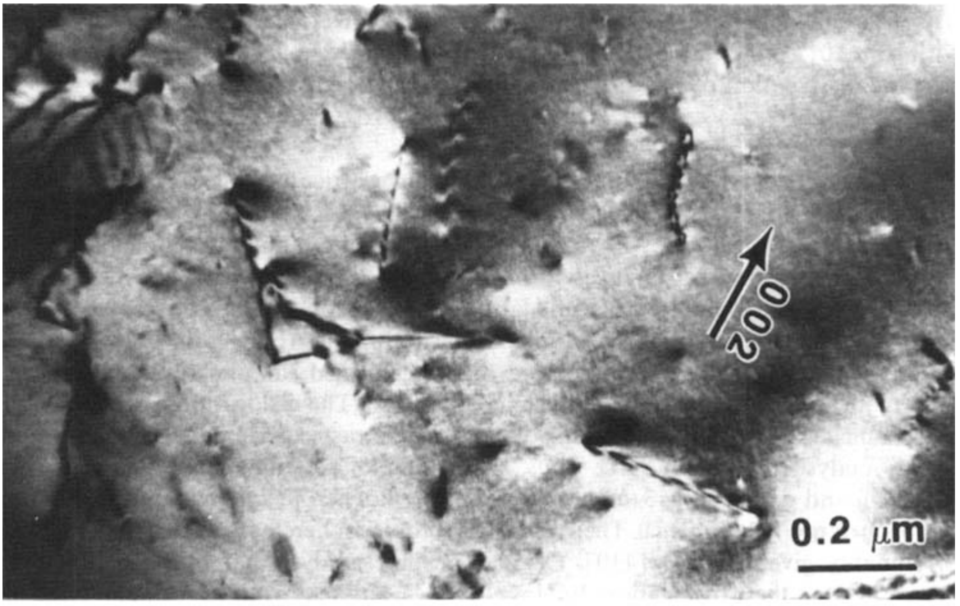
Fig. 3



(a)



(b)



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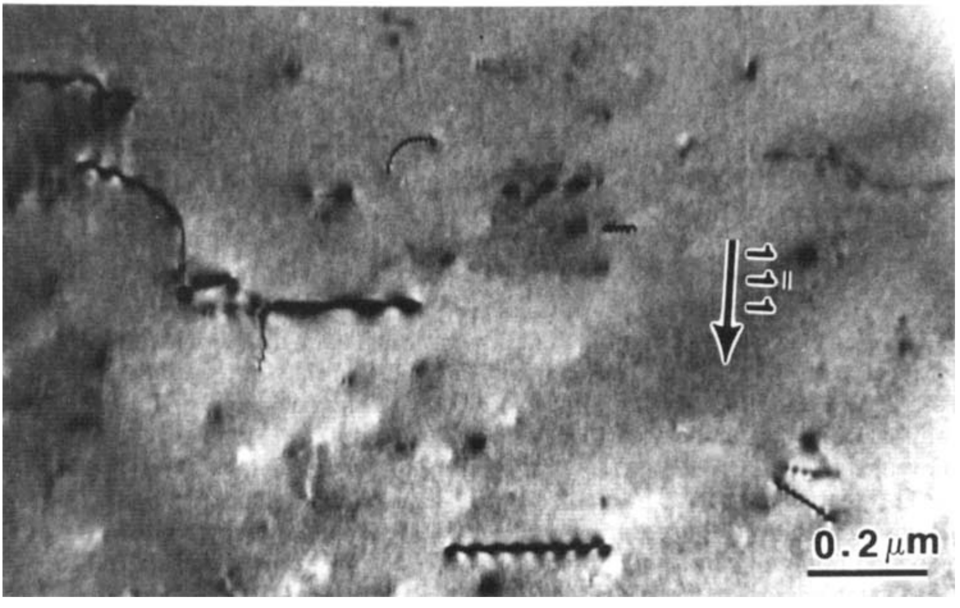

(d)

Bright-field electron micrographs taken from a sample of Ti–46Al (at.%) heat-treated according to the schedule given in section 2 and deformed in compression to ≈3% strain. Images recorded with (a) $\mathbf{g} = 11\bar{1}$, $\mathbf{B}$ near [011]; (b) $\mathbf{g} = 200$, $\mathbf{B}$ near [011] (c) $\mathbf{g} = 002$, $\mathbf{B}$ near [010], and (d) $\mathbf{g} = 1\bar{1}1$, $\mathbf{B}$ near [011].

dislocations. In fact, the latter dislocation (with $\mathbf{b}=\frac{1}{2}[\bar{1}12]$) is indeed dissociated into a triplet, as shown in fig. 2 (*a*). These images are analogous to those of superdislocations with $\mathbf{b}=\langle 101]$ presented by Hug, Loiseau and Veyssière (1988), where in the present case the dissociation of the dislocation appears to be consistent with:

$$\tfrac{1}{2}[\bar{1}12]\rightarrow\tfrac{1}{6}[\bar{1}12]+\mathrm{SISF}+\tfrac{1}{6}[121]+\mathrm{APB}+\tfrac{1}{2}[\bar{1}01]$$

on the $(1\bar{1}1)$ plane, where SISF and APB denote superlattice intrinsic stacking-fault and antiphase boundary respectively. The superpartial with $\mathbf{b}=\frac{1}{2}[\bar{1}01]$ is dissociated in principle into two Shockley-like partial dislocations bounding a complex stacking-fault (CSF), as shown schematically in fig. 2 (*b*), but because of the energy associated with the CSF, the separation of these latter partials is too small to be resolved in the electron microscope using diffraction contrast techniques (Hug *et al.* 1988).

Observations of the microstructure of the deformed samples revealed that the deformation occurred mainly in the lamellae of TiAl. The dislocations present within given lamellae have been identified by diffraction contrast experiments, and an example of this analysis is given in fig. 3. Thus, most of the dislocations imaged with $\mathbf{g}=11\bar{1}$ (fig. 3 (*a*)) and $\mathbf{g}=200$ (fig. 3 (*b*)) are not in contrast when imaged with either $\mathbf{g}=002$ (fig. 3 (*c*)) or $\mathbf{g}=1\bar{1}1$ (fig. 3 (*d*)). Their Burgers vectors are therefore parallel to [110], and presumably given by $\mathbf{b}=\frac{1}{2}[110]$. Crystallographic analysis has been employed to determine that these dislocations lie close to [132], and so are mixed segments lying on $(1\bar{1}1)$. It is interesting to note that this plane is parallel to that of the interface between the two phases, TiAl and Ti_3Al.

The results of the present study may be summarized as follows:

(*a*) most of the dislocations in the undeformed samples have $\mathbf{b}=\frac{1}{2}\langle 110]$ and $\mathbf{b}=\frac{1}{2}\langle 112]$;

(*b*) a large number of twins are present (in the TiAl lamellae) in the microstructure;

(*c*) the deformation microstructure is dominated mainly by mixed segments of dislocations with $\mathbf{b}=\frac{1}{2}\langle 110]$.

These are interesting results since the dislocation microstructure of single-phase TiAl (at room temperature) is characterized largely by dislocations with $\mathbf{b}=\langle 101]$, and rather infrequent twinning. Indeed, dislocations with both $\mathbf{b}=\frac{1}{2}\langle 110]$ and $\mathbf{b}=\frac{1}{2}\langle 112]$ are considered to be more or less sessile in single phase TiAl of nominal purity, containing incidental interstitial impurities, because of a marked anisotropy of Peierls stresses (Court *et al.* 1989). Their presence, particularly in the deformed samples, implies that they are more mobile in TiAl which is part of a two-phase (TiAl/Ti_3Al) mixture than they are in single phase TiAl. This behaviour is similar to that exhibited by TiAl containing particles of Er_2O_3 where dislocations with $\mathbf{b}=\frac{1}{2}\langle 110]$ dominate the deformation microstructure (Vasudevan *et al.* 1989). As stated in § 1 above, in this latter work the rôle of Er is to getter internally interstitial elements from the TiAl matrix, and it was proposed that the removal of interstitial elements leads to increased mobility of dislocations with $\mathbf{b}=\frac{1}{2}\langle 110]$ and $\mathbf{b}=\frac{1}{2}\langle 112]$ because of a reduction in the directionality of bonds between the Ti atoms. Since the presence of Ti_3Al in the microstructure also getters the interstitial elements from the adjacent TiAl laths because of a greater solubility of these elements in Ti_3Al than in TiAl, it seems reasonable to account for the increased mobility of these dislocations in the present work on the basis of the same proposal. It is, then, the degree of purity of the TiAl that is affected by the presence of Ti_3Al, and which is expected to lead in part to an increased ductility, as observed by Kawabata *et al.* (1988). There are other features of the lamellar microstructure which influence ductility, and these will be discussed elsewhere.

The arguments that are used to predict an increased mobility of dislocations with $\mathbf{b}=\frac{1}{2}\langle 112]$ also apply to those with $\mathbf{b}=\frac{1}{6}\langle 112]$, and since these latter defects are order twinning dislocations in this compound (Pashley, Robertson and Stowell 1969), an increased twinning activity is expected; this is consistent with the large number of twins observed in samples of two phase $(\alpha_2+\gamma)$ mixtures. The present results are also consistent with those of Hall and Huang (1988) who report that in deformed two phase $(\alpha_2+\gamma)$ alloys, only dislocations with $\mathbf{b}=\frac{1}{2}\langle 110]$ are observed. Their result may be interpreted in a similar manner to those of the present study, namely that the presence of Ti_3Al tends to act as a getter for interstitial elements in TiAl.

§4. Summary

The microstructure of a heat-treated two phase $(\alpha_2+\gamma)$ alloy is shown to be dominated by dislocations with $\mathbf{b}=\frac{1}{2}\langle 110]$ and $\mathbf{b}=\frac{1}{2}\langle 112]$, as well as twins, and deformed samples are characterized mainly by dislocations with $\mathbf{b}=\frac{1}{2}\langle 110]$. Since these dislocations have high Peierls stresses in single phase TiAl (of norminal purity), it appears that the presence of Ti_3Al reduces their Peierls stresses. This has been attributed to the removal of interstitial elements from the TiAl matrix, since these elements have a larger solubility in Ti_3Al. It is proposed that interstitial elements enhance the degree of directionality of the Ti↔Ti bonds, and so their removal reduces this directionality and therefore the anisotropy of the Peierls stresses. The relatively frequent observation of twins is also understood on this basis, since these defects would be promoted by the increased mobility of dislocations with $\mathbf{b}=\frac{1}{6}\langle 112]$, the twinning dislocations. The enhanced ductility exhibited by two-phase $(\alpha_2+\gamma)$ mixtures may be due in part to the increased mobility of dislocations with these Burgers vectors.

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