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Effect of trace impurities on the creep behavior of a near α titanium alloy

H. Mishra, D.V.V. Satyanarayana, T.K. Nandy* and P.K. Sagar

Defence Metallurgical Research Laboratory, Hyderabad 500 058, India

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The creep behavior of a near α titanium alloy was evaluated for the $\alpha + \beta$ and β heat-treatment conditions at two different levels of iron and nickel. Stress exponents varied from 4.6 to 7.9, suggesting that dislocation climb was the rate-controlling mechanism. The alloy containing lower iron plus nickel content exhibited superior creep resistance. The results were rationalized in terms of activation energy in the framework of dislocation creep.

Keywords: Titanium alloys; Optical microstructures; Creep; Activation analysis; Trace impurities

A major factor responsible for limiting the use of titanium alloys up to a temperature of 600 °C is their poor creep resistance, which obviously stems from the relatively low-temperature $\alpha \rightarrow \beta$ phase transformation. The high-temperature β (body-centered cubic) phase with an open structure exhibits anomalously high diffusion rates leading to accelerated unraveling of dislocations at elevated temperatures, thereby rendering the alloy unsuitable for applications that require good creep resistance. This problem can be partly mitigated by raising the transformation temperature. Based on this premise, several near α titanium alloys such as TIMETAL®685 (Ti–6%Al–5%Zr–0.5%Mo–0.25%Si), TIMETAL®6242S (Ti–6%Al–2%Sn–4%Zr–2%Mo–0.08%Si) and TIMETAL®834 (Ti–5.8%Al–4%Sn–3.5%Zr–0.7%Nb–0.5%Mo–0.35%Si–0.06%C), being the last in the series, have been developed for high-temperature applications [1]. While TIMETAL®6242S is used in both β treated (transformed β structure) and $\alpha + \beta$ treated (transformed $\beta + 30$–40% primary α) conditions, TIMETAL®685 is used only in the β treated condition. During β heat treatment, since the transformation temperature is high, the grain growth of β phase is quite rapid, especially in the absence of any inhibiting factor. In this microstructural condition, though the alloy exhibits good creep resistance, it has lower ductility due to its large prior β grain size, resulting in poorer low-cycle fatigue (LCF) properties. This has been addressed by developing TIMETAL®834 alloy, which is used in the duplex $\alpha + \beta$ heat-treated condition. Here, the presence of primary α (10–20%) at β grain boundaries impedes rapid growth of β grains, resulting in a tradeoff between creep resistance and LCF properties. Despite the optimization of creep and fatigue, the creep resistance in the $\alpha + \beta$ heat-treatment condition is relatively inferior due to the presence of the primary α [2–7]. Additionally, the presence of trace impurities such as Fe and Ni are known to impair the creep resistance of high-temperature titanium alloys [3,4,8–11]. Therefore, any further improvement in creep properties demands stringent control of these impurities.

Deterioration of creep resistance in titanium alloys with higher levels of trace impurities has been attributed to accelerated diffusion in the Ti-rich α phase. The group IVB elements such as Zr and Ti are considered open metals due to their unusually large ratios of ionic to atomic radii. Fe, Ni and Co, which normally occupy substitutional positions, are able to dissolve interstitially in the group IVB elements because of the large size difference. These elements (Fe, Ni, Co) diffuse at a faster rate through the host lattice (Table 1) which leads to acceleration of lattice self-diffusion [12,13]. This leads to deterioration of creep resistance in titanium alloys as reported by Hayes et al. [10] in a near α alloy, TIMETAL®6242S. In an earlier study [14], we demonstrated the deleterious effect of iron and nickel on the creep resistance of TIMETAL®834 for different microstructural conditions. In the present work, an attempt has been made to provide a rationale for the results by

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* Corresponding author. E-mail: nandytapash@gmail.com

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Table 1. Diffusivity of iron and nickel in α titanium and self-diffusion data of nominally pure and ultrahigh-purity titanium [12,13]

Element	D_o ($m^2\ s^{-1}$)	Q ($kJ\ mol^{-1}$)	D at 873 K ($m^2\ s^{-1}$)
Fe, $_{\perp}c$	4.7×10^{-7}	112.3	9.0×10^{-14}
Fe, $_{\parallel}c$	6.4×10^{-6}	144.2	1.5×10^{-14}
Ni, $_{\perp}c$	5.6×10^{-6}	137.2	3.4×10^{-14}
Ni, $_{\parallel}c$	5.4×10^{-6}	141.8	1.8×10^{-14}
Ti (nominally pure)	1.7×10^{-8}	191.9	5.5×10^{-20}
Ti (ultrahigh-purity), $\perp c$	1.35×10^{-3}	303.0	1.1×10^{-21}

$_{\perp}c$, $_{\parallel}c$ indicate perpendicular and parallel to c-axis respectively.

evaluating the creep behavior of the alloys containing different amounts of iron plus nickel.

A high-purity alloy (designated TIMET834) containing lower levels of Fe and Ni was procured from TIMET, UK and an alloy (designated DMR834HF) containing higher levels of Fe and Ni was produced in-house. The alloy was double-melted in a 40 kg vacuum arc-melting furnace using a standard melting procedure. The chemical compositions are shown in Table 2.

Both the alloys were forged at 1100 °C from a 140 mm starting diameter to a 65 mm square cross-section. The β transus was determined by solution treating specimens at different temperatures, followed by quenching in water. The specimens were polished, etched and observed by optical microscopy at 100× magnification. The lowest temperature at which primary or equiaxed α was observed was taken as the transus temperature (β_T). The two alloys were then rolled at β_T-90 °C to 14 mm diameter bars and subsequently given standard heat treatments to obtain two types of structures. The heat-treatment schedules are given in Table 3. The first treatment gave a fully transformed β microstructure in both TIMET834 and DMR834HF, and the second treatment resulted in a duplex microstructure comprising 10–15% equiaxed or primary α in transformed β matrix. Typical microstructures in one of the two alloys for both treatments are shown in Figure 1.

Creep testing was carried out in a range of temperatures (873–973 K) and stresses (150–500 MPa). In order to reduce the number of specimens, the stress jump technique was used. The specimen was allowed to deform long enough to attain steady-state creep before jumping to next stress level.

Figure 2a shows typical creep curves of TIMET834 at 600 °C at two stress levels for both microstructural conditions. Steady-state strain rates were extracted from the creep curves for all combinations of stress and temperature and plotted against applied stress on double-logarithmic scales (Figs. 2b and 3). The slopes of these lines, which give the values of the stress exponent (n), vary from 4.7 to 7.8. Arrhenius plots of $\log \dot{\varepsilon}$ vs. $1/T$ (K) at constant stress (150 MPa), which was used to determine the activation energy (Q), are shown in Figure 4a and b. Values of stress exponent and activation energy are listed in Table 4.

The β transus for DMR834HF is about 1070 °C compared to 1055 °C for TIMET834. This is consistent with higher aluminum equivalent (=Al + Sn/3 + Zr/4 + 10[O])[15] of DMR834HF (Table 2). Additionally, the amount of β stabilizer (Mo + Nb) is marginally lower in DMR834HF. Therefore, the ST temperature for both β and α + β treatment in the former alloy are kept

Table 2. Chemical composition of different melts

Melt	Al	Sn	Zr	Nb	Mo	Si	C	O	N	Fe	Ni	Al eq.
DMR834HF	6.04	3.98	3.20	0.60	0.50	0.31	0.060	0.130	0.0100	0.024	0.0170	9.5
TIMET834	5.75	4.02	3.54	0.71	0.50	0.31	0.065	0.097	0.0012	0.0065	0.0020	8.5

Table 3. Heat-treatment details of the as-processed bars

Treatment	Volume fraction of primary α (vol.%)	DMR834HF (AC = air cool)	TIMETAL834 (AC = air cool)
β	0	1080 °C/2 h/AC + Age	1065 °C/2 h/AC + Age
α–β	15	1040 °C/2 h/AC + Age	1025 °C/2 h/AC + Age

Age: 700 °C/2 h/AC.

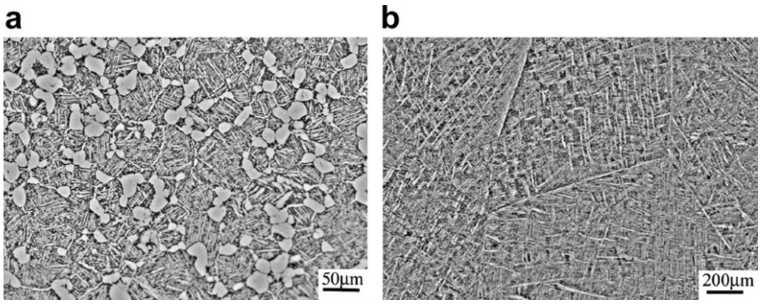

Figure 1. Microstructure of TIMET834 after (a) the α + β heat-treatment (1025 °C/2 h/AC + 700 °C/2 h/AC) condition, and (b) the β heat-treatment (1065/2 h/AC + 700/2 h//AC) condition.