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Abstract

Stress-corrosion cracking has been studied in commercial titanium alloys Ti-8Al-1Mo-1V, Ti-6Al-4V, and Ti-4Al-3Mo-1V. Electron fractography has been used in conjunction with electron metallography to establish the fracture path. A sharp transition from ductile to cleavage failure occurs in the alpha phase between 4 and 6 weight percent aluminum, while the beta phase fails by ductile rupture in all cases. Thin-foil electron microscopy has been used to show that the dislocation arrangements produced in the alpha phase range from tangles in the 4 weight percent aluminum alloy to coplanar arrays in the 6 and 8 weight percent alloys. The relation between dislocation arrangements and fracture mode offers a qualitative explanation of the variations in environmental susceptibility of various titanium alloys.

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Introduction

Stress-corrosion cracking (SCC) of titanium alloys in salt water environments has recently been the subject of intensive investigations.[1,2] As a result, the relative susceptibility of many of the commercial alloys has been catalogued, although no attempt has been made to explain the observed differences in susceptibility. Alloys that are susceptible to stress-corrosion cracking fail by slow growth of a crack under sustained load, followed by fast fracture when the critical crack length is reached. The region of slow crack growth is visually detectable on the fracture surface; an example of this is shown at A in Fig. 1. This study shows that slow-crack-growth regions always exhibit mixed cleavage and ductile fracture; thus susceptibility, as used here, indicates occurrence of slow crack growth and the existence of cleavage in slow-growth regions. In previous work, it has been shown that the degree of susceptibility can qualitatively be related to the areal percent of cleavage fracture in slow-growth regions.

In this study, attention was focused on explaining the occurrence of cleavage fracture in salt water environments of some titanium alloys in terms of metallurgical factors. Electron fractography, electron metallography, and thin-foil electron microscopy were employed to study three commercial titanium alloys. Of these alloys, Ti-8Al-1Mo-1V has been shown to be the most susceptible, Ti-6Al-4V is less susceptible, and Ti-4Al-3Mo-1V is highly resistant based on the criteria outlined above, i.e. slow crack growth and cleavage in the slow-growth zone.

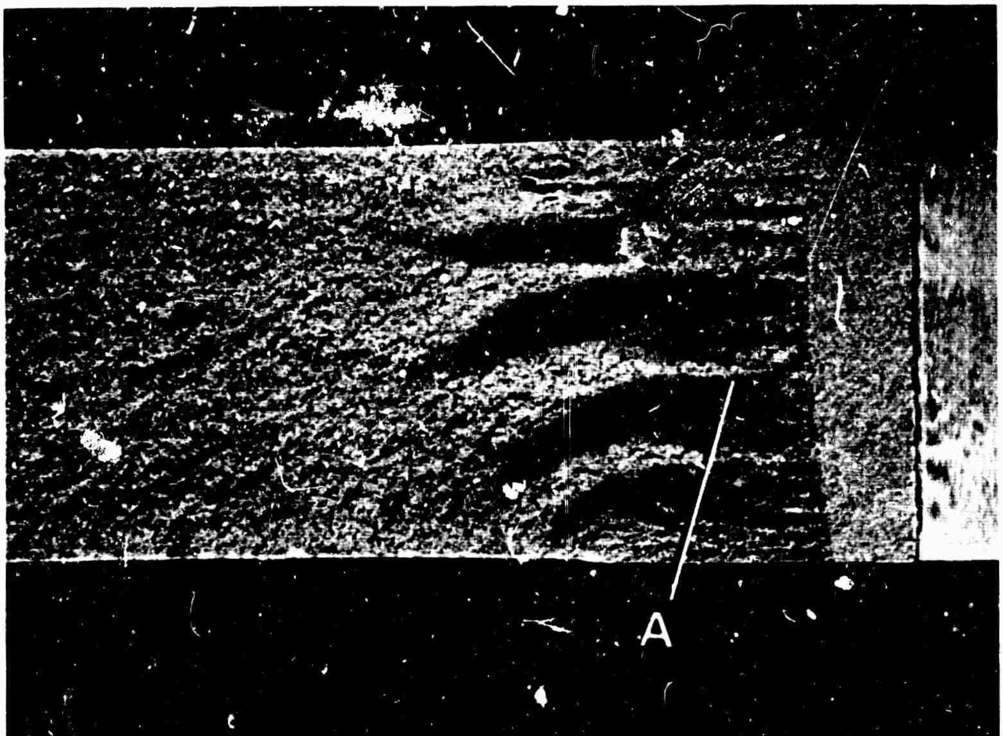


Fig. 1 Specimen Failed in 3½% NaCl Solution, Showing Fatigue Crack, Slow Crack Growth, and Fast Fracture Regions. Magnification: 3X

Experimental Procedures

Two-stage cellulose acetate replicas were used in all cases. Thin foils were prepared using a technique previously described.[3] The heat-treat conditions of the specimens used in this study are listed in Table 1. These specimens were chosen only because they furnished an excellent relation between fracture topography and microstructure. However, the conclusions drawn from tests of these specimens were shown to be valid for a wide variety of heat treatments during a separate, extensive investigation that showed that heat treatment affects the crack propagation rate, but the mode of fracture is the same in all cases.

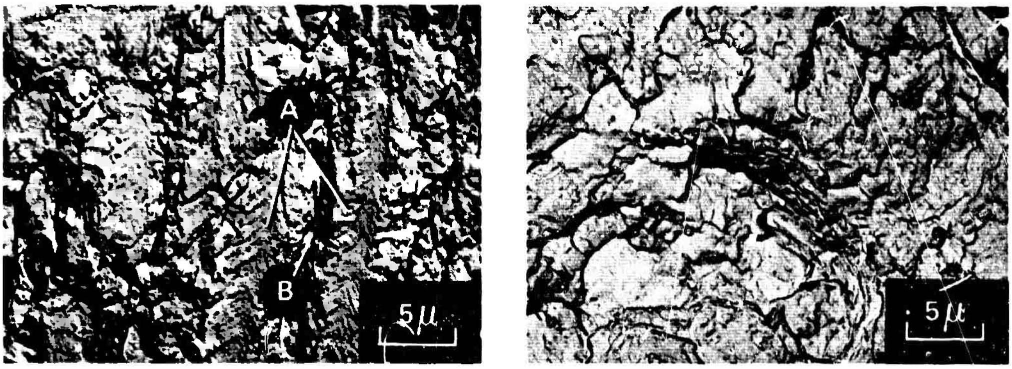
Table I Specimen Heat Treatments and Microstructures

Alloy	Heat Treatment	Microstructures
Ti-4Al-3Mo-1V	Mill annealed (furnace-cooled from 1300°F)	Equiaxed alpha and beta
Ti-6Al-4V	Solution-treated and aged (1725°F/water-quenched + 1250°F/4 hr/air cooled)	Acicular alpha and transformed beta consisting of tempered martensite
Ti-8Al-1Mo-1V	Mill annealed (furnace-cooled from 1450°F)	Acicular alpha and beta; alpha contains small (<50 Å dia) particles of ordered Ti_3Al phase

Results and Discussion

Replicas of the slow-growth regions of Ti-8Al-1Mo-1V and Ti-6Al-4V showed alternating regions of cleavage and ductile fracture as seen at A and B respectively in Fig. 2(a). Since no well-defined slow-growth zone was present on the fracture surface of the Ti-4Al-3Mo-1V, replicas were made adjacent to the fatigue precrack. Figure 2(b) shows that no cleavage fracture was evident in this alloy.

The occurrence of two fracture modes (ductile and brittle) in the slow-growth region was interpreted as possible evidence for different behavior of the alpha and beta phases in the presence of a salt water environment. It was necessary to verify this observation and to identify the phase that fails in a brittle manner. To accomplish this, microstructure replicas were prepared from metallographic samples of Ti-8Al-1Mo-1V and Ti-6Al-4V, and the fracture topography and microstructure of these alloys were compared (Figs. 3 and 4) at constant magnification. The ductile regions of Fig. 3(a) correspond to the thin lamellae of the beta phase in Fig. 3(b), and the ductile regions of Fig. 4(a) correspond to the transformed beta regions in Fig. 4(b). From this it is concluded that in both alloys the primary alpha phase is the constituent that fails by cleavage, and, in general, the beta phase and transformed beta have been found to behave in a ductile manner in all heat-treat conditions examined. The V-shaped



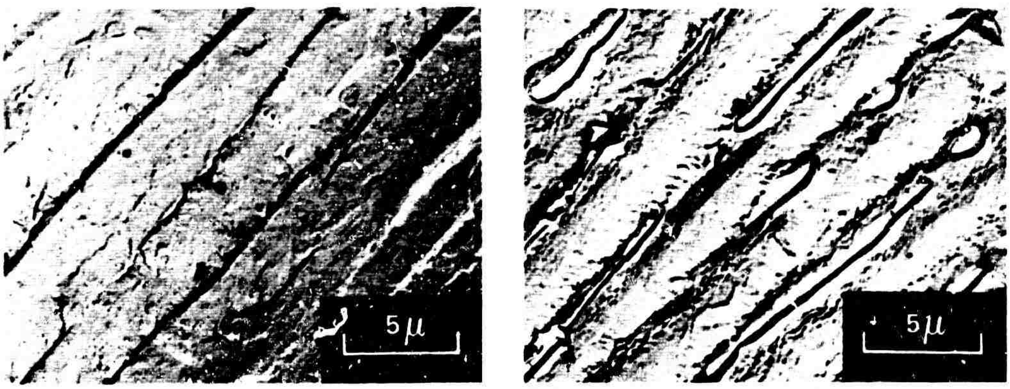
(a) Alternating Regions of Cleavage and Ductile Rupture in a Ti-6Al-4V Specimen

(b) Ti-4Al-3Mo-1V Specimen Failed in 3½% NaCl Showing Absence of Cleavage

Fig. 2 Electron Fractograph of Specimen Failed in 3½% NaCl Solution

markings, such as shown at A in Figs. 2(a) and 4(a), are considered to represent successive positions of the crack front and have been shown to point in the direction of crack propagation since they point in the opposite direction to tear dimples in the adjacent ductile bands. Thus, the crack front lags behind at the alpha-transformed beta interface as if it were pinned, indicating that the beta or transformed beta regions retard crack propagation by acting as ductile inclusions. This result can be generalized for all alpha-beta alloys to predict that a reduced susceptibility for a given alloy will accompany either an increase in volume fraction or a more uniform dispersion of beta or transformed beta phase.

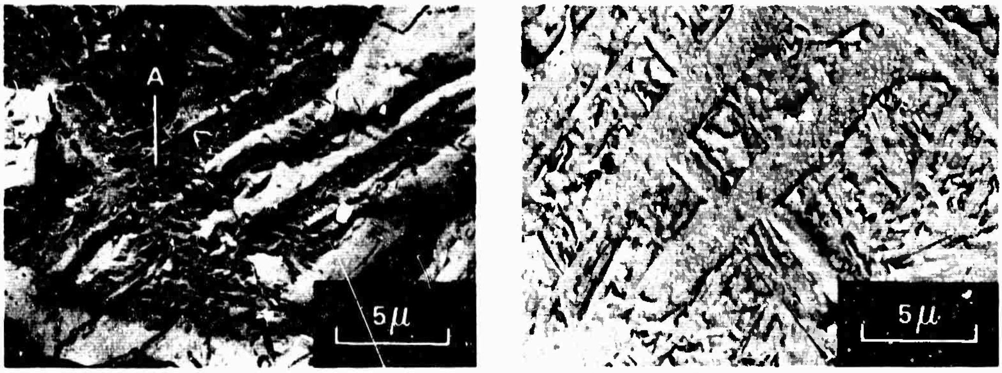
Occurrence of transgranular SCC in other alloy systems has been related to large, localized components of the applied stress normal to the fracture plane that result from coplanar dislocation arrays.[4] Other workers' results show that such arrays can occur in titanium alloys due to preferential slip on a single slip system. These results



(a) Electron Fractograph

(b) Electron Micrograph

Fig. 3 Ti-8Al-1Mo-1V Failed in 3½% NaCl Solution, Showing Correlation Between Microstructure and Fracture Topography



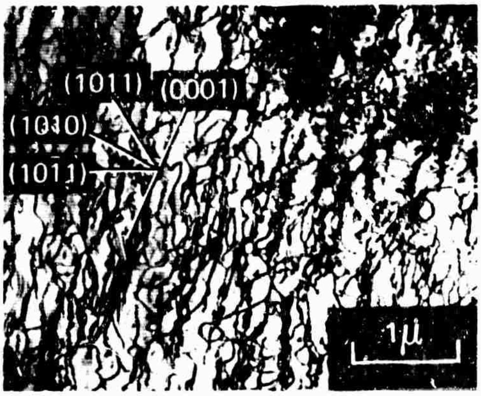
(a) Electron Fractograph *(b) Electron Micrograph*

Fig. 4 Ti-6Al-4V Failed in 3½% NaCl Solution, Showing Correlation Between Microstructure and Fracture Topography

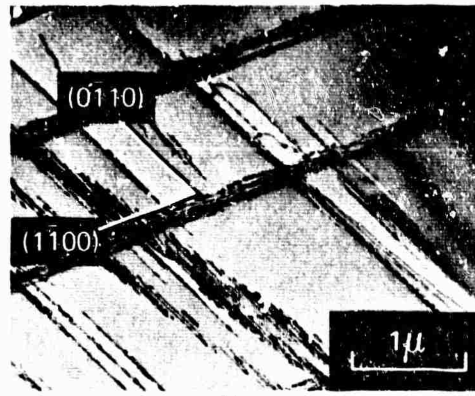
also show that changes in aluminum and oxygen content affect the relative values of the critical resolved shear stresses on the three possible slip systems: $(0001)\langle 11\bar{2}0\rangle$, $\{10\bar{1}1\}\langle 11\bar{2}0\rangle$, and $\{10\bar{1}0\}\langle 11\bar{2}0\rangle$. Generation of coplanar arrays may influence the SCC susceptibility of titanium alloys and, for this reason, the dislocation arrangements that occur in deformed Ti-8Al-1Mo-1V and Ti-4Al-3Mo-1V have been studied using thin-foil electron microscopy.

Levine[5] shows that for pure titanium (<50 parts/million oxygen) at room temperature the CRSS (critical resolved shear stress) on the $(0001)\langle 11\bar{2}0\rangle$ slip system is 3 to 4 times that on the $\{10\bar{1}0\}\langle 11\bar{2}0\rangle$ system. Churchman[6] has shown that increasing the oxygen content raises the CRSS on the three possible slip systems but produces greater changes on the $\{10\bar{1}\}\langle 11\bar{2}0\rangle$ and $\{10\bar{1}1\}\langle 11\bar{2}0\rangle$ systems, resulting in approximately equal CRSS on all three systems of 0.1 to 0.2 weight percent oxygen.

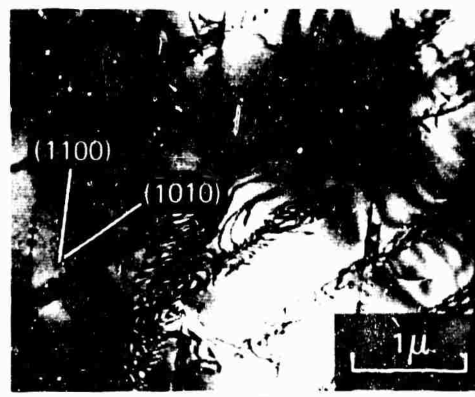
Recently, Blackburn [7] showed that aluminum contents greater than 5 weight percent promote preferential slip on $\{10\bar{1}0\}\langle 11\bar{2}0\rangle$ slip systems in binary titanium-aluminum alloys, resulting in dislocation pileup at grain boundaries and at points of slip-band intersection. Although the commercial alloys used in the present study contain 0.1 to 0.2 weight percent oxygen as compared to 0.02 to 0.04 weight percent oxygen in the alloys used by Blackburn, it was shown by thin-foil electron microscopy that aluminum has qualitatively the same effect on slip mode in these alloys as in the purer alloys. Figure 5 shows the dislocation arrangements in Ti-4Al-3Mo-1V, Ti-6Al-4V, and Ti-8Al-1Mo-1V after a 3- to 4-percent deformation. Dense tangles occur in Ti-8Al-3Mo-1V with no apparent tendency toward coplanar dislocation arrangements. Ti-6Al-4V and Ti-8Al-1Mo-1V exhibit coplanar dislocation arrangements. The principal difference is the reduced thickness of the individual slip bands in Ti-8Al-1Mo-1V.



(a) Ti-4Al-3Mo-1V, [1$\bar{2}$10] Zone Normal



(b) Ti-6Al-4V, [$\bar{1}$123] Zone Normal



(c) Ti-8Al-1Mo-1V, [$\bar{1}\bar{1}$23] Zone Normal

Fig. 5 Dislocation Arrangements After 3- to 4-percent Deformation

An expression for the normal stress at a dislocation pileup was derived by Stroh[8] who presented the following equation and Fig. 6:

$$\sigma = \sigma_s \left(\frac{L}{r}\right)^{1/2} f(\theta) \quad \text{where } (r << L)$$

where

σ = normal stress across OP at diameter r from the head of the pileup O.
L = length occupied by the dislocations in the slip plane, and
$f(\theta)$ = a factor depending on the orientation of OP.

Recalling also the shear stress σ_s,

$$\sigma_s = \frac{Gnb}{L\pi\,(1-\nu)}$$

where

G = shear modulus,
n = number of dislocations in the pileup, and
ν = Poisson's ratio,

it can be seen that σ increases linearly with n.

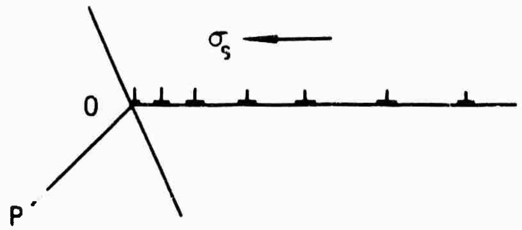


Fig. 6 Dislocation Pileup at a Barrier (from Ref. 8)

Considering the dislocation arrangements shown in Fig. 5, it is apparent that n is much larger for Ti-8Al-1Mo-1V and Ti-6Al-4V, since these alloys exhibit coplanar dislocation arrays and since, for a given strain, the dislocations are contained in fewer slip planes of principally the $\{10\bar{1}0\}$ -type. Thus, occurrence of cleavage failure in these alloys can be qualitatively explained if a minimum or threshold normal stress is required to promote SCC, since, by the previous arguments, the normal stresses in alloys that exhibit coplanar dislocation arrays will be greater. Conversely, ductile failure of Ti-4Al-3Mo-1V is indicative of normal stresses less than the threshold value. Such a threshold stress value does exist for titanium alloys since specimens loaded to less than a critical stress intensity[1] can sustain this load for prolonged times (more than 360 min) without failing. (2)

McEvily and Johnston (9) have pointed out that slip bands of reduced thickness impinge on a smaller grain-boundary area resulting in a reduced probability of activating dislocation sources in the adjacent grain, causing larger stress concentrations before relaxation by slip can occur. Since the principal difference between dislocation arrangements in Ti-6Al-4V and Ti-8Al-1Mo-1V appears to be the thickness of individual slip bands, then the increased susceptibility of Ti-8Al-1Mo-1V can be qualitatively explained in terms of increased difficulty in propagation of slip across a grain boundary.

Summary and Conclusions

1. The SCC of two susceptible titanium alloys has been shown to occur by cleavage of the alpha phase and by ductile rupture of the beta phase.

2. Evidence has been presented to show that beta or transformed beta acts as a crack arrestor.

3. Coplanar dislocation arrangements have been shown to exist in Ti-6Al-4V and Ti-8Al-1Mo-1V but not in Ti-4Al-3Mo-1V. Variations in susceptibility have qualitatively been shown to correlate with the development of such dislocation arrangements.

4. Generation of large stresses normal to the fracture plane has been shown to be associated with coplanar dislocation arrangements, and these stresses are thought to promote SCC.

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[1]The term "stress intensity" as used here is the fracture mechanics parameter. See Ref. 10 for an example.

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