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FINAL PROGRESS REPORT

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MECHANISMS OF
HIGH TEMPERATURE DEFORMATION AND RUPTURE
UNDER MULTIAXIAL LOADING CONDITIONS

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I. SUMMARY OF THE RESEARCH

This research program is concerned with the mechanisms that control high temperature fracture processes in specimens subjected to multiaxial loading conditions. The experimental program has consisted of detailed studies of the influence of multiaxial stress states on high temperature damage development which determine the life of the specimens. The objective of the work is to achieve new insight into the predominant failure mechanisms and to develop novel methods for predicting the failure of high temperature components.

The four-year program was initiated on July 1, 1993; the present report covers the period from July 1, 1996 - June 30, 1997. The principal developments for this reporting period were from investigations of the multiaxial high temperature fracture behavior of two materials: 7075 Al, in which grain boundary sliding is known to occur, and fully-lamellar titanium aluminide, in which grain boundary sliding is inhibited. The work has been performed by graduate students Ali Yousefiani and Nancy Johnson, respectively, under the supervision of the principal investigators.

High-temperature deformation and rupture behavior of commercial high strength 7075 Al has been studied under uniaxial, biaxial, and triaxial stress states. Tests were conducted at 648 K, over four orders of magnitude of strain rate. Rupture times for the different stress states were compared with respect to four different mechanistic multiaxial stress parameters, which are each linked to a particular physical mechanism controlling the creep rupture process. The results indicate that the principal facet stress parameter successfully correlates the data over the entire stress range investigated. This finding suggests that the creep rupture process is primarily controlled by cavitation coupled with a localized deformation process along inclined directions. This strain localization leads to stress redistribution to cavitating regions of the microstructure. It appears that the localized mechanism redistributing stress in 7075 Al at elevated temperatures is grain boundary sliding. A detailed description of our research on 7075 Al is presented in Section II.A of the present report.

Heat treatments of a TiAl alloy have been established for producing different microstructures in high temperature crack growth specimens. In particular, we have examined the high temperature behavior of a fully lamellar microstructure in which grain/colony boundary sliding appears to be inhibited. Despite the absence of this localized deformation mechanism, the data are best correlated by the principal facet stress suggesting that another localized mechanism is operational. Candidates for such a mechanism include lath boundary sliding and fast creep within the softer α_2 laths. The success of the principal

facet stress in correlating rupture data for fully-lamellar TiAl as well as that for many other alloys indicates the prevalence of localized deformation at elevated temperatures. This finding implies that homogeneous (continuum) creep in the absence of localized deformation is more the exception than the rule. A detailed description of our study of fully lamellar titanium aluminide is presented in Section II.B of the present report. This section is followed by a listing of the publications, presentations and dissertations resulting from the program; a listing of the graduate research assistants; and a listing of other research support.

II. RESEARCH REPORTS

A. HIGH-TEMPERATURE DEFORMATION AND RUPTURE OF 7075 Al UNDER MULTIAXIAL STRESS STATES

A.1. Introduction

Comprehensive studies concerning failure of engineering materials at high-temperatures have commonly been conducted under uniaxial stress conditions. These investigations have provided substantial information regarding the factors governing creep rupture. Different failure modes have been classified, and systematic descriptions for the mechanics and mechanisms of creep fracture have been well established [1-3]. It has been observed [4] that, for uniaxial creep conditions, a fundamental power law exists which simply relates the time to fracture, t_f , and the applied stress, σ_{app} , as follows:

$$t_f = M\sigma_{app}^{-X} \quad (A.1)$$

where M and X are stress independent constants for a given material and testing condition.

In high-temperature applications, however, the majority of the components are subject to stress states varying in both time and position. Under such complex states, the stress used in Eqn. A.1 must be modified, to correctly predict the fracture time. Due to the difficulties in establishing stress distributions for creep deformation of multiaxially stressed

components, the general effort has been geared towards obtaining representative stress parameters that can forecast t_f , utilizing data from conventional uniaxial creep rupture tests.

Most stress parameters developed for predicting creep life under multiaxial stress states are based on continuum mechanics approaches [5-8]. Multiaxial stress terms are combined in a weighted formula, and the relative contribution of each term is described by adjustable factors which are determined from data generated at different stress states. For example, it has been shown that, under complex stresses, Eqn. A.1 can be generalized as [5]:

$$t_f = M(\alpha\sigma_I + \beta\sigma_e + \gamma\sigma_H)^{-X} \quad (\text{A.2})$$

where σ_I is the maximum principal stress, σ_H is the hydrostatic component of stress given as

$$\sigma_H = (\sigma_1 + \sigma_2 + \sigma_3) \quad (\text{A.3})$$

expressed in terms of the principal stresses, $\sigma_1 > \sigma_2 > \sigma_3$, and σ_e is the von Mises effective stress defined by

$$\sigma_e = \sqrt{\frac{(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2}{2}} \quad (\text{A.4})$$

Through analysis of experimental data, coefficients α , β , and γ are adjusted to obtain the optimum correlation between the rupture time data from different stress states. Multiaxial stress terms σ_I and σ_e , as will be discussed later, are known to be the driving force for different processes contributing to creep rupture. At high-temperatures, cavity growth can also be controlled by σ_H . This provides a physical basis for the appearance of these stress terms in Eqn. A.2. Contributions of individual stress terms in this approach are considered to be independent. Other representative stress parameters in this category are developed on a similar foundation.

Multiaxial stress parameters advanced by the continuum mechanics approach may be utilized satisfactorily to predict the creep life of components. However, it should be considered that; (1) determination of these parameters will require extensive multiaxial

creep and rupture data, and (2) limited information is provided regarding the physical mechanisms involved in the creep rupture process. Such limitations can be overcome using the multiaxial stress parameters developed through mechanistic approaches. Characterized by the fact that their determination does not involve adjustable terms, these stress parameters are each associated with a specific set of physical mechanisms that control the creep rupture process. These mechanism-specific parameters are briefly discussed below.

The Maximum Principal Stress.

Acting alone, this stress component has been considered as a multiaxial stress parameter since: (1) diffusive growth of intergranular cavities is driven by tensile stresses acting normal to the grain boundaries, and (2) intergranular fracture has usually been observed to initiate on grain boundaries perpendicular to the maximum principal stress. This parameter usually dominates when cavities nucleate readily and distribute homogeneously on all grain boundaries. Under such conditions, cavitation is not constrained by creep.

The von Mises Effective Stress.

This stress component is usually considered as a correlating parameter when shear stresses governing creep deformation become the driving force for processes controlling creep rupture. Such a situation occurs when cavity distribution is inhomogeneous, and cavitating grain boundary facets are isolated from one another [9]. It has also been suggested that σ_e has a significant role in determining the life of microstructurally unstable components which undergo strain softening during creep deformation [10].

Cane's Representative Rupture Stress.

An alternative stress parameter was developed for conditions where grain boundary cavitation is constrained by the creep rate of the surroundings [11]. In this case, a

combination of stress components has been used to correlate multiaxial rupture data. It is considered that the creep rate depends on the effective and deviatoric stress components, and the cavity spacing is related to σ_I . The resulting representative stress parameter is given by

$$\sigma_c = (1.5)^{\frac{1}{n+0.5}} (1-A)^{\frac{n}{n+0.5}} (\sigma_I)^{\frac{0.5}{n+0.5}} (\sigma_e)^{\frac{n-1}{n+0.5}} (\sigma_I')^{\frac{1}{n+0.5}} \quad (\text{A.5})$$

where n is the stress exponent for power law creep and A is the area fraction of cavitating grain boundaries. The maximum deviatoric stress is defined as

$$\sigma_I' = \sigma_I - \sigma_H. \quad (\text{A.6})$$

It has been noted that $A \approx 0.33$ when constrained growth occurs on all grain boundaries normal to the maximum principal stress.

The Principal Facet Stress.

This stress parameter was derived for situations where cavitation is coupled with highly localized deformation processes, such as grain boundary sliding [12]. During sliding, shear stresses on inclined boundaries are relieved, giving rise to a redistribution of normal stresses. The average tensile stress on grain boundary facets perpendicular to σ_I is therefore amplified, and it is suggested that this enhanced stress level drives the rupture process. The principal facet stress parameter is given by

$$\sigma_F = 2.24\sigma_I - 0.62(\sigma_2 + \sigma_3). \quad (\text{A.7})$$

The preceding discussion suggests that, in addition to predicting creep life of a component under multiaxial stress states, mechanistic stress parameters can be utilized to gain a better understanding of the physical processes governing high-temperature rupture. Therefore, the successful criteria will best predict rupture times for multiaxial states from uniaxial creep data, and will also help identify the dominant creep fracture mechanism.

Multiaxial creep fracture behavior of Al and its alloys is currently being investigated. The preliminary results obtained for 7075 Al are examined in the present work. Validity of

the aforementioned mechanistic criteria are determined using stress state as a variable, and the results are compared with those of related investigations on Al and its alloys.

A.2. Procedures

The present work was performed on commercial 7075 Al. The material was received as rolled plates in T6 condition. Samples were solutionized at 723 K for 4 h and subsequently air cooled to room temperature, where they were kept for at least a week before testing. Creep tests were conducted in air and under constant stress conditions. Testing temperature was maintained at 648 ± 2 K, and the applied stress ranged from 5 to 30 MPa. All specimens were machined with their long axis parallel to the longitudinal rolling direction.

Three types of specimens, each corresponding to a different stress state, were utilized in this investigation (Table A.1). Uniaxial stress state was obtained using conventional tensile creep specimens. The state of biaxial shear was achieved using a modified double shear specimen [13], and a Bridgman circular notched tensile bar was applied to produce a triaxial state of stress. Finite element analysis of creep deformation in the notches has been carried out for both biaxial [14] and triaxial [15] specimens. Stress states were determined for the establishment of relative steady state flow, and the results are outlined in Table A.1.

A.3.1. High-Temperature Deformation

The results for the uniaxial and biaxial creep tests are concordant to the well established methods used to study the creep behavior of materials. In order to similarly utilize the triaxial data, the effective stress and effective strain rate should ideally be used. This requires more detailed work and has not been considered in the present investigation. Nonetheless, in this case, the response of the maximum principal strain and strain rate has been presented for comparison.

Table A.1- Specimen geometry, stress analysis and corresponding stress parameters.

<i>Geometry and stress state</i>	<i>Stress Analysis</i>	<i>Stress parameters</i>
<p>Uniaxial Tension</p>	$\sigma_{app} = F/A_{min}$ $\sigma_1 = \sigma_{app}$ $\sigma_2 = 0$ $\sigma_3 = 0$	$\sigma_1 = \sigma_{app}$ $\sigma_e = \sigma_{app}$ $\sigma_c = 0.69\sigma_{app}$ $\sigma_F = 2.24\sigma_{app}$
<p>Biaxial Shear</p>	$\tau_{max} \cong F/2A_{min}$ $\sigma_1 = \tau_{max}$ $\sigma_2 = 0$ $\sigma_3 = -\tau_{max}$	$\sigma_1 = \tau_{max}$ $\sigma_e = 1.73\tau_{max}$ $\sigma_c = 1.12\tau_{max}$ $\sigma_F = 2.86\tau_{max}$
<p>Triaxial Tension</p>	$\sigma_{nom} = F/A_{max}$ $\sigma_1 = 2.7\sigma_{nom}$ $\sigma_2 = 0.33\sigma_1$ $\sigma_3 = 0.33\sigma_1$	$\sigma_1 = 2.7\sigma_{nom}$ $\sigma_e = 1.8\sigma_{nom}$ $\sigma_c = 1.29\sigma_{nom}$ $\sigma_F = 4.94\sigma_{nom}$

Results of creep tests conducted for different stress states are presented in Fig. A.1. The true tensile strains, ϵ , and shear strains, γ , correspond to values converted from measured displacements. The applied stress ranged from 5 to 30 MPa, covering over four orders of magnitude of strain rate. It should be noted that the instantaneous strain, which generally increased with increasing stress, was subtracted from the total strain of the specimen. Therefore, the creep curves shown in Fig. A.1 start at the origin of the coordinates and represent only the strain due to creep.

Fig. A.1 Creep data for 7075 Al at 648 K.

In general, similar creep behavior was recognized for all types of specimens. All creep curves exhibit: (1) a primary creep stage which contributes more, relative to t_f , as the applied stress increases, (2) almost no observable secondary creep, which explains the use of the minimum strain rate instead of steady-state creep rate where applicable, and (3) an extended tertiary creep stage that accounts for most of the sample life.

For practical purposes, the stress dependence of steady state creep rate can be represented by a creep power law, which can be generalized for multiaxial stress states as follows [16]:

$$\dot{\epsilon}_e = B\sigma_e^n \quad (\text{A.8})$$

where B is a temperature and microstructure dependent constant, n is the stress exponent, σ_e is the effective stress, and $\dot{\epsilon}_e$ is the effective strain rate. The steady state creep data for all three conditions can be correlated using Eqn. A.8. However, this requires further work, and has not been considered here. Using uniaxial and biaxial data separately, the value of the stress exponent for 7075 Al was determined to be ~ 5.3 .

Exact determination of the mechanism controlling creep deformation requires further work, but the results imply that, under present testing conditions, 7075 Al exhibits the same characteristics of dislocation climb controlled creep as observed in pure metals and class II alloys. Early work on this alloy [17] indicates that its creep characteristics are microstructure and temperature dependent. Their results at high temperatures (723 K) obeyed a power law relation, showing a relatively high stress exponent ($n=6.3-7.8$). Whereas, at low temperatures (373-423 K), no simple power law was observed, and n varied from values close to 3.5 at low stresses to anomalously high values at higher stresses.

A.3.2. High-Temperature Rupture

High-temperature rupture data for the alloy are presented in Fig. A.2. The plots compare each of the representative stress parameters with respect to rupture time on a double logarithmic scale. Using the general form of Eqn. A.1, a regression analysis was applied to calculate the least square fit through the rupture data fields for all three stress states. The best fit and the corresponding correlation coefficients, as shown in the figure, can be used to accurately determine the validity of each stress parameter. It is evident that the principal facet stress ($R^2=0.95$) correlates the rupture time for 7075 Al better than any

of the other three parameters, bringing all rupture data onto a single curve. This result indicates that, under the present experimental conditions, creep fracture is dominated by cavitation that is coupled with a localized deformation process along inclined directions.

Fig. A.2 Multi-axial rupture data for 7075 Al.

In order to investigate the cavitation behavior, both optical microscopy and scanning electron microscopy (SEM) were performed. Figure A.3 shows the cross sections of four specimens tested to rupture under uniaxial conditions, with applied stresses ranging from 7.88 MPa to 29.96 MPa. Each image is a montage of optical micrographs, covering a large area (≈ 2 mm) of the polished fracture tip. In this way, both the neck profile and cavitating regions away from the tip can be examined and compared. Figure A.4 shows SEM micrographs of the rupture surfaces of uniaxial specimens tested under applied stresses of 7.88, 13.7, and 21.51 MPa.

Examination of Fig. A.3 reveals that (1) widespread cavitation occurs in all samples, (2) the extent of cavitation decreases with increasing stress, (3) cavity distribution

Fig. A.3 Optical micrograph of polished cross sections of uniaxial samples tested at (a) 7.88, (b) 8.86, (c) 19.25, and (d) 29.96 MPa.

is more homogeneous at lower stresses, and blunt microcracks that developed due to cavity coalescence are clearly visible, (4) cavity morphology changes at higher stresses, appearing more elongated towards the tensile axis, and (5) as observed from the necking profiles, deformation becomes much more localized with increasing stress. Similar observations and trends can be recognized in the SEM micrographs shown in Fig. A.4. The dimpled nature of the fracture surfaces indicates that failure resulted from extensive cavitation. Also, the change in depth, size, and ridge sharpness of the dimples with increasing stress

Fig. A.4 SEM micrographs of the rupture surface of uniaxial specimens tested at (a) 7.88, (b) 13.7, and (c) 21.51 MPa.

confirms the trends outlined above. These observations suggest that, apparently at higher stresses, the final fracture mode is influenced by deformation controlled processes.

It would be tempting to argue that the change in cavity morphology discussed above originates from a transition in the high-temperature rupture mechanism to one dominated by creep deformation. It could accordingly be suggested that, at high stresses, related stress parameters (σ_e or σ_C) should be used to correlate the multiaxial rupture data. However, the results shown in Fig. A.2 indicate that such an assumption is obviously misleading, since the principal facet stress is clearly superior in correlating all the rupture data over the entire stress range. A reasonable explanation can be proposed with the understanding that (1) in most cases where cavitation is extensive, void growth by creep deformation usually becomes dominant in the final stages of fracture, and (2) during creep fracture, different mechanisms typically occur in a coupled manner, each being dominant for a certain fraction of the creep life [2]. For the alloy under investigation, it is therefore most probable that cavitation coupled with grain boundary sliding governs most of the creep life, but final fracture is controlled by creep deformation. Consequently, the principal facet stress will be successful in bringing the rupture data together, since it predominates for most of the rupture time. On the other hand, the strain to failure and/or the final fracture mode may be categorized by creep deformation processes.

The principal facet stress parameter has also been successful in correlating the high temperature rupture data for several other alloys [12,18]. It has been shown that the stress redistribution necessary for the validity of this parameter could be originated from boundary displacements on the order of the elastic repositioning of the grains. This highly localized mode of deformation can be obtained through sliding of grain boundaries or softened precipitate free zones in the vicinity of the boundaries. In the case of 7075 Al, it has been clearly shown that, at very high temperatures, grain boundary sliding becomes the major mode of creep deformation [17]. Also in support of the previous discussion, a detailed metallographic investigation [19] explicitly indicates that, although grain boundary

sliding is dominant at very high temperatures, it is not the only process involved. Plastic flow was found to occur as an accommodating process which is coupled with grain boundary sliding.

It is interesting to note that early multiaxial creep rupture studies [20, 21] have found the von Mises effective stress to be the best rupture criterion for a number of Al alloys. Based on these results, Al alloys have been repeatedly referred to as model materials that correlate with σ_e . This contrasts with the results obtained in the present work. The difference can simply be rationalized by considering the temperature ranges (453-483 K) in which the previous investigations were conducted. It is suggested that at these relatively low temperatures, creep deformation in the alloys is not controlled by grain boundary sliding and other more homogeneous modes of deformation are dominant. The concept has been verified for 7075 Al [17], where it has been shown that grain boundary sliding dominates at relatively high temperatures (723 K), while between 373-423 K no sign of boundary sliding can be observed. Therefore, it can be concluded that the probable cause for failure of the principal facet stress with regard to other Al alloys tested at lower temperatures is the absence of a localized deformation process such as grain boundary sliding, which is one of the main requirements for the application of this parameter.

A.4. Conclusions

Multiaxial creep rupture of commercial 7075 Al was investigated at 648 K, with applied stresses ranging from 5 to 30 MPa. Results from the creep data indicated relatively similar behavior for all stress states, and the creep exponent for the alloy was found to be approximately 5.3. The rupture data have been compared with respect to different physically based multiaxial parameters, each linked to general mechanisms known to control high-temperature rupture of materials. The success or failure of these criteria has been used as a means to identify the governing creep fracture mechanism.

It was found that the principal facet stress successfully correlates the rupture time for 7075 Al, bringing the rupture data for all stress states onto a single curve. This indicates that, under the present conditions, creep fracture is dominated by cavitation coupled with localized deformation processes along inclined directions. Microstructural observations of cavitation presented in this work, and direct evidence of grain boundary sliding reported elsewhere, lend to support this finding.

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B. MICROSTRUCTURE DEVELOPMENT AND HIGH TEMPERATURE DEFORMATION AND RUPTURE OF TITANIUM ALUMINIDE UNDER MULTIAXIAL STRESS STATES

B.1. Introduction

Creep deformation and rupture are important phenomena for the many engineering components that operate under service conditions at elevated temperatures. The bulk of available creep data has been generated under uniaxial tension loadings [Cane, 1982 #26]. However, engineering structures are often submitted to multiaxial loading conditions. The uniaxial rupture data cannot be directly extrapolated to the more general multiaxial case. Because high temperature deformation and rupture of structural materials is often the result of multiaxial loading, it is important to understand and be able to predict creep behavior under multiaxial stresses.

Intermetallic materials have received significant attention recently due to their attractive high temperature properties [1]. In particular, titanium aluminide is being considered for use in high temperature applications such as gas turbines and engine casings [2] because of its low density and high temperature strength. However, titanium aluminide possesses low ductility at room temperature. Many applications in which titanium aluminide looks to be an attractive material require high toughness. Through heat treatment, high temperature strength and room temperature ductility can be balanced to provide a more usable material [1]. The high degree of multiaxial stresses present in many of its potential applications makes the understanding of high temperature behavior especially important.

B.2 Titanium Aluminide

Titanium aluminide with an aluminum composition between 44 and 52 at% Al demonstrates three morphologies. The Ti-Al phase diagram in the area of interest is shown

in Figure B.1. The two phases of interest, α_2 and γ , are shown in Figure B.2. TiAl (γ) has a face-centered tetragonal structure ($L1_0$) while Ti_3Al (α_2) has a ordered hexagonal structure (DO_{19}). Depending on heat treatment, the microstructure will be one of the following: single phase equiaxed γ grains, fully lamellar grains consisting of alternating α_2 (Ti_3Al) and γ (TiAl) laths, or a duplex structure with equiaxed γ grains as well as lamellar ($\alpha_2 + \gamma$) grains. The duplex structure has been shown to exhibit higher yield strengths and greater ductility below 600°C while the fully lamellar structure has superior high temperature strength and creep resistance [3]. Chromium and niobium are common alloying additions in titanium aluminide. Chromium has been shown to enhance ductility of duplex TiAl at room temperature [4]. The addition of niobium increases the strength of the alloy and improves its oxidation behavior [5] by stabilizing the Al_2O_3 layer [6].

Figure B.1. The titanium-aluminum phase diagram.

Figure B.2. Crystal structures of a) face-centered tetragonal γ and b) hexagonal α_2 Ti-Al phases.

B.3. Procedures

The material used in the present study was supplied by Titanium Metals Corporation of America with a nominal atomic composition of Ti-48Al-2Cr-2Nb. The ingot chemistry is shown in Table B.1. The as cast material exhibited a duplex structure composed of equiaxed γ grains and lamellar $\alpha_2 + \gamma$ grains with an average grain size of 8 μm [7].

Table B.1 - Chemical composition of titanium aluminide used in the present investigation.

Element	Al	Nb	Cr	Fe	O	N	Ti
Composition (wt %)	33.4	4.70	2.30	0.050	0.071	0.005	balance

All heat treatments were performed under a high purity argon atmosphere. Microstructures were examined using optical microscopy. Samples were etched with a solution composed of 10 ml lactic acid, 10 ml nitric acid, 1 ml hydrofluoric acid, and 100 ml distilled water for 20 to 60 seconds [6].

Creep samples were machined in three configurations: uniaxial tension, triaxial tension, and biaxial shear according to the geometries in Table A.1. The creep samples were wrapped in tantalum foil which acted as an oxygen getter and heat treated in a high purity argon atmosphere according to the schedules described below. Creep tests were performed in air under constant load and constant stress conditions as noted. The bulk of the samples were crept at 1088 ± 2 K. Two uniaxial samples were crept at 1033 ± 2 K.

B.4 Results

Heat Treatment

The as-received material was heat treated under a variety of conditions to find an optimal microstructure for creep testing. The heat treatment schedules and resulting microstructure morphologies are listed in Table B.2. For clarity, heat treatments will be referred to by the designations listed in the Table. Figure B.3 shows the equiaxed γ microstructure of the as-cast material. Heat treatment A ($1400^\circ\text{C}/1\text{hr}$, 5°C min furnace

Table B.2. Heat treatment schedules for Ti-48Al-2Cr-2Nb.

Designation	Heat Ramp	Heat Dwell	Cool Ramp	Microstructure
A	11.6°C/min	$1400/1\text{ hr}$	5°C/min	primarily fully lamellar with some previous γ at grain boundaries, large grains
B	11°C/min	$1400/1\text{ hr}$	22°C/min	duplex
C	12°C/min	$1425/1\text{ hr}$	22°C/min	duplex
D	12°C/min	$1440/1\text{ hr}$	22°C/min	duplex
E	15°C/min	$1450/1\text{ hr}$	10°C/min	duplex
F	15°C/min	$1425/1\text{ hr}$	2°C/min for 2 hr (to 1185°C), 5°C til RT	primarily equiaxed gamma, larger grains than as cast
G	15°C/min	$1440/1\text{ hr}$	air cooled to 1233°C then water quenched	fully lamellar
H	15°C/min	$1440/1\text{ hr}$	water quenched	fully lamellar

cool) resulted in the large lamellar grains with no previous γ grains in the bulk of the sample shown in Figure B.4a. However, as shown in Figure B.4b, part of the sample still had γ grains along the lamellar grain boundaries. Figure B.5 shows the duplex microstructure that resulted from heat treatment D (1440/1 hr, air cool). The fully lamellar structure with interlocking laths along the grain boundaries that resulted from heat treatment H (1440/1 hr, water quench) is shown in Figure B.6.

Figure B.3. Optical micrograph of the as-cast microstructure.

Figure B.4. Optical micrographs of heat treatment A showing lamellar grains with interlocking boundaries with a) no previous γ grains and b) previous γ grains at boundaries between lamellar grains.

Figure B.5. Optical micrographs of heat treatment D showing duplex structure.

Figure B.6. Optical micrograph of heat treatment H. The material is fully lamellar with interlocking laths along grain boundaries.

Creep Rupture Tests

High-temperature rupture data for the alloy are shown in Fig. B.7. As before, a regression analysis was applied to the data in the general form of Eqn. A.1 to calculate the least square fit through the rupture data for all three stress states. The maximum principal stress gives scattered results ($R^2 \approx 0.07$). The von Mises effective stress results in a close fit ($R^2 \approx 0.94$), but the principal facet stress ($R^2 \approx 0.98$) gives the best fit and brings all rupture data onto a single curve. This result indicates that the creep rupture is dominated by localized deformation processes that result in stress redistribution. Although there are interlocking boundaries between the lamellar grains, it is possible that the lamellae are sliding along each other during primary creep and allowing the formation of cavities or that deformation the weaker α_2 lamellae gives rise to stress redistribution that is consistent with the principal facet stress theory. Lath boundary sliding in lamellar TiAl alloys was proposed by Hazzledine and co-workers that involves the interaction of superdislocations and certain misorientations of adjacent γ/γ laths. Experimental observations by Es-Souni et al. [10] and numerical results by Chakraborty and Earthman [11] indicate that lath

boundary sliding does not occur during secondary creep in fully lamellar TiAl due to the low mobility of superdislocations at elevated temperatures. However, these findings do not preclude the possibility of lath boundary sliding in the early stages of primary creep which could ultimately affect the rupture time. Metallographic work is currently underway to determine whether lath boundary sliding and rapid creep within α_2 laths occur under the present conditions.

Conclusions

A heat treatment schedule for Ti-48Al-2Cr-2Nb that results in a fully lamellar microstructure with interlocking laths at colony boundaries was developed. Titanium aluminide in this condition was creep tested at 815 °C under uniaxial tension, biaxial shear and triaxial tension stress states. Rupture data for the material correlated well with the principal facet stress indicating that a localized deformation process gives rise to stress redistribution within the microstructure. Although grain boundary sliding is suppressed due to interlocking, lath boundary sliding or creep within weaker α_2 laths could facilitate the required stress redistribution.

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III. PUBLICATIONS, PRESENTATIONS AND DISSERTATIONS

Journal Articles

K. L. Murty, B. V. Tanikella, and J. C. Earthman, "Effect of Grain Shape and Texture on Equi-Biaxial Creep of Stress Relieved and Recrystallized Zircaloy-4," *Acta Metallurgica et Materialia*, 42, 3653-3662 (1994).

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W. E. Churley and J. C. Earthman, "Damage Mechanics Approach for Predicting High Temperature Crack Growth under Mixed-Mode Conditions," *Materials Science and Engineering*, A202, 36-42 (1995).

A.-B. A. El-Nasr, F. A. Mohamed, and J. C. Earthman, "High Temperature Rupture of a SiC Particulate Reinforced Al Composite Under Multiaxial Stress States," *Materials Science and Engineering*, A214, 33-41 (1996).

W. E. Churley and J. C. Earthman, "High Temperature Crack Growth in 304 Stainless Steel under Mixed-Mode Loading Conditions," *Metallurgical and Materials Transactions A*, 28A, 763-773 (1997).*

A. Yousefiani, A. A. El-Nasr, F. A. Mohamed and J. C. Earthman, "High-Temperature Deformation and Rupture of Unreinforced and Particulate-Reinforced 7075 Al Under Multiaxial Stress States," *in preparation*.

A. Yousefiani, F. A. Mohamed and J. C. Earthman, "High-Temperature Deformation and Rupture of 2024 Al Under Multiaxial Stress States," *in preparation*.

Papers in Conference Proceedings

A. Yousefiani, A. A. El-Nasr, F. A. Mohamed and J. C. Earthman, "High-Temperature Deformation and Rupture of 7075 Al Under Multiaxial Stress States," *Proceedings of the Seventh International Conference on Creep and Fracture of Engineering Materials and Structures*, J. C. Earthman and F. A. Mohamed, eds., TMS, 439-448 (1997).*

N. L. Johnson and J. C. Earthman, "Numerical Simulations of Primary Creep in a Particulate Reinforced Metal Matrix Composite Under Multiaxial Stress States," *Proceedings of the Seventh International Conference on Creep and Fracture of Engineering Materials and Structures*, J. C. Earthman and F. A. Mohamed, eds., TMS, 729-728 (1997).*

Published Abstracts

N. L. Johnson and J. C. Earthman, "Evolution of Internal Multiaxial Stress States in Particulate Reinforced Metal Matrix Composites Under Creep Conditions," *Proceedings of the International Conference on Composites Engineering*, New Orleans, (1994).

A.-B.A. El-Nasr, F. A. Mohamed and J. C. Earthman, "High Temperature Rupture of a SiC Particulate Reinforced Al Composite Under Multiaxial Stress States," *1995 TMS Fall Meeting*, Cleveland, Oct. 30 - Nov. 3.

V. OTHER RESEARCH SUPPORT

James C. Earthman

Electric Power Research Institute, "Corrosion Prevention by Regenerative Biopolymers," \$345,870 for the period 3/95-2/98 (Co-investigator: T. K. Wood). The primary objective of this research is to perform and evaluate biological techniques for preventing corrosion attack by sulfate-reducing bacteria (SRB) as well as other corrosion inducing agents. It is anticipated that the present research will lead to the development of new biological methods that could economically reduce corrosion rates in many applications.

National Science Foundation, "Role of Impurities on Superplastic Flow and Cavitation," \$370,704 for the period 8/95-7/98 (PI: F. A. Mohamed). The primary objective of the investigation is to develop a better understanding of the ways in which impurity segregation at interfaces influences the characteristics of superplastic flow at low stresses. The experimental approach involves a detailed examination of the effect of selected impurity elements on dislocation motion, plastic instability, interfacial sliding, microstructure, and cavitation processes. In particular, cavity nucleation and the relationships between deformation and cavity growth are closely investigated.

Farghalli A. Mohamed

National Science Foundation, "Role of Impurities on Superplastic Flow and Cavitation," \$370,704 for the period 8/95-7/98 (Co-investigator: J. C. Earthman). The primary objective of the investigation is to develop a better understanding of the ways in which impurity segregation at interfaces influences the characteristics of superplastic flow at low stresses. The experimental approach involves a detailed examination of the effect of selected impurity elements on dislocation motion, plastic instability, interfacial sliding, microstructure, and cavitation processes. In particular, cavity nucleation and the relationships between deformation and cavity growth are closely investigated.

United States Army Research Office, "Spray Deposition Processing of Al/SiC MMC Armor Materials and Ta Alloys for Army Applications," \$300,000 for the period 7/95-6/98 (PI: E. J. Lavernia). The Primary objective of the program is to use a spray atomization and deposition synthesis approach to process ceramic particulate reinforced Al alloy metal matrix composites (MMCs) as armor materials and Ta alloys for Army applications such as shaped charged liners and explosively formed projectile. The scientific Objective of the investigation is to enhance our understanding of the fundamental mechanisms that govern the processing-microstructure-mechanical behavior synergism in Al alloy based MMCs and Ta alloys.