Thermal-cycling creep of γ -TiAl-based alloys

C. Schuh^a, D.C. Dunand^{a,*}, A. Wanner^b, H. Clemens^b

^aDepartment of Materials Science and Engineering, Northwestern University, 2225 N. Campus Drive, Evanston, IL 60208, USA

^bInstitut für Metallkunde, Universität Stuttgart, Seestr. 71, D-70174 Stuttgart, Germany

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Abstract

In two-phase TiAl-based alloys, the coexisting α_2 and γ phases exhibit a thermal expansion mismatch, so that increased creep rates during thermal cycling may be expected. Creep deformation of two γ -TiAl-based alloys was investigated during thermal cycles between 900 and 300 or 350°C with applied tensile stresses of 32.5 or 37.0 MPa. Measured thermal cycling creep rates were compared with isothermal creep rates calculated at the effective average temperature. No creep enhancement was measured upon thermal cycling within uncertainties of $1.6 \times 10^{-3}\%$ strain per cycle. © 2000 Elsevier Science Ltd. All rights reserved.

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1. Introduction

Titanium aluminide alloys based on γ -TiAl have been developed to exhibit an optimum combination of room-temperature ductility and high-temperature strength and toughness [1,2]. With recent developments in shape-forming processes such as rolling, extrusion, and superplastic forming [3–5], these materials are suited for application in the aerospace (e.g. compressor and turbine blades) and automotive (e.g. valves) industries. Because of their two-phase structure composed of α_2 - and γ -phase lamellae, γ -TiAl alloys exhibit high creep resistance [2,6]. However, the difference in thermal expansion coefficient between the two phases generate internal mismatch stresses upon thermal cycling (i.e. engine start-up and shut-down). For example, Appel and Wagner [7] performed an in-situ TEM study of a γ -TiAl-based alloy with fully lamellar microstructure during a thermal cycle from room temperature to 721°C and back to room temperature. They observed the generation of dislocations in the γ phase at 721°C which they attributed to thermal mismatch between the phases.

As reviewed in Refs. [8–10], materials which are subject to both an external stress and repeatedly-generated internal stresses exhibit a unique mode of deformation

called *thermal-cycling creep*. This phenomenon is most commonly observed during thermal cycling of (a) polycrystalline single-phase materials with anisotropic thermal expansion, (b) two-phase materials, where internal stresses are generated by thermal expansion mismatch, and (c) allotropic materials, where repeated phase transformation produces internal mismatch stresses (i.e. transformation superplasticity). When the applied external stress is small compared to the internal stress, higher deformation rates are measured during thermal cycling as compared to isothermal creep at an equivalent temperature. Therefore, this type of deformation is undesirable for engineering components for use at elevated temperatures.

Sato and Kuribayashi [11] developed a micro-mechanical model for thermal-cycling creep, based upon the thermal-expansion mismatch between a rigid spherical particle and a creeping matrix. During thermal cycling, internal mismatch stresses result in an additional strain rate $\dot{\epsilon}$, which augments the strain rate due to normal creep mechanisms. Sato and Kuribayashi derived the following expression for $\dot{\epsilon}$:

$$\dot{\epsilon} = \frac{2 \cdot n \cdot (n + 4)}{2^{1/n} \cdot 5} \cdot \frac{f \cdot (1 - f^{1/n})}{f^{1/n} \cdot (1 - f)^2} \cdot (\Delta\alpha \cdot \dot{T})^{1-1/n} \cdot \dot{\epsilon}_{\text{eff}}^{1/n} \quad (1)$$

where the matrix phase creeps according to a power-law with stress exponent n , the particle phase has volume fraction f , the thermal expansion coefficient mismatch is

* Corresponding author. Tel.: +1-847-491-5370; fax: +1-847-491-7820.

E-mail address: dunand@nwu.edu (D.C. Dunand).

$\Delta\alpha$, and $\dot{T}$ is the heating and cooling rate of a triangular thermal cycle. The effective strain rate $\dot{\epsilon}_{\text{eff}}$ is the strain rate during thermal cycling due to creep of the matrix without a superimposed internal stress state, and can be found from:

$$\dot{\epsilon}_{\text{eff}} = \frac{1}{\Delta t} \int_0^{\Delta t} \dot{\epsilon}(T) \cdot dt \quad (2)$$

where t is time, Δt is the thermal cycle duration, and $\dot{\epsilon}(T)$ is the isothermal creep rate at absolute temperature T .

The distribution of the α_2 phase within γ -TiAl-based alloys (such as those used in the present study) is more complex than in the model by Sato and Kuribayashi, which thus cannot be directly applied to these alloys. However, we used this model to estimate whether or not creep enhancement due to thermal cycling is expected at all for the present experiments. We take the volume fraction of α_2 phase $f \approx 0.1$ [1–3], the thermal expansion mismatch between the γ and α_2 phases $\Delta\alpha \approx 4 \times 10^{-6} \text{ K}^{-1}$ [12–14], and creep data for the coarse-grained γ -phase TiAl (56 μm grain size) from Ref. [15] to represent the matrix creep behavior. For consistency with our experimental procedures (as described below), we take an applied stress of 35 MPa, and heating or cooling rates between 35 and 200 K/min, with 20-min cycles between 300 and 900°C. Use of these parameters in Eqs. (1) and (2) predicts strain rate enhancements due to thermal-cycling creep of $\dot{\epsilon} \approx 2 \times 10^{-8}/\text{s}$. This result is in qualitative agreement with the observations of thermal-mismatch-induced dislocation motion in these alloys reported by Appel and Wanger [7]. Furthermore, the above predicted rate may become significant when considering the expected lifetimes of engine components. As mentioned before, Eq. (1) has been formulated for spherical reinforcement, while the α_2 phase commonly exists in the form of lamellae in creep-resistant γ -TiAl-based alloys. Kitazono and Sato [16] have demonstrated that the above model over-predicts the thermal-cycling creep effect by about an order of magnitude for a material containing long fibers. Thus, reasonable rates to be expected based on the model are in the range 2×10^{-9} – $2 \times 10^{-8}/\text{s}$. Additionally, the α_2 phase can be expected to creep during thermal cycling, although the model above does not account for creep of the reinforcement. This may further reduce the mismatch between γ and α_2 , diminishing the thermal-cycling creep effect.

Because γ -aluminides proposed for high-temperature usage are commonly composed of two phases, and have been reported to manifest high levels of internal stress [7], they may exhibit thermal cycling creep, leading to undesirable strain accumulation and premature failure. For example, Sato et al. [17] have observed increased creep rates during thermal cycling of Mar-M247 superalloy, which they attributed to thermal expansion

mismatch between the γ and γ' phases. Since the simple model described above offers only a broad upper-bound estimate for the strain rate during thermal-cycling creep, experiments are required to verify the presence or absence of this deformation mechanism in γ -TiAl-base alloys. Additionally, since other promising engineering materials (i.e. directionally solidified composites) have been found unstable during thermal cycling [8,18], a study of thermal cycling creep has important implications for the impending service of TiAl alloys. Thus, the purpose of the present paper is to examine the tensile deformation of two TiAl alloys during thermal cycling between 900 and 300 or 350°C. The measured deformation rates during cycling are compared with those expected from isothermal creep without internal stress effects.

2. Experimental procedures

The two TiAl alloys used in the present work were supplied in the form of hot-extruded billets by Plansee AG, Austria. The first alloy had the nominal composition Ti-47Al-4(Cr,Nb,Ta,B) (at%), and exhibited a duplex microstructure (Fig. 1a). The second alloy was of

Fig. 1. Optical micrographs of (a) Ti-47Al-4(Cr,Nb,Ta,B) with duplex (D) structure, and (b) Ti-46.5Al-4(Cr,Nb,Ta,B) with fully lamellar (L) structure.

the nominal composition Ti–46.5Al–4(Cr,Nb,Ta,B) (at%), and had a “designed fully lamellar” structure (see, for example, [2,19]) as shown in Fig. 1b. In the remainder of this paper the two alloys are referred to as “D” (duplex) and “L” (lamellar), respectively. More processing and microstructural details of these materials can be found in [2,20]. One tensile specimen of each alloy, with cylindrical gauge section of 25 mm length and 4 mm diameter, was machined from an extruded billet.

Isothermal and thermal-cycling creep experiments were conducted using a customized creep frame described elsewhere [21], under an atmosphere of purified argon. Both the D and L specimens experienced similar test schedules. After heating to 900°C and soaking for 75 min, a stress of 32.5 MPa (D) or 37.0 MPa (L) was applied for a sufficient amount of time to establish a steady-state creep rate. Following at least 5 h of steady-state creep, the specimens were subjected to thermal cycles between an upper temperature of 900°C and a lower temperature of 300°C (D) or 350°C (L). Typical 20-min thermal cycle profiles for the two alloys are shown in Fig. 2. The heating and cooling ramps were as rapid as experimentally possible, and are slightly different due to different convection and radiation conditions between the two tests. The specimens experienced 56 (D) or 45 (L) thermal cycles after which the temperature was allowed to equilibrate at 900°C, where a steady-state creep rate was again assessed for comparison with the rate measured before cycling.

3. Results and discussion

Selected portions of the deformation history of specimen D with applied stress 32.5 MPa are shown in Fig. 3,

Fig. 2. Representative experimental thermal cycles.

with the measured creep rates tabulated in Table 1. The creep rates before and after thermal cycling were both $\sim 1.0 \times 10^{-6}$ /s, indicating that no significant microstructural changes were effected by cycling, and a true steady-state was established prior to thermal cycling. Consequently, the contribution of primary creep to strain development during thermal cycling is negligible. A similar deformation history was obtained for the L specimen with applied stress 37 MPa; the measured creep rates are given in Table 1. The more creep-resistant lamellar structure of the L specimen resulted in much lower deformation rates as compared to the D specimen at a similar stress. This result is in qualitative agreement with several previous reports [22–25] for different TiAl alloys at higher stresses (250–430 MPa).

It is apparent from Fig. 3 and Table 1 that the deformation rate during cycling is much lower than the isothermal rate at 900°C, for both the D and L alloys. However, a lower creep rate may be expected during cycling, as the average cycling temperature is well below 900°C. The contribution of thermal-cycling creep to deformation is found by subtraction of the average isothermal creep rate during cycling, $\dot{\epsilon}_{\text{eff}}$, as defined by Eq. (2), from the measured thermal cycling creep rate, $\dot{\epsilon}_{\text{tot}}$. Evaluation of Eq. (2) requires knowledge of the temperature profile during a representative thermal cycle as well as the temperature dependence of the isothermal creep rate of the material. In the present case the temperature profile is measured experimentally (Fig. 2), and the isothermal creep rates of both alloys are

Fig. 3. Selected portions of the measured displacement-history of the D alloy, including isothermal creep at 900°C before and after a period of thermal cycling (11 of 56 cycles shown). Displacement scales have been shifted for convenience, and the creep rates corresponding to each segment of the experiment are shown.

Table 1

Measured creep rates at 900°C and during thermal cycling. (The creep rate predicted by Eq. (2) is listed for comparison)

Alloy	Stress (MPa)	Creep rate at 900°C before cycling (/s)	Creep rate at 900°C after cycling (/s)	Measured creep rate during cycling, $\dot{\epsilon}_{\text{cyc}}$ (/s)	Calculated creep rate during cycling, $\dot{\epsilon}_{\text{eff}}$ (/s)
Ti–47Al–4(Cr,Nb,Ta,B) “D”	32.5	9.9×10^{-7}	1.1×10^{-6}	1.4×10^{-7}	$1.32 \times 10^{-7} - 1.44 \times 10^{-7}$
Ti–46.5Al–4(Cr,Nb,Ta,B) “L”	37.0	1.8×10^{-7}	2.9×10^{-7}	2.2×10^{-8}	$8.29 \times 10^{-9} - 2.23 \times 10^{-8}$

known at 900°C (Table 1) for the respective stresses of interest. We extrapolate the measured isothermal creep rates at 900°C to lower temperatures by an Arrhenius relationship, i.e. $\dot{\epsilon}(T) \propto \exp(-Q/RT)$ with Q the apparent activation energy for creep and R the ideal gas constant, assuming no change in creep mechanism at lower temperatures.

Using the above Arrhenius relationship, we evaluate Eq. (2) by numerical integration over the experimentally-measured thermal cycle profile (Fig. 2), the results of which are shown in Table 1. The calculated values are presented as a range of strain rates, corresponding to values of Q between 250 and 450 kJ/mol, which are reported for creep of various two-phase TiAl alloys [6,22,25,26]. The calculations are only weakly sensitive to these reasonable variations in Q .

For both the D and L alloys, the measured creep rates during thermal cycling are fully predicted from Eq. (2), as shown in Table 1. We conclude that the measured deformation during thermal cycling is most probably due to typical creep mechanisms (e.g. dislocation creep, diffusional creep, etc.), with no measurable enhancement due to the contribution of internal stress. Based upon data from Table 1, errors in the experimental strain rates ($\pm 5\%$), and errors in the evaluation of Eq. (2), we estimate an upper-bound strain rate from the contribution of thermal-cycling creep of $\sim 1.4 \times 10^{-8}/\text{s}$. This rate is close to the broad upper-bound calculated from Eq. (1) above. It corresponds to a strain of 1.6×10^{-5} accumulated after each thermal cycle, and implies that, in the worst case, over 600 thermal cycles can be performed before 1% strain is accrued due to this mechanism.

In addition, we note that the relative increase of creep rate due to thermal cycling is expected to contribute significantly to deformation only under the action of small applied stresses [9]. Thus, the present experimental conditions were chosen to represent extreme conditions, i.e. high temperatures, large thermal cycle amplitudes, and low stresses, to maximize the effect of thermal cycling. In practice, lower maximum temperatures (650–750°C), smaller temperature excursions in the creeping range of the alloy, and larger applied stresses (up to 250 MPa) would further diminish the effects of thermal cycling creep relative to typical creep mechanisms. Further long-term experiments (with several thousand

thermal cycles) would be necessary to establish the existence and magnitude of this effect under service-relevant conditions.

4. Conclusions

Prior microstructural evidence on γ -TiAl-base alloys suggests the possibility of large thermal-expansion mismatch stresses on heating into the range of service temperatures. Upon thermal cycling, then, such alloys could exhibit enhanced creep due to internal stresses as also observed in many two-phase materials. While simple models of thermal-cycling creep predict a small but significant effect, the microstructural complexity of all two-phase TiAl alloys makes such predictions unreliable. Thus, an experimental investigation of two TiAl alloys, with duplex and “designed fully lamellar” microstructures, was conducted under thermal-cycling conditions to elucidate this point.

Isothermal creep rates at 900°C before and after cycling were measured and compared to rates during thermal cycling between 900 and 300 or 350°C. Within experimental errors of $\sim 1.6 \times 10^{-3}\%$ strain per cycle, isothermal creep was found to account for all of the strain accumulated during thermal cycling, with no measurable enhancement due to internal mismatch-stresses. Given that experimental conditions were chosen to maximize the thermal-cycling effect, the contribution of thermal-cycling creep can be expected to be further diminished for technologically representative service conditions.

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