The above discussion of twinning and pressure dependent diffusion controlled creep indicate that either can account for the large magnitude of the SD effect in α-Pu. At present the author favors the diffusion controlled model since creep of α-Pu does appear to be diffusion controlled in this stress range and no twins were observed. However, since the mechanisms of plastic deformation in α-Pu are not completely understood, further microstructural and phenomenological observations of creep are necessary to resolve this question. Clearly the relation of the mentioned structural changes during deformation to flow stress is not yet understood. The SD effect in α-Pu is larger than that in any other metal and this warrants further study.

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Corrections to *Met. Trans.*, 1972, vol. 3

The Influence of Oxygen Concentration on the Internal Stress and Dislocation Arrangements in α Titanium, by J. C. Williams, A. W. Sommer, and P. P. Tung, pp. 2979-84.

Pages 2981-84

Beginning at the bottom of page 2981, the published text was not in the proper sequence. The following should be substituted. Please refer to above reference for additional text and figures.

B) Thin Foil Electron Microscopy

Thin foils were prepared from the gage sections of tensile specimens of all five types of material which had been deformed at temperatures between 77 and 540 K. Twinning was observed in types I through IV materials deformed at all temperatures studied. The twin thickness decreased with increasing temperature and oxygen content as did the twinning frequency. Selected area electron diffraction and trace analysis showed that nearly all the twins were of the $\{10\bar{1}2\}$ type. In all cases where twins were observed, profuse slip had also occurred in both the matrix and the twin. Burgers vector determination experiments showed that the deformation had almost exclusively occurred by **a** vector slip. There were also appreciable numbers of **c**+**a** type dislocations present, however these dislocations were confined to regions near the twin: matrix interfaces. All materials showed a trend toward more uniform distributions of dislocations with increasing temperature or with decreasing oxygen concentration at low temperatures ≤ 300 K. With the exception of the types I and II material, a transition from planar to wavy glide was observed; the temperature range over which the transition occurs increases with increasing oxygen concentration. When planar glide occurs, the preferred glide planes are $\{10\bar{1}0\}$ and $\{10\bar{1}1\}$ with essentially no well-developed basal bands being observed. Figs. 3(*a*) to (*c*) and 4(*a*) to (*c*) illustrate the slip mode transition below and above room temperature in types III and V material, respectively. The types I and II material did not show a well-defined transition to planar glide even at 77 K, as is shown in Figs. 5(*a*) to (*c*). In specimens of types III through V material deformed at temperatures within the transition range, increased numbers of dislocations cross slipping out of the bands are seen and large, square-ended loops are frequently observed, as illustrated in Fig. 6(*a*). Burgers vector determination on these types of structures showed that the square-ended loops have the same **a** vector as the dislocations within the bands. In the planar arrays, most of the dislocations are jogged screws, as shown in Fig. 6(*b*). The numbers of the jogs and the size of the jogs increase with increasing deformation temperature. Large numbers of dipoles and loops are also observed at all temperatures, the size of the loops and dipoles increases with increasing temperature. The dipoles are frequently lined up along [0001], probably indicating that they lie on $\{10\bar{1}0\}$ planes. Fig. 7 illustrates these loops and dipoles. Occasional hydrides, which are seen in the foils, Fig. 3, can be precluded by electropolishing at temperatures $< -40°C$. Thus, the hydrides observed are ascribed to hydrogen pick up during electropolishing.

The small regions of second phase observed by optical metallography in the type III material have been identified in thin foils as the bcc β phase using selected area electron diffraction techniques. The existence of β phase in commercial purity titanium has been previously documented by Curtis, *et al.*[7] who ascribe the β phase to the iron content of these alloys.

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DISCUSSION

The preceding results have shown several trends in α titanium deformation behavior which require further discussion. These are: the rapid increase in τ_{int} with decreasing temperature in the higher oxygen materials, the wavy to planar slip mode transition and the maximum in the τ^*-T plots. The following discussion will suggest qualitative explanations for the first two of these results.

If the τ_{int}-T and τ^*-T curves, Figs. 1 and 2, are compared with the dislocation arrangements at various temperatures and oxygen contents, Figs. 3 to 5, it becomes apparent that the temperature which marks the onset of planar glide also marks the rapid rise in τ_{int} and the maximum in τ^*. Further, it appears that the magnitude of both the low temperature increase in τ_{int} and the maximum in τ^* is larger in materials which exhibit narrower slip bands, *i.e.*, in higher oxygen materials. No slip mode transition was observed in the types I and II material, and, as seen in Figs. 1 and 2, no marked rise in τ_{int} or maximum in τ^* is observed in these materials. Additionally, it should be noted that planar glide persists to higher deformation temperatures with increasing oxygen concentration, and that the τ_{int} values for the higher oxygen materials only approach those for the lower oxygen levels when wavy glide occurs in both materials. Previous investigations have attributed the strong temperature dependence of the flow stress to rapid increases in τ^*[8-11] whereas our results indicate that the temperature dependence of τ_{int} is also an important contributing factor, at least in materials where a slip mode transition occurs.

It is useful to consider why increasing oxygen might cause a transition to planar glide in α titanium at low temperatures. In this connection, an attractive suggestion is that the ordering tendencies exhibited in more concentrated Ti-O solid solutions[10] are preceded by short range ordering in the more dilute solid solutions studied here thereby producing coplanar slip. Direct proof of SRO (or lack thereof) is experimentally very difficult since oxygen has a low scattering power for X-rays and electrons. Such an explanation for coplanar slip at low temperature would also require a disordering of the dilute solid solution in the vicinity of room temperature to be consistent with our observations. This may not be unreasonable since the yield point observed in materials III to V during initial straining returns after short aging times at room temperature and above, indicating that the oxygen has some mobility in this temperature range. In this connection, the dependence of τ_{int} on oxygen concentration, c, varies with temperature in the materials we have studied. τ_{int} is approximately proportional to c at temperatures in the planar glide regime for material types III through V and it is approximately proportional to $c^{1/2}$ at higher temperatures. Such a change in concentration dependence is certainly consistent with a rearrangement of the oxygen atoms.

The interrelation between slip mode, τ_{int} and τ^* cannot be explicitly set forth at present. However, inspection of the dislocation arrangements in Figs. 3(*a*) and (*c*) shows that, when compared to the tangled dislocation arrangements, the average distance between dislocations is less within the planar bands and that the dislocations (all of which have the same **b**) within one band are almost parallel. Both of these factors can possibly contribute to the observed increase in average internal stress which opposes the motion of dislocations when they are confined to planar slip bands. The connection between the maximum in τ^* and the onset of planar slip is not clear and, in fact, the maximum in a τ^*-T plot is itself anomalous in pure metals. A maximum in strain rate sensitivity-temperature plots has been reported in the Nb-H system,[12] but the origin of this maxima has not been analyzed in terms of slip mode.

Throughout the paper we have called τ_{int} the component of the flow stress obtained by stress relaxation measurements or from negative relaxation rate experiments, both of which have been described. Theories of thermally activated flow formally divide the flow stress into thermally activated and athermal components, the latter being designated τ_a. The exact relationship between τ_a and the experimentally determined parameter, τ_{int}, has recently been questioned[13] but in the present paper we have not attempted to interrelate the measured quantity and its theoretical counterpart. Instead we have taken an experimentally measurable quantity, τ_{int}, and have shown that the change in temperature dependence of this parameter corresponds to a change in slip mode. In this connection, it should be noted that the values of τ_{int} obtained from two unrelated techniques are in good agreement. Further, it has been recently shown that τ_{int} values in LiF obtained by these two methods plus strain rate cycling all show good agreement.[14] It thus appears that τ_{int} is physically significant, but the precise interpretation of this parameter, as well as its ultimate designation, is still an unsettled question.

CONCLUSIONS

1) An increase in the temperature dependence of the relaxed portion of the flow stress, designated τ_{int}, has been correlated with a wavy to planar slip transition.

2) The maximum in τ^* occurs at the temperature which corresponds to the onset of planar glide.

3) The transition from planar to wavy slip is the result of increased difficulty of cross slip at lower temperatures or at higher oxygen content.

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Corrections to *Met. Trans.*, 1973, vol. 4

Diffusion of Hydrogen in Titanium Alloys Due to Composition, Temperature, and Stress Gradients, by J. L. Waisman, George Sines, and L. B. Robinson, pp. 291-302

Page 300, Equation [2.8]

The number 3 was omitted from the denominator of this equation. The equation should read:

$$J_m = -D\left[\frac{dC_m}{dx} + 3.55C_m\frac{dN_{AL}}{dx} - \frac{2.0C_m}{3RT}\frac{d\tau}{dx} + \frac{(5300-583N_{AL})}{RT^2}C_m\frac{dT}{dx}\right] \qquad [2.8]$$

An Analysis of Plastic Instability in Pure Shear in High Strength AISI 4340 Steel, by Kohichi Tanaka and Joseph W. Spretnak, pp. 443-54

Page 443

The equation on page 443 should read:

$$d\sigma = \left(\frac{\partial\sigma}{\partial\epsilon}d\epsilon + \frac{\partial\sigma}{\partial\dot{\epsilon}}d\dot{\epsilon} + \frac{\partial\sigma}{\partial T}dT\right) = 0$$