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## Research Article

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**Posted Date:** September 9th, 2021

**DOI:** <https://doi.org/10.21203/rs.3.rs-885396/v1>

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# $\gamma$ -TiAl alloy: Tensile creep deformation behaviour and creep life at 832 °C

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## Abstract

Creep deformation in single phase  $\gamma$ -TiAl alloy manufactured using different processing techniques has been an extensively studied topic since the late 1970s. The present work revisits the original work on understanding the tensile creep deformation behaviour of wrought single-phase  $\gamma$ -TiAl alloy by Hayes and Martin [1] and is aimed to develop an understanding of steady state creep. Besides, it is also aimed to investigate the creep life for stress levels of 69.4 and 103.4 MPa at 832 °C using Monkman-Grant [2] approach.

**Keywords:**  $\gamma$ -TiAl, Monkman-Grant parameter, Coble creep, Nabarro-Herring creep

## 1. Introduction

A series of reports on the room temperature tensile deformation behaviour of  $\gamma$ -TiAl alloys ( $L1_0$  structure) show that the near- $\gamma$ , two phase compositions having ~ 48 at.% Al possess very high toughness (ultimate tensile strength~ 844-1010 MPa and tensile ductility~ 3-4.6% at room temperature) [1], [3]–[10]. Extensive creep deformation studies have been carried out on a number of two phase near-  $\gamma$ -TiAl alloys produced by various processing routes. In addition, compression creep based investigations have been performed on a number of single phase  $\gamma$ -TiAl alloys and a number of literatures have reported that minimum strain rates during creep testing at different regimes of temperature and stress, hugely influence the grain size of single phase  $\gamma$ -TiAl alloys (during creep testing) [11]–[24]. Minimum strain rate of such intermetallic-based alloys may be defined in terms of Mukherjee-Bird-Dorn (MBD) equation [19], [25], [26].

The work of Hayes and Martin [1] discusses an analysis of the minimum strain rate deformation of a wrought single phase  $\gamma$  -TiAl alloy within temperature range of 760-1000 °C and stress range of 32–345 MPa. The work by Hayes and Martin [1] also predicts the main mechanism for creep rupture at different temperatures, in a given stress range, using Larson-Miller (L-M)

[27] and Monkman-Grant (M-G) plots [2] and extensive microstructural characterisation using Optical microscope (OM) and the Transmission Electron Microscope (TEM). Besides, a recent report (from the author) on the tensile creep deformation behaviour of  $\gamma$ -TiAl has determined the stress exponent and creep activation energies between 760 °C and 900 °C at 69.4 and 103.4 MPa [28]. Moreover, based on creep activation energies and stress exponents, it has been reported that there is a transition from dislocation-glide to dislocation-climb controlled creep at very low creep stress levels ( $\sim 66.68$  MPa) [28] and that there is no steady-state creep observed for both interrupted and uninterrupted creep tests at 832 °C [1]. However, there is limited understanding of creep behaviour and creep rupture life as a function of stress at 832 °C. To this end, the present work is aimed towards determining creep behaviour and rupture life for stress levels of 69.4 and 103.4 MPa at 832 °C using M-G approach.

## 2. Theoretical analysis

Based on the single-phase  $\gamma$ -TiAl alloy composition and tensile creep test parameters used by Hayes and Martin [1], the minimum strain rate ( $\dot{\epsilon}_{\min}$ ) vs strain ( $\epsilon$ ) and  $\ln(\dot{\epsilon}_{\min})$  vs  $\epsilon$  plots have been determined at 832 °C for two different stress levels viz. 69.4 and 103.4 MPa for the purpose of understanding the influence of creep stress and temperature on the three creep regimes (viz. primary, secondary and tertiary) and the influence of the same on the creep life of the material. For creep life determination, M-G approach was used. For M-G approach, the governing equation is: [16], [29]

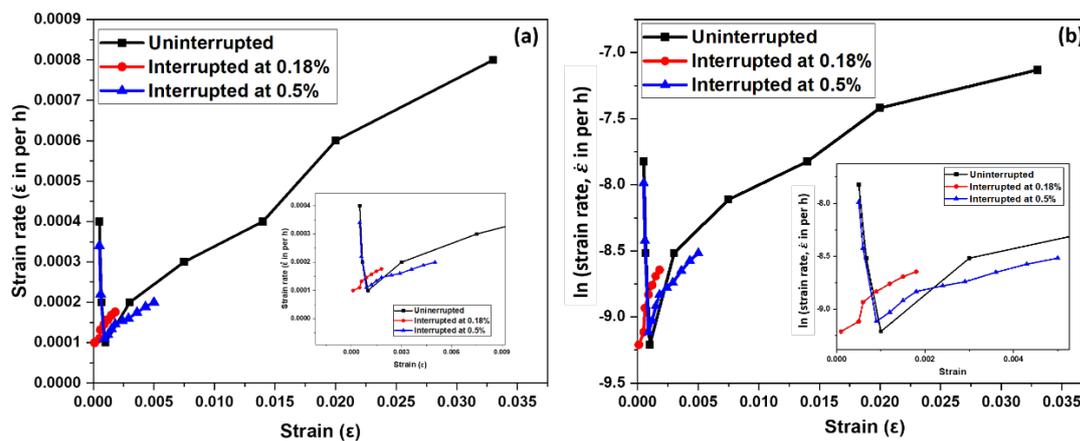
$$\dot{\epsilon}_s t_r = C_{MG} \quad (2)$$

where,  $\dot{\epsilon}_s$  is the steady state creep rate ( $\sim \dot{\epsilon}_{\min}$  based on the justification that many metallic materials may not exhibit steady-state creep (provided in Ref. [28])),  $t_r$  is the rupture time (expressed in seconds) and  $C_{MG}$  ( $\sim 1.40$  at 832 °C) is the M-G constant (obtained from Ref. [28]).

## 3. Results and Discussions

**Figs. 1(a and b)** show the variation of  $\dot{\epsilon}$  with  $\epsilon$  (in both linear and logarithmic scales respectively) for three cases viz. uninterrupted creep testing and interrupted creep testing with termination strains of 0.18 and 0.5%. **Fig. 1(b)** shows the creep regimes at 832 °C (for stress levels of 69.4 and 103.4 MPa respectively) with a higher simplicity. Based on **Fig. 1**, it is observed that the present material (45.9Ti-0.91Nb-52.9Al (at.%)) shows a very early onset of tertiary creep at 832 °C for uninterrupted creep test and for creep test terminated at 0.5% strain

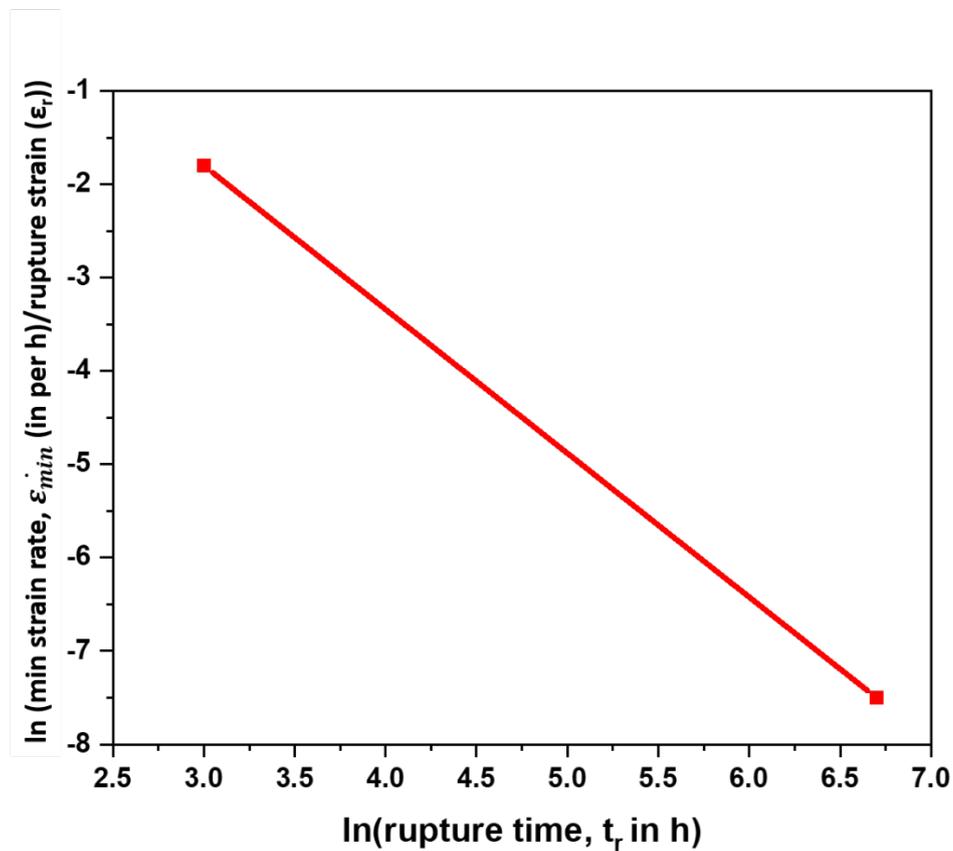
whereas there is no primary (or transient) creep observed during creep test interrupted at strain of 0.18% (**Fig. 1**) indicating that the material (used in the present work) does not exhibit resistance to creep deformation for the aforementioned deformation condition. Besides there is a limited regime of tertiary creep observed for creep tests terminated a 0.18% and 0.5% strain (**Fig. 1**). In addition, the material shows a dominant tertiary creep regime for the case of uninterrupted tensile creep testing (**Fig. 1**). Secondary (or steady-state creep) is not observed in any case (**Fig. 1**). In other words, the interrupted creep tests lead to a higher rate of vacancy formation at grain boundaries (GBs) normal to the tensile stress subsequently followed by void growth and coalescence leading to an intergranular fracture, as mentioned in Refs. [30]–[32]. Explanation of the above tendencies during interrupted and uninterrupted tensile creep test (at 832 °C) is subject to extensive microstructural investigations which is beyond the scope of the present discussion.



**Fig. 1** Variation of (a)  $\dot{\epsilon}$  with  $\epsilon$ , and (b)  $\ln(\dot{\epsilon})$  with  $\epsilon$ . In parts (a) and (b), the plot uninterrupted creep test has been represented using black colour whereas the creep tests terminated at 0.18% and 0.5% have been represented using red and blue colours respectively. The inset images in parts in (a) and (b) represent the different creep regimes in both interrupted and uninterrupted creep tests with a high level of clarity.

For uninterrupted creep testing, there is a dominant tertiary creep regime (**Fig. 1**). Moreover, it has been reported that the extent of tertiary creep hugely influences the creep life of the material [31]. To this end, creep life determination has been performed for samples (with uninterrupted creep testing) at 832 °C. **Fig. 2** shows the creep life plots (for uninterrupted creep testing) based on M-G approach for stress levels of 69.4 and 103.4 MPa at 832 °C. From **Fig. 2**, it is observed that  $t_r$  decreases with increasing  $\dot{\epsilon}_{min}$  at 832 °C (based on equation 2). Moreover, the slope (p)

of the M-G curve (in **Fig. 2**) is calculated as -1.33 which suggests that the mechanism of creep rupture is power-law breakdown at 832 °C (using the criteria mentioned in Ref. [14], [16], [33]–[37]). Hayes and Martin [1] and Saha [28] have reported that  $\dot{\epsilon}_{min}$  decreases with increasing stress levels (from 69.4-103.4 MPa) at 832 °C. Moreover, Saha [28] has reported that the creep mechanisms operating at different stress levels (69.4 and 103.4 MPa) at 832 °C are independent of each other and hence, are not sequential. Hence, combining the trend observed between  $\dot{\epsilon}_{min}$  with  $t_r$  (**Fig. 2**) with the aforementioned reports (Refs. [1], [28]), it may be inferred that M-G based approach to determine the creep life also predicts that  $t_r$  decreases with increasing stress levels (from 69.4 to 103.4 MPa) at 832 °C.



**Fig. 2** M-G plots for determining creep life for 69.4 MPa and 103.4 MPa at 832°C.  $\dot{\epsilon}_{min}$  has been normalised with rupture strain ( $\epsilon_r$ ) determined from **Fig. 1**. A justification of the above normalisation (for M-G plots) has been provided in Ref. [27].

Based on the Ashby's model [13], [14], [38], [39], the constitutive equation for NH (or lattice diffusion creep) is given as:

$$\dot{\epsilon} = 9.3 \frac{D_l G b}{kT} \left(\frac{b}{d}\right)^2 \left(\frac{\sigma}{G}\right) \quad (3)$$

where,  $\dot{\epsilon}$  is the strain rate,  $D_l$  is the lattice diffusivity,  $b$  is the burgers vector,  $d$  is the grain size,  $k$  is the Boltzmann constant,  $\sigma$  is the creep stress and  $G$  is the shear modulus of the material. Similarly, using Ashby's model, the constitutive equation for Coble creep (or GB diffusion creep) is given as:

$$\dot{\epsilon} = 33.4 \frac{D_{GB} G b}{kT} \left(\frac{b}{d}\right)^3 \left(\frac{\sigma}{G}\right) \left(\frac{\delta}{b}\right) \quad (4)$$

where,  $\delta$  is the GB width and  $D_{GB}$  is the activation energy for GB diffusion. Both  $D_{GB}$  and  $D_l$  are highly temperature-dependent [14]. However, the magnitude of  $D_{GB}$  is higher than that of  $D_l$  at lower homologous temperatures whereas the magnitude of  $D_l$  is higher than that of  $D_{GB}$  at higher homologous temperatures [40], [41]. This is because GBs offer a higher pathway for diffusion at lower temperatures when compared with that of the lattice [42]. Moreover, a comparison of the equations (3) and (4) shows that Coble creep has a higher grain-size dependence as compared to that of NH creep. In other words, although both Coble and NH creep may operate simultaneously, Coble creep is the dominant diffusion creep mechanism for fine grain sizes at lower temperatures whereas NH creep dominates for comparatively coarser grains at higher temperatures. However, both these mechanisms show a similar sensitivity towards  $\sigma$  (as shown in equations (3) and (4)). Based on the previous reports [1], [28],  $d$  has minimal influence on  $\dot{\epsilon}$ . The direct proportionality between  $\dot{\epsilon}$  and  $\sigma$  in equations (3) and (4) also explains the previous observations (in Refs. [1], [3], [28]) on the increasing magnitude of  $\dot{\epsilon}$  with increasing stress levels from 69.4-103.4 MPa. Besides,  $G$ ,  $\sigma$ ,  $T$  ( $\sim 832$  °C) and  $b$  are constant in the present work. Based on the values of NH and Coble creep activation energies ( $Q_l$  and  $Q_{GB}$  respectively) reported for 69.4 MPa ( $Q_{GB} \sim 210$  kJ/mol and  $Q_l \sim 350$  kJ/mol at 832 °C in Ref. [28]) and for 103.4 MPa ( $Q_{GB} \sim 151.88$  kJ/mol and  $Q_l \sim 253.13$  kJ/mol at 832 °C in Ref. [28]), it may be inferred that a decrease in the creep activation energy (for both Coble and NH creep) significantly reduces the creep life of the material with increasing stress levels from 69.4-103.4 MPa.

#### 4. Conclusions

Based on the present work, the following may be concluded:

- At 832 °C,  $\gamma$ -TiAl alloy (45.9Ti-0.91Nb-52.9Al (at.%)) shows a dominant tertiary creep regime for uninterrupted creep testing at 69.4 MPa and 103.4 MPa whereas there is a limited tertiary creep regime for creep tests (under the same temperature and stresses) for creep tests interrupted at 0.18% and 0.5% indicating that during interrupted creep testing,

there is a higher rate of vacancy formation along GBs normal to the applied tensile stress followed by void coalescence and growth leading to intergranular fracture.

- M-G approach (for creep life determination) shows that creep life (for uninterrupted creep testing) decreases with increase in stress levels from 69.4 to 103.4 MPa. In addition, M-G plots indicate the mode of rupture (for uninterrupted creep test) at 832 °C (for stress levels of 69.4 MPa and 103.4 MPa) is power-law breakdown.
- A decrease in creep activation energy (for both Coble and NH creep) reduces the creep life of the material with increasing stress level from 69.4 to 103.4 MPa.

### Acknowledgement

MS would like to acknowledge the Department of Metallurgical and Materials Engineering, NIT Durgapur, for detailed discussions during the preparation of the manuscript.

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