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1996 TMS Fall Meeting/Materials Week
October 6-10, 1996
Cincinnati, OH
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MECHANICAL RESPONSE AND MICROCRACK FORMATION IN A FINE-GRAINED DUPLEX TiAl AT DIFFERENT STRAIN RATES AND TEMPERATURES

Zhe Jin*, Carl Cady*, George T. Gray III*, and Young-Won Kim**

*Los Alamos National Laboratory, Materials Science and Technology Division, MS G755, Los Alamos, NM 87545, U.S.A.
**Universal Energy Systems Inc., Materials & Processes Division, 4401 Dayton-Xenia Road, Dayton, OH 45432, U.S.A.

Abstract

The compressive mechanical behavior of a fine-grained duplex gamma TiAl alloy was studied at strain rates of 0.001 sec\(^{-1}\) and 2000 sec\(^{-1}\) and temperatures from -196 °C to 1200 °C. The temperature dependence of the yield stress was found to depend on the strain rate: At the quasi-static strain rate, 0.001 sec\(^{-1}\), the yield stress decreases with increasing temperature with a plateau between 200 °C and 800 °C. At the high strain rate, 2000 sec\(^{-1}\), the yield stress exhibits a positive temperature dependence at temperatures above 600 °C. The strain hardening rate decreases dramatically with temperature in the very low and high temperature regions with a plateau occurring at the intermediate temperatures for both strain rates. As the strain rate increases the strain hardening rate plateaus were seen to extend to higher temperatures. The strain rate sensitivity was found to increase slightly with temperature (but less than 0.1) for strain rates above 0.001 sec\(^{-1}\). However, at a strain rate of 0.001 sec\(^{-1}\), there is a dramatic increase in the strain rate sensitivity with increasing temperature. In particular, at temperatures above 1100 °C, the rate sensitivity becomes much larger. Microcracks occurring in the grain interiors and at grain boundaries were observed at all strain rates and temperatures. The formation and distribution of microcracks were found to vary depending on the strain rate and temperature of deformation.
Introduction

Significant progress has been made in development of γ-TiAl alloys toward high temperature and high performance applications in the last decade [1-3]. Various γ-TiAl components for both gas-turbine engines and automotive engines have been identified and engine tests on some components have been performed successfully [4-6]. However, the low room-temperature remains an obstacle for a wide scale use of γ-TiAl alloys [4]. The advances in thermomechanical treatment on γ-TiAl alloys have led to the development of uniformly distributed fine-grained γ-TiAl alloys [1,3]. The mechanical properties of these fine-grained γ-TiAl alloys under quasi-static loading show a dramatic increase in their ductility to as much as 4% [3,7-9]. However, the mechanical properties of fine-grained γ-TiAl alloys at high strain rates have not been investigated to date. The purpose of this paper is to study the mechanical response of a fine-grained γ-TiAl alloy at high strain rates over a range of temperatures and compare the high-strain-rate properties with those at quasi-static strain rates. A detailed analysis of the deformation mechanisms and microcrack formation in a fine-grained γ-TiAl alloy at high and quasi-static strain rates and at temperatures ranging from -196 °C to 1200 °C will be presented.

Experimental Procedure

The material used in this study was a fine-grained duplex γ-TiAl alloy with a nominal composition of Ti-46.5Al-2Cr-3Nb-0.2W (at.%). A fine-grained duplex microstructure was obtained using a thermomechanical process followed by heat treating at 1280 °C/3hrs/FC to 800 °C/AC. Cylindrical specimens for compression testing, 6 mm in diameter by 6 mm in height, were machined with the cylindrical axes parallel to the forging direction. Compressive tests were performed at a quasi-static strain rate of 0.001 sec⁻¹ with temperatures ranging from -196 °C to 1200 °C using an Instron screw-driven test machine. High strain rate tests at a strain rate of 2000 sec⁻¹ and temperatures from -196 °C to 1085 °C were conducted using a Split-Hopkinson Pressure Bar [10]. For microstructure and microcrack observations, the deformed specimens were cut parallel and perpendicular to the specimen cylindrical (loading) axis, the specimens were then ground, and polished using standard metallographic methods. The polished surfaces were slightly etched using Kroll’s reagent to view microcracks that developed during testing.

Experimental Results

The initial microstructure of the material studied consists of uniformly distributed fine and equiaxed duplex grains: γ-grains plus lamellar grains, as shown in Fig. 1. The average grain size is about 10-15 μm in diameter. No apparent grain size changes were observed following deformation.

Mechanical Response at High and Quasi-Static Strain Rates

Fig. 2 shows the stress-strain curves at strain rates of (a) 0.001 sec⁻¹ and (b) 2000 sec⁻¹. For the quasi-static strain rate tests (Fig. 2 (a)), the flow stress reaches a saturation stress very quickly at temperatures above 800 °C. For temperatures exceeding 1100 °C no strain hardening is observed and the flow stress is seen to be essentially constant with plastic strain. At temperatures below 800 °C, the quasi-static flow stress continuously increases with strain and no flow stress saturation occurs up to fracture as evidenced by the room temperature stress-strain curve in Fig. 2 (a). The stress-strain curves for temperatures from 25 °C to 600 °C exhibited small but apparent upper yield points. High strain rate test samples were deformed to about 10%, as shown in Fig. 2 (b). The stress-strain curve at 1085 °C and 2000 sec⁻¹ exhibits a similar characteristic to that of the quasi-static strain rate at 800 °C, indicating that the temperature for the flow stress saturation shifts to
Fig. 1 The initial duplex microstructure of fine-grained gamma TiAl alloy.

![Microstructure Image]

**Fig. 2** Stress-strain curves at (a) 0.001 sec\(^{-1}\) and (b) 2000 sec\(^{-1}\).
higher temperatures at high strain rates.

An estimate of the yield stress at the high strain rates was calculated by taking an average of the stress-strain curves at the yield points. The temperature dependence of the yield stress at high and quasi-static strain rates is shown in Fig. 3, where the flow stress at 2% strain for the high strain rate is also presented. The temperature dependence of the yield stress in this fine-grained \( \gamma \)-alloy is seen to be completely different between high and quasi-static strain rates at high temperatures. At temperatures below 600 °C, the yield and flow stresses show a negative temperature dependence (the stresses decrease with increasing temperature) at both strain rates. At temperatures above 600 °C, the yield and flow stresses exhibit a positive (or anomalous) temperature dependence at high strain rates but an increasingly negative temperature dependence at quasi-static strain rates.

![Graph showing the temperature dependence of yield stress at 0.001/s and 2000/s.](image)

**Fig. 3** The temperature dependence of yield and flow stresses at 0.001 sec\(^{-1}\) and 2000 sec\(^{-1}\).

The work hardening rate \( \frac{d\sigma}{d\varepsilon} \) was measured by taking an average slope of the stage-II portions of stress-strain curves in Fig. 2. The work hardening rate is seen to decrease dramatically with temperature at low and high temperatures but exhibits a plateau at intermediate temperatures at both strain rates, as shown in Fig. 4. The work hardening rate plateau extends to higher temperatures at the high strain rate; the work hardening rate plateau starts to drop at 600 °C for the quasi-static strain rate but at 1100 °C at high strain rate. The work hardening rates at high strain rate are shown to be much larger than that at low strain rate. The minimum difference in the work hardening rates between the two strain rates is approximately 1360 MPa in the plateau region between 200 °C and 600 °C (Fig. 4). At higher and/or lower temperatures the difference is even larger. At temperatures greater than 1100 °C no strain hardening was observed in the quasi-static strain rate tests (Fig. 2 (a) and Fig. 4).

The temperature dependence of strain rate sensitivity, \( \frac{\partial(\ln \sigma)}{\partial(\ln \dot{\varepsilon})} \), is shown in Fig. 5. The strain rate sensitivity for low strain rate tests (Fig. 5 (a)) was calculated using the data obtained at strain rates of 0.001 sec\(^{-1}\) and 0.1 sec\(^{-1}\). The strain rate sensitivity for the low strain rate shows an apparent temperature and strain dependence. The increment in strain rate sensitivity is greater than 0.1 when the strain goes from yield to 40% at temperatures below 1100 °C. No strain dependence of the rate sensitivity is observed at 1200 °C. However, the temperature dependence of strain rate sensitivity is remarkable particularly at temperatures above 1100 °C, which is graphically illustrated in Fig. 5 (a) as the temperature approaches 1200 °C where the strain rate sensitivity reaches 0.45. The strain rate sensitivity for the high strain rate tests (Fig. 5 (b)) was calculated using data at strain rates of 35 sec\(^{-1}\) and 2000 sec\(^{-1}\). At the high strain rates, the strain rate sensitivity is less than 0.1 over the entire range of temperatures. Although the strain rate sensitivity increases slightly with increasing temperature, the strain dependence of the strain rate sensitivity at high strain rates is not
**Fig. 4** The temperature dependence of work hardening rate at 0.001 sec\(^{-1}\) and 2000 sec\(^{-1}\).

**Fig. 5** The temperature dependence of strain rate sensitivity at (a) low strain rate and (b) high strain rate.

**Microcracks at Different Strain Rates and Temperatures**

Fig. 6 shows microcracks observed in specimens deformed at (a) 0.001 sec\(^{-1}\) and -196 °C, (b) 0.001 sec\(^{-1}\) and 1200 °C, (c) 2000 sec\(^{-1}\) and -196 °C, and (d) 2000 sec\(^{-1}\) and 1085 °C, respectively. The micrographs in Fig 6 (a), (b) and (d) are obtained from the cross sections perpendicular to the loading directions, and Fig. 6 (c) is from the longitudinal section of the specimen. At quasi-static strain rate and low temperature (Fig. 6 (a)), the microcrack distribution is not uniform. Two distinct regions are seen in this figure: one is dominated by grain boundary microcracks and the other is dominated by grain interior microcracks. The grain interior microcracks are more or less parallel among neighboring grains, exhibiting a band like arrangement. However, the grain boundary microcracks are formed randomly without any preferential orientation relationships among them. At high temperatures, both types of microcracks are uniformly distributed.
Fig. 6 Microcracks in specimens deformed at (a) 0.001 sec\(^{-1}\) and -196 °C, (b) 0.001 sec\(^{-1}\) and 1200 °C, (c) 2000 sec\(^{-1}\) and -196 °C, and (d) 2000 sec\(^{-1}\) and 1085 °C.

throughout the specimen (Fig. 6 (b)). The grain interior cracks are formed within the isolated individual grains and the parallelism of grain interior cracks between grains is also observed.

For the high strain rate and low temperature tests, the microcrack distribution is significantly different from that of the low strain rate tests. Microcracks at high strain rate and low temperatures form uniformly throughout the entire specimen and are dominated by grain interior cracks (Fig. 6 (c)). Most of these microcracks are inclined 45° to the compression axis. The grain boundary microcracks become more significant when the temperature increases, as shown in Fig. 6 (d). However, the uniformity of the microcrack distribution at a high strain rate and a high temperature is between those of the quasi-static low temperature test and the quasi-static high temperature test. The grain interior cracks within a gamma grain are parallel and uniformly distributed throughout the grains at all testing conditions.
Analysis and Discussion

Mechanical Response and Deformation

No apparent grain size change after deformation indicates that no remarkable recrystallization occurred during the deformation. However, the apparent flow stress saturation stage at 800 °C and 0.001 sec⁻¹ (Fig. 2 (a)) suggests that dynamic recovery is a significant component of deformation at temperatures above 800 °C for the quasi-static tests. The propensity for flow stress saturation at large strains at 1085 °C and 2000 sec⁻¹ (Fig. 2 (b)) indicates that dynamic recovery processes are potentially affecting deformation at temperatures above 1085 °C at high strain rate. Accordingly, at low temperatures (<800 °C for the quasi-static strain rate and <1085 °C for the high strain rate), dynamic recovery is not a dominant mechanism affecting defect storage.

The strain rate sensitivity at the quasi-static strain rate is very large (0.45) at 1200 °C. This suggests that the deformation of this fine-grained γ-TiAl alloy may be dominated by superplastic deformation mode at 1200 °C [11]. Because the grain size is small (10 - 15 μm in diameter) and the deformation temperature (1200 °C) is much higher than 0.5Tm, the occurrence of superplastic deformation dominated by grain boundary sliding (GBS) seems reasonable. Evidence of this is also observed at 1100 °C for large strains, where the rate sensitivity is close to 0.3. In fact, the strain rate sensitivity shown in Fig. 5 (a) is an average value of the rate sensitivities between the strain rates of 0.001 sec⁻¹ and 0.1 sec⁻¹. Thus, the significance of a superplasticity contribution to the deformation at 0.001 sec⁻¹ is largely smeared by the averaging with the 0.1 sec⁻¹ data. The strain rate sensitivity corresponding to the strain rate of 0.001 sec⁻¹ should be larger than that calculated from the average. The stress-strain curve for 1100 °C at 0.001 sec⁻¹ (Fig. 2 (a)) also indicates that the strain in this material is completely controlled by superplastic deformation at 1100 °C. These results are consistent with the superplasticity studies on duplex microstructures of TiAl alloys reported in the literature [12-16]. Superplastic deformation in TiAl has been shown to occur at 1025 °C between strain rates of 8.3 x 10⁻⁴ and 1.6 x 10⁻³ for a grain size of 8 μm, where the strain rate sensitivity was equal to $0.33 \approx 0.43$ [12]. It has also been reported that the superplasticity temperature can be as low as 800 °C at 8.3 x 10⁻⁴ sec⁻¹ when the grain size is less than 1 μm [14]. For a grain size of 20 μm, the superplasticity temperature has been shown to increase to 1210 °C at 1 x 10⁻⁴ sec⁻¹ [16]. Because the grain size of the current gamma alloy is between 10-15 μm, the initiation of superplastic behavior above 1100 °C at a strain rate of 0.001 sec⁻¹ is therefore considered reasonable.

The strain rate sensitivity at the high strain rate is less than 0.1 up to 1200 °C (Fig. 5(b)). This suggests that no superplastic deformation is involved in plastic deformation at high strain rates even at temperatures as high as 1200 °C. The deformation at this strain rate should be dominated by dislocation slip and mechanical twinning [17] and obey normal Taylor theory for polycrystalline deformation.

The work hardening rate ($\theta$) is normalized with the Taylor Factor ($M=9.425$) for polycrystalline materials. The ratio of the shear modulus ($G=69620$ MPa [18]) of TiAl to the normalized work hardening rate ($\theta/M$) versus temperature is plotted in Fig. 7. This ratio ranges from 84 to 214 for the high strain rate, which is consistent with the stage-II hardening in FCC metals. At 0.001 sec⁻¹ and temperatures below 800 °C, stage-II hardening dominates. At 25 °C, this hardening extends to fracture as studied previously [19]. This indicates that the stage-II hardening dominates the deformation from yield to fracture (60% strain) at 25°C. The ratio ($GM/\theta$) of 1773 at 800 °C and 0.001 sec⁻¹ suggests third stage work hardening, which starts at a very small strain (about 4% in
Fig. 7 Normalized work hardening rate versus temperature at 0.001 sec\(^{-1}\) and 2000 sec\(^{-1}\).

Fig. 2 (a)). This indicates that dynamic recovery is a significant deformation mechanism above 800 °C quasi-statically in TiAl. The GM/θ ratio becomes essentially infinite due to the zero hardening at temperatures above 1100 °C. This suggests that a completely different deformation mechanism, superplasticity, occurs at high temperatures for quasi-static strain rates, which is consistent with the strain rate sensitivity analysis.

According to the above analysis, it is concluded that the plastic deformation of fine-grained gamma alloy is controlled by stage-II hardening (1) at 0.001 sec\(^{-1}\) and temperatures below 800 °C, (2) at 2000 sec\(^{-1}\) and temperatures up to 1085 °C. At 0.001 sec\(^{-1}\), superplasticity is the deformation mechanism at temperatures above 1100 °C. At temperatures between 800 °C and 1100 °C, the third stage hardening dominates the stress-strain response at 0.001 sec\(^{-1}\).

The significant anomalous temperature dependence of yield and flow stresses at the high strain rate shown in Fig. 3 was also observed in a relatively course-grained Ti-48Al-2Ni-2Cr duplex gamma alloy [20]. The pinned <101> screw superdislocations were observed in specimens deformed at 2000 sec\(^{-1}\) and temperatures above 400 °C [20]. Accordingly, the anomalous temperature dependence of the yield and flow stresses at high strain rates is possibly due to the blocking of superdislocation slips by various blocking mechanisms proposed in the literature [21-23]. However, it has been also reported [24] that superdislocations are observed only at temperatures below room temperature in Ti-48Al and Ti-47.5Al-2.5Cr. At high temperatures, the 1/2<110> normal dislocation slip and mechanical twinning on multiple systems are responsible for deformation process [24]. The anomalous temperature dependence observed in Ti-48Al and Ti-47.5Al-2.5Cr is thus interpreted by the thermally activated formation of jogs along the 1/2<110> dislocation lines which results in a larger stress for dislocation slip [24]. Therefore, the detailed mechanisms controlling the anomalous temperature dependence of the yield stress in Ti-rich (γ + α\(_2\)) two phase alloys should be further investigated.

Microcrack Formation

In Fig. 6 (c), most microcracks are inclined 45° to the loading direction, i.e., in the direction having the maximum resolved shear stress. This indicates that the microcracks observed in this study are most likely formed by shear displacement on particular crystallographic planes.
Numerous translamellar and interlamellar microcracks in a fully-lamellar TiAl alloy were also observed to form by slip plane (or slip band) decohesion [25]. The parallelism of microcracks within each grain in Fig. 6 also indicates that the microcracks are formed by shear displacement on a single crystallographic plane. Thus, the formation of grain interior microcracks may be directly related to deformation by dislocation slip in individual grains. Grain boundary microcracks have no particular orientation relationship with the loading direction. However, the grain boundary microcracks were seen to preferentially form at grain boundaries of grains not possessing grain interior cracks (Fig. 6 (a), (b) and (d)).

The GBS at 1200 °C and 0.001 sec⁻¹ (Fig. 6 (b)) results in the individual grains behaving like isolated single crystals, so the deformation of individual grains to accommodate the overall sample shape change is isolated from one grain to another. The grain interior microcracks formed by the shear displacements on the particular crystallographic planes in each grain are, therefore, restricted within individual isolated grains, as shown in Fig. 6 (b). Although the number of microcracks in Fig. 6 (a) is similar to that in Fig. 6 (b), the strain in Fig. 6 (b) is more than triple that in Fig. 6 (a). Therefore, the propensity for microcracking at -196 °C should be much larger than that at 1200 °C for the quasi-static strain rate.

Microcracks formed at 1085 °C and 2000 sec⁻¹ (Fig. 6 (d)) are similar to those at 1200 °C and 0.001 sec⁻¹, except that the distribution of microcracks in Fig. 6 (d) is more heterogeneous than that in Fig. 6 (b). This may be due to the lack of GBS at the high strain rate. Microcracks formed at -196 °C and 2000 sec⁻¹ (Fig. 6 (c)) are primarily grain interior cracks, which are different from the microcracks shown in Fig. 6. The similar number of microcracks in Fig. 6 (c) and (d) suggests that the deformation temperature may not affect the total number of microcracks at high strain rates but alter the microcrack characteristics. However, if the number of microcracks is normalized with strain, it is seen in Fig. 6 that the total number of microcracks increases as the strain rate increases.

Conclusions

1. At a high strain rate, 2000 sec⁻¹, the stress-strain response of fine-grained gamma TiAl alloy is controlled by stage-II work hardening at -196 °C to 1085 °C. At the quasi-static strain rate of 0.001 sec⁻¹, stage-II work hardening dominates plastic deformation at temperatures from -196 °C to 800 °C; Stage-III work hardening controls the stress-strain behavior at temperatures between 800 °C and 1100 °C; At temperatures above 1100 °C, superplasticity is a dominant deformation mechanism at 0.001 sec⁻¹.

2. The temperature dependence of the yield and flow stresses depends on strain rate. At the quasi-static strain rate of 0.001 sec⁻¹, the yield stress decreases with a plateau between 200 °C and 800 °C as temperature increases. At the high strain rate of 2000 sec⁻¹, the yield and flow stresses exhibit a minimum at 600 °C. At temperatures above 600 °C the yield and flow stresses exhibit an anomalous (positive) temperature dependence.

3. Both grain interior microcracks and grain boundary microcracks are observed in this study. The formation of grain interior microcracks is thought to be due to shear displacements along particular crystallographic planes. The distribution of microcracks becomes uniform and the total number of microcracks decreases as deformation temperature increases at quasi-static strain rate. At high strain rate (2000 sec⁻¹) and low temperature (-196 °C), no grain boundary microcracks are observed and grain interior cracks are uniformly distributed. No apparent temperature dependence of the total number of microcracks is observed at high strain rates.

Acknowledgment

The authors are very grateful to R.W. Carpenter II and M. Lopez for their help with the mechanical tests. This work was performed under the auspices of the U.S. Department of Energy.
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