Creep behavior of Ti-600 alloy solutioned at β and α+β region

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Abstract. Creep tests were carried out on one kind of near alpha titanium alloy named after Ti-600 at 550 ℃, 600 ℃ and 650 ℃ with the same stress of 300 MPa. The calculated activation energies for the alloy solutioned atβand α＋β region are 490.1kJ/mol and 473.5 kJ/mol, respectively. Transformed β grains become coarsened and twisted, and the short primary α grains grow up, α plus β lamellar in transformed β grains elongate apparently after creep tests.

Introduction

In recent years, more and more attention has been paid on titanium and titanium alloys which can be used at 600 ℃ or even higher temperature for a long time[1,2]. Several near α titanium alloys have been developed to meet the requirements for advanced aerospace industry for increased performances and higher service temperatures, including IMI834, Ti-1100, BT36, Ti60, Ti-600, etc[3,4]. Ti-600, developed at Northwest Institute for Nonferrous Metal Research (NIN) in China, is a near alpha titanium alloy designed for components used in turbine engines at 600 ℃. Tensile properties at ambient temperature and at its service temperature, thermal stabilities, oxidation properties, and fatigue properties of the alloy have been investigated widely by several researchers[3,5,6].

A major factor responsible for limiting the use of titanium alloys up to 600 ℃ is their poor creep resistance. So, the high temperature creep property of conventional high temperature titanium alloys should be studied profoundly, for the creep property has significant influence on the life endurance for Ti alloys and the safety for key components. Creep resistance has great relation with the volume fraction of β transus phases. In this paper, the role of both β and α+β microstructures on creep for Ti-600 alloy at three temperatures with the same stress of 300MPa are to be investigated.

Experimental materials and procedures

A 200kg ingot of Ti-600 alloy was produced by electrode consumption vacuum arc furnace. The β transus temperature for the alloy is 1010 ℃ or so. The ingot alloy was conventionally forged and rolled to diameter 12mm bars. The creep samples were cut from the rolling bars and were solutioned at 1000 ℃ and 1020 ℃ for 1 h, then aged at 650 ℃ for 8 h, air cooling.

Creep tests under constant tensile load in air were carried out on the specimens of 5mm gauge diameter using a RD-30 typed creep-rupture machine at 550 ℃, 600 ℃ and 650 ℃ with the same stress of 300 MPa. The creep elongations were measured by means of one linear variable differential transformer, allowing an accuracy of 0.1%. Microstructures for the specimens before and after creep tests were examined by Olympus PMG3 optical microscopy.

Experimental materials and procedures

Creep properties of Ti-600 alloy. Representative creep curves of Ti-600 alloy with two different thermal treatments are displayed in Fig.1. The creep curves exhibit all the three well-defined stages, i.e. primary, secondary and tertiary creep regimes. In the initial stage, or primary creep, the strain rate is relatively high, but slows with increasing strain. The strain rate eventually reaches a minimum and becomes near constant. This stage is known as secondary or steady-state creep. In tertiary creep, the strain rate exponentially increases with strain because of necking phenomena. An accentuated reduction of primary creep period with increase of the applied stress and test temperatures is also observed in Fig.1.

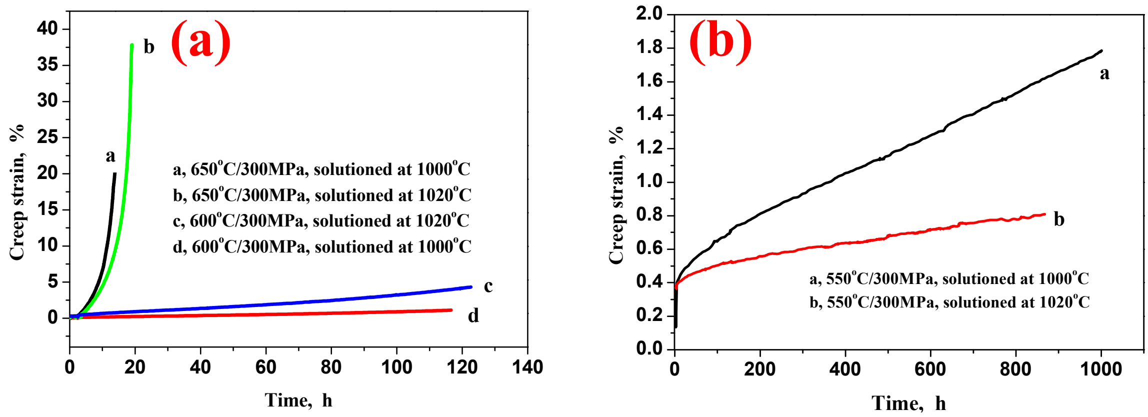

Fig.1. Typical creep curves of Ti-600 alloy, samples solutioned at 1020 ℃ and 1000 ℃, then crept at 600 ℃, 650 ℃ (a) and at 550 ℃ (b) with the same stress of 300 MPa.

Activation energy of Ti-600 alloy. The combined stress and temperature dependence of steady state creep rate is frequently described by a power law of the form[7]:

$$\dot{\varepsilon}_s = A\sigma^n \exp\left(-\frac{Q_{App}}{RT}\right) \tag{1}$$

where A is a constant, n the apparent stress exponent, R the gas constant, T the absolute temperature and Q_{App} is the apparent activation energy. When the stress is stable, the apparent activation energy can be described as follows through differential calculus to Equation (1):

$$Q_{App} = -R\left[\frac{\partial \ln \varepsilon_s}{\partial \ln(1/T)}\right]_\sigma \tag{2}$$

Arrhenius plots of log $\dot{\varepsilon}_s$ vs. 1/T (K) at constant stress of 300 MPa, which was used to determine the activation energy, are shown in Figure 2. The activation energy for the alloy solutioned at β and α+β region is evaluated to be 490.1kJ/mol and 473.5 kJ/mol, respectively. From the result above, it can be seen that for the alloy solutioned at β region exhibit higher activation energy and show better creep resistance with β treated structures as compared to the alloy solutioned at α+β region by about 17kJ/mol, which consists with the established fact in titanium alloys[8].

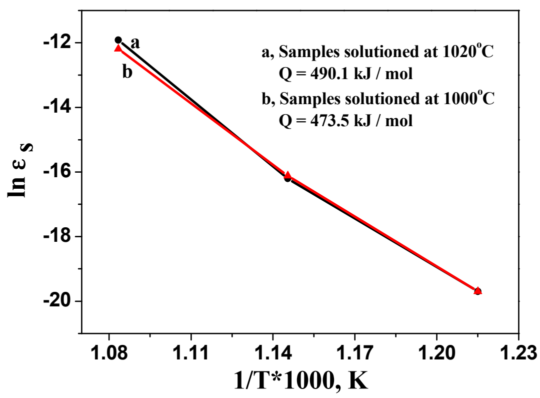

Fig.2. Arrhenius plots for the determination of activation energy.

Optical microstructures Ti-600 alloy. The typical microstructures for the alloy before and after creep tests are shown in Fig.3. A duplex microstructure comprising less than 5% equiaxed or primary α in transformed β matrix can be seen in the alloy solutioned at β region. 1-3 μm wide primary α lamellas (hcp structure) are embedded in a partially transformed β matrix (bcc structure). The α laths are either tangled in basketweave zones or formed 50-100 μm wide aligned colonies inside 50-200 μm wide former β grains, as shown in Fig.3a.

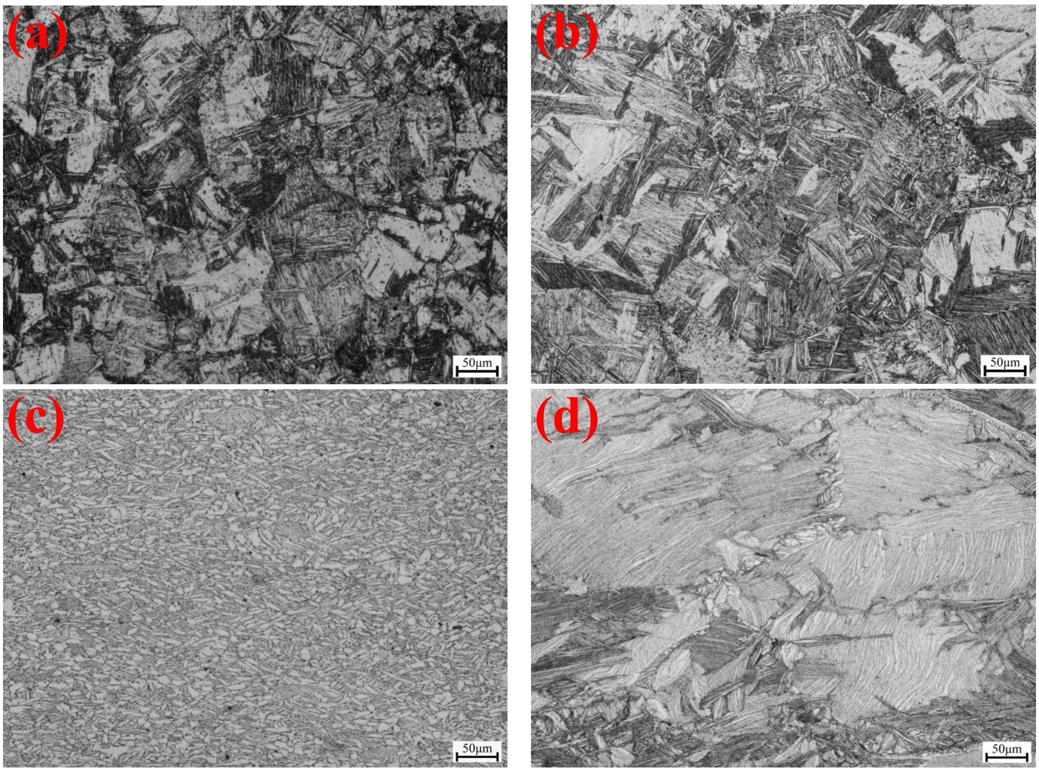

Fig.3. Microstructures of Ti-600 alloy bars, samples were solutioned at 1000 ℃ (a, c) and at 1020 ℃ (b, d) for 1h, then aged at 650℃ for 8h, air cooling, (a, b) before creep tests, (c, d) samples crept at 650 ℃ with the same stress of 300MPa.

When the solution temperature increases to above the β transus temperature (e.g. 1020 ℃), evenly distributed transformed β structures can be found in the alloy, and the microstructure for the alloy is fine lamellar α plus β, as shown in Fig.3b. The origin β grains of the alloy in Fig. 3b are bigger than that of in Fig.3a, the relatively thicker laths can be found in Fig. 3b, the reason may be the origin β grains are bigger when the alloy solutioned at relatively higher temperature of 1020 ℃.

The morphologies of the alloy after creep experiments are quite different from that of the alloy before creep tests. Transformed β grains become coarsened and twisted at the same time. The short primary α grains grow up, α plus β lamellar in transformed β grains elongate apparently, as shown in Fig.3c and Fig.3d.

Creep property comparison of Ti-600 alloy with $\alpha+\beta$ and β microstructures. Ideally, "model materials" with either a fully basketweave microstructure or aligned colonies would be better for the creep resistance. During the heat treatment process at β region, since the transformation temperature is high, the grain growth of β phase is quite rapid, especially in the absence of any inhibiting factor. The high-temperature β phase (body-centered cubic) 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.

It is well known that (α+β) morphology has a strong influence in determining the creep properties in titanium alloys[9]. Apparently, the lamellar α/β interfaces in the alloy act as obstacles to dislocation motion. The relatively large initial average grain size decreases the creep strain due to reduction of grain boundary sliding, dislocation sources[7]. There exists a coplanar orientation relationship between α and β plates in the lamellar colony structure[7]. This crystallographic relationship enables slip transfer across α/β interfaces. The abrupt orientation change across a colony boundary would be expected to impede the slip transmission[10]. Thus, the nature of the colony boundaries and the slip system active in the α-phase determine the degree of resistance of the α/β interface to the process of slip transmission. Therefore, in order for dislocation transmission occur, large residual dislocation content will be generated in the interfaces and could lead to significant strain hardening and consequently to improve the creep resistance[7].

Considering the analyses elaborated by H. Mishra[8], in β treated structure, easy diffusion paths are scarce because the prior β grain size is large and equiaxed α is absent. Thus the structure that is dominated by semicoherent or coherent lath α–α or α–β interfaces does not facilitate easy diffusion, resulting in higher activation energy. Transgranular strength and grain boundary strength will both decrease at high temperature, but the strength will decrease faster at the grain boundaries, so the transgranular strength has the main effect on the alloy.

Summary

Creep behavior of Ti-600 alloy with two different thermal treatments was investigated at the temperature range of 550 ℃ - 650 ℃ with the same stress of 300 MPa. The main conclusions can be drawn as follows:

(1) The creep resistance for the alloy solutioned at β region is better than that of the alloy solutioned at α+β region. The activation energy for alloy solutioned at β and α+β region is evaluated to be 490.1kJ/mol and 473.5 kJ/mol, respectively.

(2) Transformed β grains become coarsened and twisted, and the short primary α grains grow up, α plus β lamellar in transformed β grains elongate apparently after creep tests.

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