portional to $1/\sigma$ [14]. Therefore, from dielectric loss measurements of τ against $1/T$ or from initial-rise analyses of the TSDC, an ionization energy of $[Li]^0$ centres can be evaluated.

A single peak in the dielectric loss spectrum has a relaxation energy of 0.73 eV and an oscillation frequency of 197 cm^{-1} [16], the latter being about one-half of an oscillation frequency of 385 to 485 cm^{-1} for lithium ions in various oxide glasses [9]. Equation 2 was used to calculate T_m for the polarization which is relevant to a single loss peak. T_m thus calculated, 250 K, is proximate to the 255 K peak (Fig. 3). If the polarization arises by ionic conduction within Li_2O precipitates rather than by ionization of $[Li]^0$ centres, the activation energy for conduction in Li_2O crystal must be equal to or close to 0.73 eV. A.c. conductivity measurements [18] reveal H_c of 0.72 to 0.77 eV for four samples out of six of Li_2O polycrystals. Therefore, it seems better to ascribe the relevant polarization as arising from ionic conduction within Li_2O precipitates rather than ionization of $[Li]^0$ centres within micro-galaxies.

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Grain-boundary deformation behaviour in a metastable beta-titanium alloy

Recent studies of tensile and bending fracture in heat-treated Ti–4.5 Al–5 Mo–1.5 Cr (CORONA 5) weld metal have shown that room-temperature sliding can occur at grain-boundary alpha/transformed-beta interfaces [1]. The mechanism of sliding differs fundamentally from that reported for grain-boundary sliding during high-temperature creep, and appears to involve highly localized, shear-induced deformation at the grain-boundary alpha/transformed-beta interface. In order to determine the universality of this phenomena for titanium alloys the present programme investigated the deformation behaviour of the metastable-beta titanium alloy Ti–15 V–3 Cr–3 Al–3 Sn.

Studies were conducted on sheet material approximately 2 mm thick which was given one of two different heat treatments:

(1) solution heat-treated at 788° C/3 h-WQ to obtain an entirely beta microstructure;

(2) solution heat-treated at 788° C/3 h-WQ + aged at 675° C/6 h-AC, which resulted in a Widmanstätten plus grain-boundary alpha microstructure in a retained beta matrix.

Following thermal treatments, blanks were machined into bend specimens, mechanically polished through 600 grit SiC paper, and electropolished at −45° C in a perchloric acid

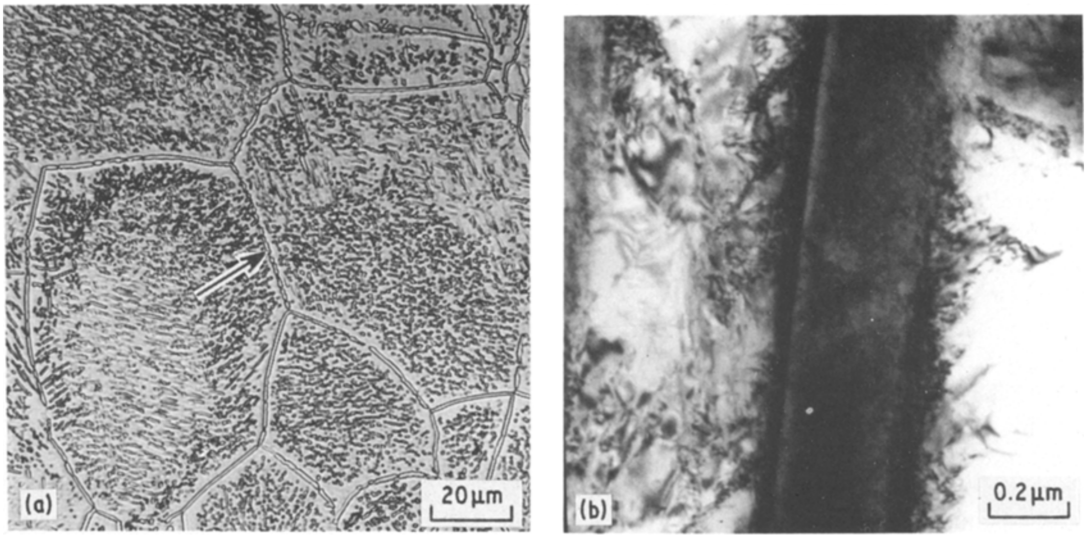


Figure 1 Widmanstätten plus grain-boundary alpha microstructure of solution heat-treated and aged Ti–15–333: (a) light micrograph, arrow indicates beta grain boundary; (b) TEM showing grain-boundary alpha (dark), a dense dislocation substructure is evident.

electrolyte described by Williams and Blackburn [2]. Deformation was provided by progressive bend testing around dies of known radius. Optical, scanning-electron, and transmission-electron microscopy, and electron microprobe analysis techniques were employed in microstructural characterization and in the study of deformation and crack-initiation behaviour.

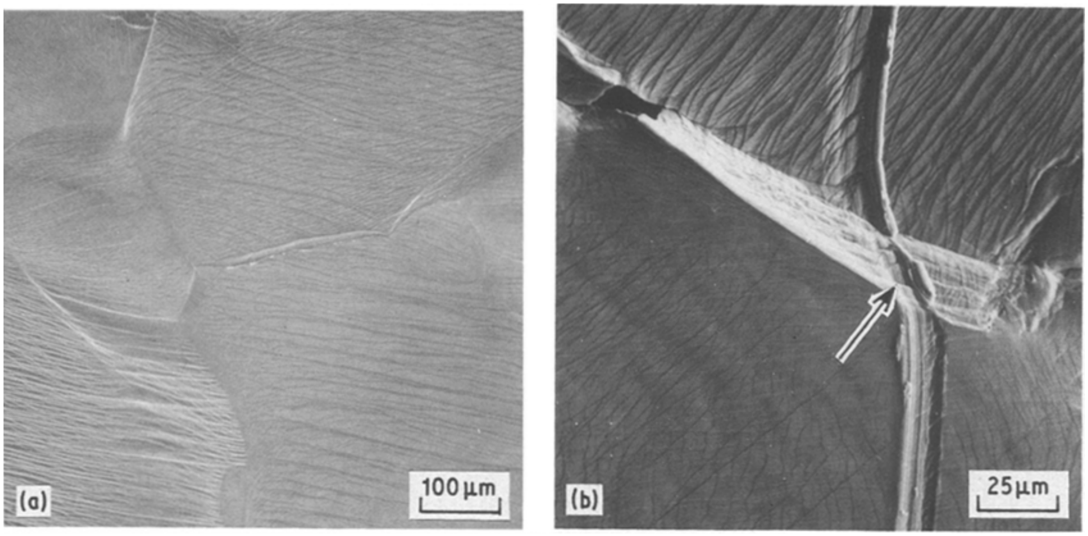
Beta grains in the solution heat-treated alloy exhibited an omega-type instability, evidence of which was diffused intensity lines on the TEM diffraction patterns. Microprobe analysis of the single-phase beta alloy indicated an absence of alloying element segregation at grain boundaries. The microstructure of the aged alloy consisted of Widmanstätten plus grain-boundary alpha in a beta matrix (Fig. 1a). The TEM in Fig. 1b of a specimen which was severely deformed shows a large dislocation density and an absence of interphase phase at the grain-boundary alpha/beta interface. As expected, the alpha plus beta microstructure was characterized by a partitioning of solute elements to the alpha and beta phases (Table I).

TABLE I Microprobe analysis of grain-boundary region in solution heat-treated and aged Ti–15–333

	Composition (wt %)	
Alloy element	GBα	Beta adjacent GBα
Al	4.12	2.63
V	7.00	16.28
Cr	1.39	4.08
Sn	2.38	1.93

Fig. 2a shows extensive slip occurring throughout beta grains in the deformed solution heat-treated microstructure. Grain-boundary separation, which resulted from slip incompatibilities, is illustrated in Fig. 2b. In general, it was observed that fractured surfaces at the crack initiation sites in the single-phase, retained-beta microstructure were covered with irregular-shaped dimples (Fig. 2c). Deformation of the two-phase microstructure was typically characterized by sliding at the grain-boundary alpha/retained-beta interfaces. A large step formed by the sliding process is shown in Fig. 3a and b. It is of interest to note: (1) the lack of observable macroscopic deformation in the adjacent grains; and (2) the absence of a crack at the interface (this was confirmed by carrying out various tilting experiments in the SEM). From Fig. 3c it appears that grain-boundary alpha/beta interface sliding was followed by extensive deformation, which can be seen by gradual bending of the alpha platelets. Finally, Fig. 3d shows a step formed by grain-boundary alpha/beta interface sliding which experienced a transition to dimple formation at the interface, possibly because of a change in the state of stress at the crack tip.

The present work suggests that sliding requires



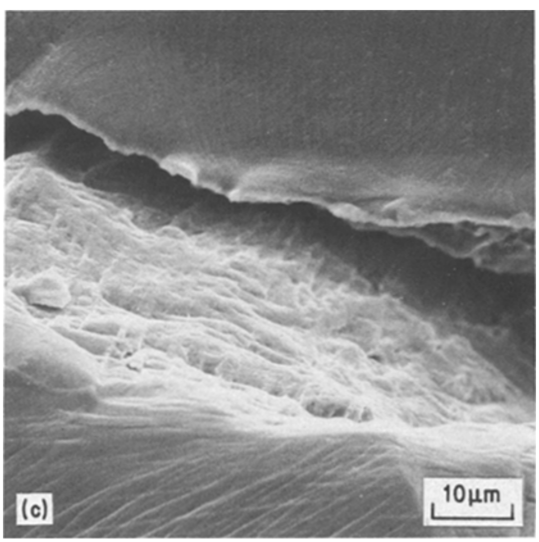


Figure 2 Deformation and crack initiation characteristics of solution heat-treated Ti–15–333: (a) extensive homogeneous slip; (b) inhomogeneous nature of the slip at beta grain boundary as indicated by bending of fiducial scratch (arrow); (c) crack initiation at beta grain-boundary.

the co-existence of alpha and beta phases, and that it can occur without the presence of interface phase. It is of interest to note that visible sliding was observed only at grain-boundary alpha/beta interfaces and not at Widmanstätten alpha/beta interfaces.

Generally, the formation and growth of grain-boundary alpha is associated with the rejection of beta stabilizers across the interface into the beta phase, which leads to the creation of steep concentration gradients. Such gradients have, in fact, been observed in the studies of CORONA 5 [1] using electron-microprobe analysis and in Ti–6 Al–2 Sn–4 Zr–2 Mo [3] using a high resolution "dedicated" scanning-transmission electron microscope. Considering these results, it might be expected that the region in the vicinity of the interface would have a composition intermediate between the alpha and the beta phases, including critical compositions which have been shown to exhibit interesting features in titanium alloys, such as a very low elastic modulus. This critical composition could also have a lower elastic shear modulus given by $C' = (C11 - C12)/2$, which may be related to the occurrence of (reversible) martensitic phase transformations, and thereby be a useful parameter for estimating the relative stability of b c c crystal structures [4–6]. Thus, the application of stress could cause localized, stress-induced phase transformation [7–11] along the alpha/beta interfaces and lead to a large amount of shear-induced sliding. In addition, it is also possible that interfacial dislocations may participate in the shearing process. Finally, since the alpha/beta interfacial sliding has been observed to occur at low stresses and was most typically associated with an absence of observable macroscopic plastic deformation in grain-boundary alpha, it can be stated that such deformation may not play a major role in sliding. In conclusion,

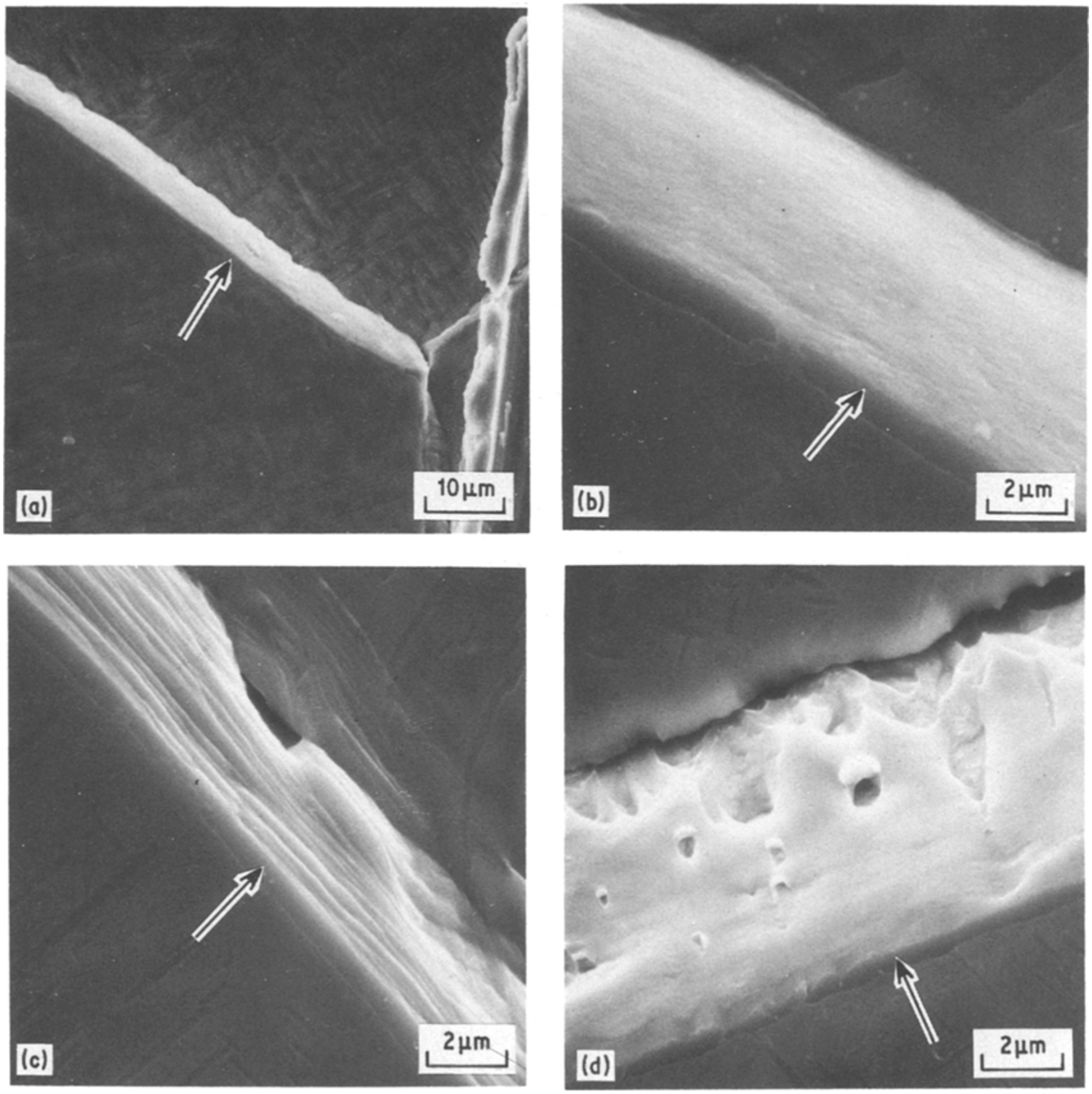


Figure 3 Deformation and crack-initiation characteristics in solution heat-treated and aged Ti–15–333; (a) and (b) step formation at grain-boundary alpha/beta interface; (c) evidence of slip on previously formed step; (d) transition from interface sliding to crack opening and dimple formation, arrows indicate grain-boundary alpha.

the observation of grain-boundary alpha/beta interface sliding in Ti–CORONA 5 [1], Ti–Mn alloys [12], and Ti–Mo [3] alloys indicates universality of this phenomena in titanium alloys under certain conditions.

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