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HIGH TEMPERATURE DUCTILITY LOSS IN TITANIUM ALLOYS - A REVIEW

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It is well known that two phase titanium alloy systems suffer from an abrupt drop in ductility at elevated temperatures in the range of 1000 to 1150°K. This loss of ductility is manifested by easy decohesion of polycrystalline aggregates along the grain boundaries of the high temperature beta phase. If the alloy is in a state of tensile stress at the aforementioned temperatures, cracks initiate at the grain boundaries and propagate readily through the alloy, leading to premature failure. This phenomenon is a cause of major concern in titanium alloy fabrication and welding. Several mechanisms have been proposed to explain high temperature crack nucleation and growth along the boundaries. A critical review of the phenomenon and possible mechanisms responsible for the observed behavior will be discussed.

INTRODUCTION

Several near α and α - β titanium alloys have been found to have poor hot workability under certain processing conditions. This poor workability, and sub-solidus weld cracking, result from a ductility loss in the temperature range of the β to α transformation. This phenomenon has been studied and found to vary with alloy content, specimen thermal history, and a number of microstructural characteristics. This paper will review and summarize the results of several investigations of this high temperature ductility drop.

I. Background

The high temperature ductility loss observed in α - β titanium alloys is illustrated in Figure 1. This data, from Lewis and Caplan (1), plots %RA for tensile tests performed both on heating and on cooling during a simulated GTA weld thermal cycle. The key features of the thermal cycle were a peak temperature above

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Figure 1. High temperature ductility of several titanium alloys, some showing anomalous on cooling ductility loss (1).

the β transus and a moderate cooling rate, approximately 10°C/s between 1000°C and 750°C . When tested on heating, the ductility of all of the alloys increased smoothly with temperature. On cooling however, the alloys had ductility minimums between 750°C and 850°C . No minimum was seen in unalloyed titanium. The specimens tested within the low ductility region were macroscopically brittle, failing at or near prior β grain boundaries, but microscopically ductile, failing by microvoid coalescence, Figure 2 (1). These characteristics of the ductility loss, the alloy and temperature dependence and the failure characteristics, are consistent throughout the investigation of the phenomena.

Figure 2. Fracture surfaces of Ti-6211 tensile tested on cooling at 821°C (1). Sample is β processed material, 10% RA.

Lewis and Caplan (1) found that the ductility loss occurred in near α and α - β alloys, alloys which typically retain 5% to 20% β at room temperature. The microstructures in these alloys will be controlled by alloy content, cooling rate, and microstructure prior to cooling, specifically the mix of α and β , and the size and shape of the β grains (2-5).

Under equilibrium conditions, the bcc β phase transforms on cooling to hcp α by nucleation and growth. When rapidly cooled however, either hexagonal α' or orthorhombic α'' martensite can form. The literature shows that the high temperature ductility loss is associated with a temperature range just below the β to α transformation finish, and β processed material with equiaxed prior β grains. Further, it has been observed that the loss is limited to alloy/thermal cycle combinations which result in room temperature microstructures containing specific mixes of α , β , and α' . The investigations which gave rise to these conclusions will be reviewed, followed by a discussion of the proposed models for the behavior.

II. Review of Hot Ductility Dip Investigations and Results

Two factors contribute to an alloy's susceptibility to the high temperature ductility loss, alloy content and prior thermomechanical history. These will be discussed individually and the results will then be summarized.

Figure 3. Ti-6211 HAZ showing microstructure prone to weld cracking (8).

several titanium alloys and found that most had a ductility minimum between 600°C and 800°C, Figure 4. In general, his results were similar to those cited above, Ti-6Al having a ductility loss, but the most severe losses being associated with the presence of 5-6 wt% aluminum and ≥ 2 wt% β stabilizer content. In the case of Ushkov, the β stabilizer element studied was Sn. Ushkov found a slight ductility loss in Ti-8Mn, and reduced ductility, but no definite minimum, in Ti-6Al-4Zr. Ushkov also showed a microstructural effect, in that all of the coarse-grained specimens had a severe loss of ductility, but in fine grained material, the severity of the loss was a distinct function of alloy content.

Suzuki, et al. (11) investigated the hot workability of a number of titanium alloys, Figure 5. Included in their study were as-cast Ti-5Al-2.5Sn, Ti-64, and Ti-6Al-6V-2Sn (Ti-662), α - β processed commercial purity (CP) titanium and Ti-64, and β processed Ti-64, 64ELI, and Ti-6Al. The results of this study confirm that the phenomena is primarily seen in α - β alloys, no loss being observed in the CP titanium or β alloys. The extent of the ductility loss in Ti-64 was affected by microstructural condition and interstitial content. The loss was most severe for the

A. Effect of Alloy Content

Lewis and Caplan (1) showed that susceptibility to the ductility loss was a function of alloy content, and that Ti-6211 was particularly susceptible. Similar results were obtained in the weldability studies of Hayduk, Damkroger, and Edwards (6-9). They found that Ti-6211 was susceptible to sub-solidus cracking in GTA and GMA welding, but Ti-64 and Zr modified Ti-64 variants were not. Both the microstructure, shown in Figure 3, and fracture surfaces associated with the Ti-6211 weld cracks were similar to those shown by Lewis and Caplan, suggesting that the same mechanism is occurring.

Ushkov (10) studied the hot ductility of

Figure 4. Ductility versus temperature results of Ushkov (10), showing ductility losses in (a) fine grained and (b) coarse grained material.

Figure 5. Ductility and strength of several titanium alloys. Data of Suzuki, et. al. (11).

as-cast and β processed standard materials, and far less severe in the α - β and β processed ELI materials. Ti-5Al-2.5Sn and Ti-662 had ductility losses in the as-cast condition, but were not tested in the other conditions.

The Suzuki (11) study also addresses the effect of microstructure, which will be discussed in detail in the next section. None of the alloys showed a ductility loss when tested in the α - β processed condition. The ductility loss was alloy-dependent in the β -processed material, and uniformly severe in the as-cast material. In the as-cast specimens however, only moderate gains in ductility are realized by further cooling of the material. It is possible that the as-cast specimens may simply have an overall reduced ductility, rather than the specific minimum seen in β processed material.

Damkroger et al. studied the behavior of several titanium alloys, including Ti-6211, Ti-64, and ternary alloys of Ti-6Al plus 2, 4, and 6% Nb, Mo, or V (13,14). CCT curves were developed for these alloys and on-cooling tensile tests were performed over a range of temperatures across the

Figure 6. Reduction in area versus test temperature for Ti-6Al-2V-2Nb tested at several cooling rates (13).

β to α transformation. All of the alloys were found to have some high temperature ductility loss, with alloys containing Nb being particularly susceptible. The data, and example of which is shown in Figure 6, shows the ductility loss as a function of alloy content and cooling rate, and finally of microstructure, as will be discussed in the following section.

Several studies have examined the

effects of trace or interstitial elements on the ductility loss. Lewis and Caplan (1) found an increased concentration of sulfur on the fracture surfaces of specimens which had been thermally cycled to above the β transus temperature, cooled, and tested at 77°K. They postulated that the sulfur could be weakening an interface, leading to the failures. Inouye and David (15,16), and Lewis, et. al. (17) noted the detrimental effect of boron additions on Ti-6211 castings and weld metal. Inouye and David found boron rich particles on the fracture surfaces of specimens thermally cycled to above the β transus, cooled, and broken at room temperature. In these studies, the ductility loss persisted to room temperature, rather than existing only over a specific temperature regime.

The effect of interstitial elements on the weld cracking of Ti-6211 was studied by Hayduk (9) who added oxygen and nitrogen to the fusion zone via specially prepared cover gases. Both fusion zone and heat-affected zone cracking were found to increase with oxygen and nitrogen content up to approximately 0.15% nitrogen or 0.2% oxygen. At higher interstitial levels, weld cracking fell to essentially zero but this was attributed to stabilizing the α phase, replacing the coarse, lenticular α' with Widmanstätten $\alpha + \beta$. Suzuki, et. al (11) also found that interstitial content affected the ductility loss, the loss being less severe for ELI (extra-low-interstitial) than for standard material. Nordin (18) found that the weld cracking susceptibility of Ti-6211 was reduced by yttrium additions, and Lewis et. al. (19), found the ductility loss reduced by yttria and oxygen additions. In both cases, the lenticular α' was replaced.

B. Effect of Prior Thermomechanical History and Microstructure

Lewis and Caplan (1) observed that the ductility loss was a function of alloy content, required prior super-transus thermal exposure, and occurred near the β to α transformation finish. These observations are consistent throughout the investigations of the high temperature ductility loss and associate the phenomena with specific microstructures. In other work, Lewis (20,21) characterized the hot ductility dip in Ti-6211 with respect to cooling rate and microstructure. These results are shown in Figure 7, a superposition of ductility regimes onto the CCT diagram of Gordine (22). The hot ductility behavior is clearly tied to the β to α

Figure 7. Ductility of Ti-6211 as tested by Lewis (20), superimposed on the CCT diagram of Gordine (22). Minimum ductility lies just below the β to α transformation finish.

Figure 8. Locus of vareststraint cracking points for several welding cycles, superimposed onto the CCT diagram of Damkroger (9). Data for Ti-6211.

transformation, with the minimum ductility falling just below the finish. Within the β to α transformation, Lewis found that the ductility decreased with both decreasing temperature and increasing cooling rate, corresponding to increasing amounts of α .

The sub-solidus weld cracking behavior of Ti-6211 shows similar trends (7-9). Figure 8 shows the locus of cracking time/temperature curves superimposed on the CCT diagram of Damkroger. As in the Lewis study, the locus of points which define the vareststraint cracking fall approximately on the β to α transformation finish. The microstructures associated with the weld cracking, and ductility loss, were shown in Figure 3. These consisted of a grain boundary α film, Widmanstatten $\alpha+\beta$, and a lenticular microconstituent identified by Lewis (23) as α' martensite, Figure 9. This microconstituent, usually found in rapidly cooled niobium-bearing alloys, seems to be associated with severe ductility losses (13). Lewis (21) showed that the coarse, lenticular α' was functioning as a void nucleation site. Figure 9 shows a number of small voids associated with individual α' platelets.

In comparison, the microstructures seen in Ti-64, Figure 10, are Widmanstatten $\alpha+\beta$, replaced by fine α' at the highest cooling rates. Ti-64 welds, which did not crack,

Figure 9. Microstructure in Ti-6211 showing coarse, lenticular α' (19).

Figure 10. Fine, α' microstructure in Ti-64 GTA weld fusion zone (12).

didn't include the coarse, lenticular α' . In Ti-6211 GMA welds, locations where microstructures which contained the coarse lenticular α' , but more highly stressed locations which didn't crack, did not. Thus, the ductility loss seems to be associated with the β to α transformation finish because of the formation of but a particular transformed microstructure.

These results were amplified in the later work of Damkroger (13,14) in which continuous-cooling-ductility (CCD) maps and CCT curves were superimposed for several α - β alloys. Examples of these maps are shown in Figure 11. For the most part, the alloys behaved in a similar fashion, shown in Figure 11a. At cooling rates which resulted in the creation of a coarse lamellar microstructure, the ductility loss was found to be particularly severe. Lower cooling rates produced an equiaxed microstructure, and higher cooling rates, fine basketweave $\alpha + \beta$, or α' , neither of which was susceptible to a ductility loss. The result was the creation of a low ductility window in time/temperature space. In the Nb bearing alloys however, intermediate cooling rates resulted in the formation of the coarse lenticular α' and the ductility loss was severe. In the case of Ti-6Al-2Nb, a window of moderate ductility remained, Figure 11b. In Ti-6211, the kinetic regimes for coarse colonies and the lenticular α' overlap, leaving no window of reasonable ductility.

Damkroger (14) showed that the kinetic behavior of the ductility loss and the effect of alloy content arise from the β to α transformation kinetics and the production of a susceptible microstructure - either coarse, lamellar $\alpha + \beta$, or in Nb containing alloys, lenticular α' . For the lamellar structures, he went on to show a general correlation between the severity of the ductility loss and the amount and size

Figure 11. Time-temperature-ductility maps and CCT curves for (a) Ti-64 and (b) Ti-6Al-4Nb (13).

of the lamellar colony (vs basketweave) microconstituent. As a result, the sensitivity to and temperature range of an alloy's ductility loss could be predicted based on a model of the transformation kinetics incorporating composition and cooling rate (24).

Stark and his co-workers investigated several microstructural aspects of the ductility loss in Ti-6211 (25-28). They studied the effects of prior β grain size and shape, morphology of the β decomposition product, and the intergranular structure. The effect of β grain size and shape were first evaluated. It was concluded that if the β grains were equiaxed, any transformation structure other than equiaxed α was susceptible to a high temperature ductility loss. It was also found that this behavior was not affected by the prior β grain size. The effect of α phase morphology was investigated by testing samples with different α phase morphologies. Equiaxed α microstructures were found to have high ductility at all test temperatures, implying that Widmanstätten or martensite plus grain boundary α structures are required to cause reduced hot ductility. The effect of slip path length was then investigated by quenching partially transformed (lamellar α + β) specimens to transform the interlath β to α' , which limited the slip path length to the width of an individual Widmanstätten α plate or the thickness of a grain boundary α film. The quenched samples were found to behave identically to fully transformed samples, suggesting that the loss of ductility is not affected by the slip path length within the transformed β grains.

The Starke studies clearly demonstrate a number of microstructural aspects of the ductility loss phenomena which are in agreement with those discussed previously. The most significant is that the ductility loss requires the creation of a transformed β microstructure in equiaxed prior β grains. However, Starke found no effect of either the transformation product present or the scale of the prior β grains or transformation product. As shown by Damkroger (13,14) however, these results apply only to Ti-6211 and to a certain extent, other Nb containing alloys. In other alloy systems, both the transformation product and its scale will affect the ductility loss. As an example, Blackburn (29) and Ushkov (10) found that when Ti-6Al specimens were modified to produce either a martensitic structure or a refinement of the α grain size, the 800°C ductility of the material was improved.

III. Mechanisms and Models

Several studies have proposed mechanisms to explain the ductility loss and low ductility failures. One interesting aspect of the phenomena is the lack of any clear discontinuity in the macroscopic stress/strain behavior. Krishnamohanrao, Kutumbarao, and Rao combined data from the literature and their own work to create failure maps for several titanium alloys (30,31). These fracture mechanism

maps plot normalized tensile stress (σ/E) versus homologous temperature (T/T_m) and identify regions where specific failure modes are dominant. The map for Ti-6Al-4V, Figure 12, shows no anomalies in the temperature range of the high

Figure 12. Failure map for Ti-64 (31).

temperature ductility loss. In Ti-64, the hot ductility loss region as shown by Damkroger (13) falls largely within the dynamic fracture and rupture regimes. This is in direct contrast with the appearance of the fracture surfaces which look more like those in the intergranular creep rupture region. In this case, the discrepancy may be due to significant differences in the microstructural and test conditions. Still, the lack of any indication of the low ductility regime is puzzling.

The lack of any inconsistencies in the macroscopic stress/strain behavior was observed by Damkroger (14) and Suzuki (11) as well. In both of these

studies, the maximum strength of the alloys tested was found to be a smooth function of test temperature, with no indication of the ductility minimum. Similarly, true stress-true strain curves show a progression from a balance between recovery and deformation at high temperatures to work-hardening behavior at lower temperatures. The only discontinuities seen were associated with the β transus for the alloy tested. Suzuki et. al (11) studied the temperature and strain rate effects on high temperature deformation of Ti-662, an alloy which did show a definite ductility dip. They derived activation energies for deformation at temperatures above and below the β transus of 318 kJ/mole and 254 kJ/mole, respectively. These values compare well with those from creep studies of alloys which show no ductility dip. These results suggest that the ductility minimum doesn't result from a discontinuity in deformation mechanics, but from microstructural causes that prevent the material from accommodating strain.

Several researchers have proposed models for the ductility loss. One model, proposed by Starke et al (28), is that at 800°C, the α allotriomorph is significantly weaker than the adjacent β or transformed β . Slip concentrates in the grain boundary α and is blocked at a non-Burgers related α/β interface, causing stress concentration and microvoid nucleation. This mechanism is similar to that shown by Greenfield and Margolin (32) to occur at room temperature, but conflicts with the generally reported strengths of the α and β phases. Lewis (1), Fujii and Suzuki (12), and Vandecastelle, et al (33) all showed that at the critical temperatures, α was the stronger phase in Ti-6211 and Ti-64. Hayduk (9) and Ushkov (10) found a more severe ductility loss in material with high levels of interstitial elements, which segregate to and strengthen the α phase. A second shortcoming of the model is that

no clear relationship exists between the size of the α allotriomorph and the severity of the ductility loss (10,14).

A second model, based on the partitioning of trace solute elements, is supported by Lewis and Caplan (1) and Inouye and David (15,16). Lewis and Caplan proposed that the failures were due to the presence of a sulfur-rich layer at the grain boundary allotriomorph/Widmanstatten side plate interface. When the material is strained, slip is blocked by the "effective interface", leading to microvoid nucleation (1). Inouye and David proposed a similar model, with boron diffusion to the β grain boundaries being the controlling mechanism, causing the embrittlement to behave like a diffusion-controlled process. They suggested that the enhanced diffusivity at high temperature and in the bcc lattice were leading to the association of the phenomenon with the super-transus thermal cycles, but that any thermal cycle with sufficient time at temperature would result in embrittlement (15). Additionally, Inouye and David proposed that the embrittlement of welds was due to the precipitation of TiB particles during cooling through the β region.

A strength of the model proposed by Lewis and Caplan is that it mates well with the mechanism of the β to α transformation as documented by Rath and Imam (34). They showed that the formation of Widmanstatten side plates does not result from a breakdown of an advancing allotriomorph/ β interface, as was generally thought (35). Rather, the side plates form on the side of the grain boundary allotriomorph opposite that which grows into the β grain. However, the solute partitioning model suffers from inconsistencies when compared with the entire body of data. For one, the low ductility failures occur in material cooled too slowly to support the idea of non-equilibrium partitioning of an interstitial element (14). Conversely, the creation of an impurity-rich interface by an equilibrium, diffusion-controlled process contradicts the results of Starke et al (25), who found increased ductility in samples held at 800°C for 30 minutes prior to testing. Further, intergranular fracture associated with grain-boundary embrittlement results in smooth fracture surfaces, rather than the dimpled fracture surfaces seen in the low ductility failures, and unlike grain boundary embrittlement, testing within a specific temperature range is required to achieve the high temperature ductility loss (25).

Suzuki et al. (12) found that the failures occur not in the grain boundary α film, but in an adjacent β film at the root of the Widmanstatten side plates, formed by an α to β retransformation in a locally enriched region. They attributed the failure to strain accumulation in the β film and proposed the relationship:

$$\epsilon_c \propto 1/L+d \quad \text{where: } \epsilon_c \text{ is the critical fracture strain}$$

L is the grain boundary α thickness
 d is the prior β grain size

This relationship is an expression of the slip path length/stress concentration mechanism that governs failure in materials prone to localized deformation, shown (36-40) to occur in α and α - β titanium alloys at room temperature and experimentally demonstrated by Terlinde and Gysler (38,39). At high temperature however, a clear prior β grain size dependence has not been demonstrated. Damkroger (14) showed that the severity of the ductility loss was instead related to the size and amount of lamellar Widmanstatten colonies, an observation that is more consistent with the critical temperatures, which are below the transformation

finish. However, several investigators have shown evidence of strain localization and planar slip in the transformed β adjacent to grain boundary α .

A key element in the Suzuki model is the existence of a continuous β film at temperatures between $0.5T_m$ and 50°K below the β transus. This was addressed by showing evidence that the material adjacent to the grain boundary allotriomorph undergoes a reverse transformation during a subtransus isothermal hold or a reheating cycle, giving rise to the continuous β film (12). The composition and thermal history effects result from the requirement of a grain boundary α plus transformed β microstructure, and the temperature dependence reflects a regime where the retransformation to β occurs only at the β stabilizer-rich zone adjacent to the allotriomorph. However, this model does not explain a ductility loss during an initial cooling from a supertransus thermal excursion. In fact, Suzuki and Fujii show "starting" microstructures as examples where no continuous β exists. These microstructures, which have no continuous β film, are in fact susceptible to the ductility loss.

Damkroger and Edwards (14) also proposed a model based on localized slip of a particular microconstituent, but that the microconstituent was a lamellar $\alpha + \beta$ colony structure. Room temperature studies of planar slip in titanium usually involve precipitation hardening alloys, where softening due to particle dissolution creates a preferred slip path (36-39). In this case, Damkroger and Edwards proposed that planar slip of the lamellar colonies occurred at high temperatures because of changes in the shape of the α phase lattice. X-ray diffraction studies (14) showed that the α phase c/a ratio for several alloys increased from approximately 1.595 to 1.610 as the temperature changed from room temperature to the ductility loss temperatures. This change was believed to enhance the tendency toward basal slip (41), resulting in gross planar deformation of the colonies. As with the models discussed previously, this model cannot adequately explain the entire body of data. In this case the questions are the behavior of Nb bearing alloys containing the coarse, lenticular α' martensite, and the validity of the proposed slip character of the α phase (42,43).

Rath et. al (35,40) have proposed that the ductility loss is associated with stresses built up during the transformation. One mechanism proposed is the lattice contraction which occurs during the bcc β to hcp α transformation (40). Damkroger (14) investigated this model by measuring the lattice parameters of the α and β phases, as a function of temperature, for a wide range of model alloys and comparing the calculated contraction stresses to hot ductility behavior. The results showed a general consistency but a clear correlation could not be established. In a later study of the β to α transformation mechanism, Rath and Imam found that the lamellar $\alpha + \beta$ structure was not created by the breakdown of an advancing planar α/β interface, but from the growth of Widmanstätten side plates into the opposite β grain (34). Because a single α microconstituent cannot have the Burgers relationship with two randomly oriented β grains, the observed transformation mechanism would give rise to a significant microstructural discontinuity. Rath and Imam proposed that this discontinuity contributed sufficiently to the local stress state to result in the low ductility failures (34).

IV. Summary

Several general conclusions can be made regarding the high temperature investigations which have been conducted. The first is that the hot ductility dip and the occurrence of weld cracking seem to be different manifestations of the same

phenomenon, general loss of ductility in the temperature range of the β transus - 50°C to β transus - 250°C. The ductility loss seems to occur in nearly all α - β titanium alloys, if they are thermomechanically processed to an appropriate microstructural state. The severity of the ductility loss seems to be a function of alloy content and is most significant in alloys having approximately 5-6 wt% aluminum and 2-3 wt% β stabilizer elements. Combined with an observed cooling rate dependence, the critical criteria then becomes the creation of a coarse lamellar Widmanstatten microstructure, usually occurring with a grain boundary α allotriomorph, within equiaxed prior β grains. These microstructural criteria are relaxed in the case of Ti-6211, and to a lesser extent, in the case of Nb bearing ternary or quaternary alloys, where the ductility loss is particularly severe, regardless of microstructure.

A great deal of disagreement exists regarding the mechanism(s) responsible for the behavior. All of the models require two things - concentration of strain and an interface that blocks the passage of slip. The models differ in how they propose that the strain concentration be developed and, in the case of the solute partitioning model, what sort of interface is required to block slip. The ductility loss is particularly severe in the case of large prior β grains, and coarse Widmanstatten colonies, leading one group to suggest that deformation of the colonies may be the governing mechanism. However, other researchers have proposed that the ductility loss is actually tied to the deformation of the grain boundary α film itself, or to a retransformed β film adjacent to the allotriomorph. Still others believe that the phenomenon is tied to stresses or compositional gradients developed during the transformation itself. It is clear that each of the proposed models for the high temperature ductility loss is consistent within the bounds of the investigation upon which it is based. However, it is equally apparent that when the greater body of data is considered, inconsistencies can be found in each of the models.

VI. Conclusions

- 1) α - β titanium alloys are often susceptible to a severe reduction in ductility at temperatures just below the β to α transformation finish.
- 2) The ductility loss is associated with a transformed microstructure consisting of grain boundary α plus coarse Widmanstatten α + β in equiaxed prior β grains.
- 3) Alloy and cooling rate dependence of the ductility loss can be explained based on the creation of a susceptible microstructure.
- 4) Nb containing alloys are susceptible to a particularly severe ductility loss, and typically have microstructures containing a coarse, lenticular α' microconstituent, itself associated with microvoid nucleation.
- 5) Several models have been proposed for the phenomenon, all of which have limited applicability.

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