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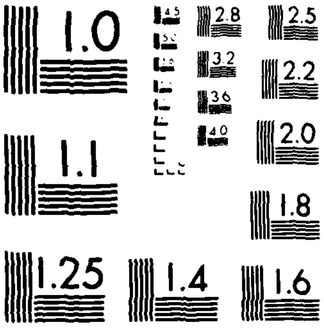

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THE INFLUENCE OF HYDROGEN ON THE MULTIAXIAL FRACTURE BEHAVIOR OF TITANIUM ALLOY SHEETS

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The influence of hydrogen on the deformation and fracture of three Ti alloys in the form of sheet has been examined as a function of the state of stress. Unnotched sheet specimmens of α-phase Ti (-60, 630, and 980 ppm H), β-phase Ti-30V (-40 and 2000 ppm H), and α-β Ti-6Al-4V (-30, 240, and 500 ppm H) have been investigated over deformation paths ranging from uniaxial to equibiaxial tension. Based on the measurements of the local fracture strains, the α-Ti data show a decrease in ductility with increasing hydrogen content

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as the degree of biaxiality of the tensile strain increases. Thus hydrogen embrittlement of Ti sheet is most severe under equibiaxial straining. In contrast, there is no pronounced effect of stress state on the localized necking and fracture behavior of either the Ti-30V or the Ti-6Al-4V at any of the hydrogen levels/microstructures examined. The results indicate that, even in equibiaxial tension, Ti alloys in the form of sheet are immune to hydrogen embrittlement if hydrides do not form. If hydrides are present, the embrittlement is most pronounced under those state of stress (e.g., plane strain and equibiaxial tension) which are characterized by large normal stresses.

The Influence of Hydrogen on the Multiaxial Fracture Behavior of Titanium Alloy Sheets

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Introduction

The susceptibility of titanium alloys to hydrogen embrittlement is well known (for reviews, see refs. 1 and 2). It is also well known that, at low hydrogen contents, such embrittlement often occurs only under conditions of a highly triaxial state of stress, such as near a crack-tip or notch. However, sheet and tubing components deform under conditions of plane stress, and triaxial states of stress are rarely present. Nevertheless, sheet and tubing products are also frequently stressed in service under multiaxial states of stress in hydrogen-containing environments, and sheet products containing hydrogen are often formed under multiaxial deformation conditions. Thus the potential embrittlement of thin-wall Ti alloy components under multiaxial stress states remains an important concern.

Previous studies have established that smooth bar specimens deformed in uniaxial tension appear relatively immune to embrittlement effects up to: ~400 wt ppm H in CP Ti [3-5], ~4000 ppm in single phase β alloys [1], and ~600 ppm in Ti-6Al-4V [1,6]. Burst pressure and preliminary fracture strain data have also shown that the hydrogen embrittlement of plastically anisotropic Ti sheet is sensitive to stress state, being much more severe in equibiaxial tension than in uniaxial tension [7,8]. The purpose of this study is to examine the influence of hydrogen on the multiaxial deformation and fracture behavior of not only α-phase commercially pure (CP) Ti, but also β-phase Ti-30V and the α-β alloy Ti-6Al-4V, all in form of sheet. Both the forming limit behavior (appropriate to stretch forming) and the fracture behavior of sheet specimens have been determined over a range of deformation paths from uniaxial to equibiaxial tension. Observations of the Ti-V and Ti-6Al-4V alloy behavior are confined to hydrogen levels at which embrittlement is not present in uniaxial tension. A more complete analysis of the hydrogen embrittlement of CP Ti sheets under multiaxial states of stress will be presented elsewhere [9].

Experimental Procedure

Both the CP Ti and Ti-6Al-4V were in the form of sheet (0.75 mm and 0.66 mm thick, respectively) kindly supplied by TIMET Corp.; the Ti-30V was cold rolled to 1.0 mm thick sheet.

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After a specimen preparation procedure which included recrystallization, the CP Ti sheet had a grain size of 0.015 mm and contained 1460 wt ppm oxygen; the solid solution Ti-30V alloy had a grain size of 0.034 mm and contained 1400 ppm oxygen; and the Ti-6Al-4V, which was in the mill-annealed condition characterized by an equiaxed α and (~27%) β-phase microstructure, had 1500 ppm oxygen.

Hydrogen was thermally charged into the test specimens using a modified Sieverts apparatus. Charging was performed at 973K. The hydrogen contents of the test specimens were: CP Ti-60, -630, or -980 wt ppm H, (Ti-30V)-40 or -2000 wt ppm H, and (Ti-6Al-4V)-30, -250, or -500 wt ppm H. After charging, the specimens were cooled at a rate of ~75°C/min; such a cooling rate should be sufficient to retain a solid solution in the Ti-30V alloy [10,11]. Optical and scanning electron microscopy showed that only the Ti-630H and Ti-980H sheets contained hydrides; these were plate-like and occurred both trans- and intergranularly (see ref. 9 for details).

The mechanical tests were based on the procedures used by Hecker for determining forming limit diagrams [12,13] and subsequent modifications developed by Bourcier and Koss for hydrogen embrittlement studies [8]. All tests were performed at room temperature at an average effective (engineering) strain-rate of $2\times10^{-3}\,s^{-1}$ for the CP Ti, $10^{-4}\,s^{-1}$ for the Ti-6Al-4V, and $10^{-5}\,s^{-1}$ for the Ti-30V sheets. Punch stretching utilized a 50 mm hemispherical punch. The major and minor principal strain in the plane of the sheet (ε_1 and ε_2, respectively) were measured from photogrids consisting of either 1.27 mm contacting circles or 0.50 mm squares. Both limit strains and local fracture strains were determined directly from grid dimensions.

Experimental Results and Discussion

The behavior in uniaxial tension of the three alloys at all levels of hydrogen is nearly independent of hydrogen content. In all cases, the yield and tensile strengths remain unaffected by hydrogen. Increasing hydrogen content does cause a very small decrease in the strain hardening exponent n obtained by fitting the stress σ-strain ε curve to a $\sigma=k\varepsilon^n$ relationship. For example, in the CP Ti the value of n decreases from 0.11 to 0.09 for uniaxial tests transverse to the rolling direction (TD) and from 0.16 to 0.13 in the rolling direction (RD). In the Ti-30V alloy, n decreases from 0.08 to 0.06 while the strain-rate hardening exponent m ($m = d\ln\sigma/d\ln\dot{\varepsilon}$) increases from 0.013 to 0.017 as the hydrogen level increases from 40 to 2000 ppm H. In the Ti-6Al-4V, there was no influence of H on either n or m for up to 500 ppm H.

Both the Ti and the Ti-6Al-4V possess crystallographic textures and exhibit a strong degree of plastic anisotropy. Using the parameters R and P to describe plastic anisotropy (R and P are $\varepsilon_{width}/\varepsilon_{thickness}$ in tensile tests of specimens in either the RD or TD directions, respectively), the following data were obtained: (1) for CP Ti, R≈2 and P≈5; and (2) for Ti-6Al-4V, R≈P≈4. The β-phase alloy is plastically isotropic [R=P≅1].

The influence of hydrogen on both the forming limit and the fracture behavior of β-phase Ti-30V alloy is shown in Fig. 1. The forming limit curves are based on the major and minor strains in the plane of the sheet (ε_1 and ε_2) at the onset of localized necking [12]. The fracture limit curves are determined from local measurements of ε_1 and ε_2 based on fracture grid elements at the fracture surfaces. The obvious conclusion from the data in Fig. 1 is that 2000 wt ppm H, which corresponds to 9 at. %, has no effect on either the forming limit or fracture behavior of the β-phase Ti-30V alloy. Fractography shows dimpled fracture surfaces characteristic of ductile fracture, the fracture surface appearance also being unaffected by H. Optical microscopy shows no hydrides are present even after straining.

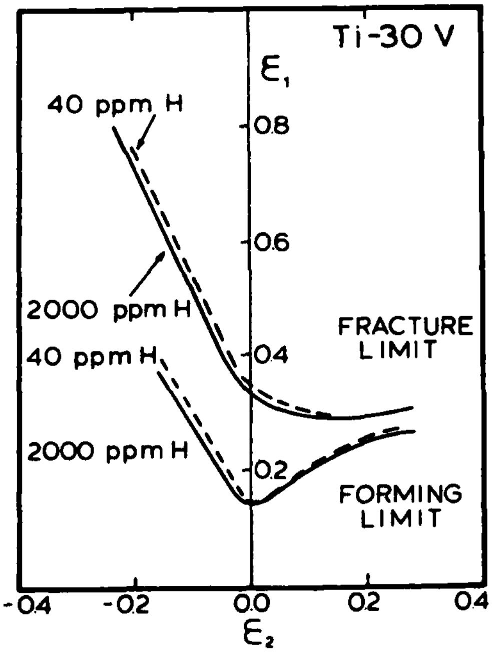

Fig. 1: A comparison of the forming limit and fracture limit diagrams for Ti-30V sheet containing either 40 or 2000 wt. ppm hydrogen.

While the above fracture behavior is consistent with previous burst pressure data on the β Ti alloys [14], the limit strain observations (Fig. 1a) are nonetheless surprising in view of the well-established sensitivity of localized necking of sheets to small changes of the strain hardening exponent n and particularly the strain-rate hardening exponent m [13,15]. In other β Ti alloys, disc pressure tests indicate embrittling effects only at hydrogen contents ≳2000 ppm and only if α-phase is present [14].

The absence of embrittling effects in either the limit or fracture strain measurements can be rationalized on the basis of the large degree of solid solubility of H in bcc Ti alloys, the apparently isotropic character of the dilation of the bcc lattice by the H atoms [1], and the high rate of diffusion of H in the bcc lattice. As a result there is very little influence of H on dislocation motion, and no hydrides are present to act as void nuclei. The minor effects which do occur tend to compete against each other in influencing macroscopic flow behavior. For example, the small decrease in strain hardening exponent n from 0.08 to 0.06 tends to be compensated by a small increase in strain-rate hardening exponent m for 0.013 to 0.017 as H is increased from 40 to 2000 ppm. Thus, H has little or no net effect on limit strain or fracture behavior in Fig. 1.

The behavior of the Ti-30V alloy suggests that if there is sufficient β-phase present to act as a solid solution trap for the H present in an α-β Ti alloy, then no embrittling effects will occur provided that the H does not accumulate locally and form hydrides. As shown in Fig. 2, this is the case even in equibiaxial tension of the α-β Ti alloy, Ti-6Al-4V. The presence of the β-phase, the absence of locally large triaxial stresses and the relatively uniform character of the stress state in the present sheet test are conditions which favor macroscopically uniform hydrogen concentrations (albeit microscopically partitioned to the β-phase) in which hydrogen remains in solution (mostly in the β-phase). Thus,

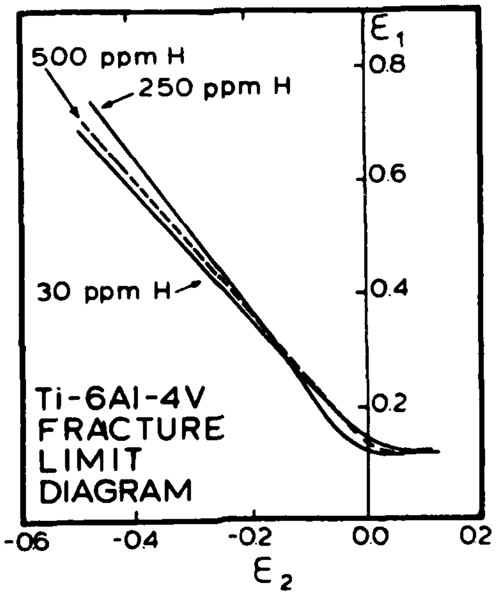

Fig. 2: Fracture limit diagrams for Ti-6Al-4V sheet containing 30, 250, or 500 ppm hydrogen.

hydrogen embrittlement does not occur even in equibiaxial tension of a sheet material whose plastic anisotropy dictates that large normal stresses are required for flow to occur. Consistent with the above, no hydrides were observed optically, even after straining, in the Ti-6Al-4V alloy containing 500 ppm H. The appearance of the dimpled fracture surfaces were also unaffected by H.

In contrast to the β and α-β Ti alloys discussed above, the α-phase CP Ti shows a decrease in ductility with increasing hydrogen content as the degree of biaxiality of the tensile stresses increases. As shown in Fig. 3, the hydrogen embrittlement of this plastically anisotropic α-phase alloy depends on deformation path and is the most severe under equibiaxial tension. A strong degree of "biaxial" embrittlement occurs even though hydrogen has little effect on the yielding or flow behavior in both uniaxial and equibiaxial tension [16].

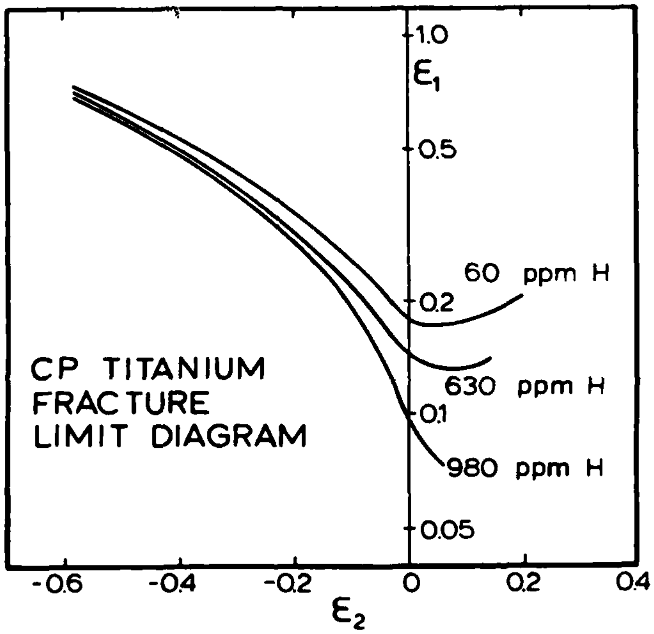

Fig. 3: Fracture limit diagrams for CP Ti sheet containing 60, 630, or 980 ppm hydrogen.

Unlike the β Ti-30V and the α-β Ti-6Al-4V, the α-phase CP Ti contains hydrides and, as shown in Fig. 4, these fracture as a result of the matrix undergoing plastic deformation. The surprising aspect is that, although the fractured hydrides clearly form voids (see Fig. 4), it is only in plane strain and especially in equibiaxial tension that hydride fracture results in large losses of ductility. In uniaxial tension, the Ti-60H sheet (which contains almost no hydrides even in highly strained material) shows similar ductility to the Ti-980H material which has a high density of large (~25μm x 2.5μm) hydrides which begin forming voids at strains

Fig. 4: An electron micrograph showing strain-induced hydride fracture and void formation in CP Ti.

much less than the fracture strain. The influence of stress state on the hydrogen embrittlement of CP Ti sheet is examined in detail elsewhere; that study concludes that there are two factors primarily responsible [9]. First, data show that equibiaxial and plane strain deformation are especially effective in fracturing hydrides and forming voids; this may be principally a result of the stress state due to the strong degree plastic anisotropy of the sheet; specifically, R and P >>1. Secondly, once the hydrides fracture and voids form, the plasticity associated with the failure process under multiaxial deformation is such that void link-up occurs via the onset of a local shear instability process; the data show that this occurs at a much lower void density in plain strain and equibiaxial deformation.

Summary

The influence of hydrogen on the deformation and fracture of three Ti alloys in the form of sheet has been examined as a function of the state of stress. Unnotched sheet specimens of α-phase Ti (- 60, 630, and 980 ppm H), β-phase Ti-30V (- 40 and 2000 ppm H), and α-β Ti-6Al-4V (- 30, 240, and 500 ppm H) have been investigated over deformation paths ranging from uniaxial to equibiaxial tension. Based on the measurements of the local fracture strains, the α-Ti data show a decrease in ductility with increasing hydrogen content as the degree of biaxiality of the tensile strain increases. Thus, hydrogen embrittlement of Ti sheet is most severe under equibiaxial straining. In contrast, there is no pronounced effect of stress state on the localized necking and fracture behavior of either the Ti-30V or the Ti-6Al-4V at any of the hydrogen levels/microstructures examined. The results indicate that, even in equibiaxial tension, Ti alloys in the form of sheet are immune to hydrogen embrittlement if hydrides do not form. If hydrides are present, ductile fracture may be accelerated by void initiation due to the strain-induced fracture of hydrides. As analyzed elsewhere [9], the sensitivity of

hydrogen embrittlement in CP Ti to the state of stress is associated with plane strain and equibiaxial deformation being (a) especially effective in fracturing hydrides and forming voids in this plastically anisotropic sheet material and (b) very susceptible to shear instability triggered by void formation.

Acknowledgements

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