Materials and Design 37 (2012) 236–242

Contents lists available at SciVerse ScienceDirect

Materials and Design

journal homepage: www.elsevier.com/locate/matdes

Creep behavior of a single crystal nickel-based superalloy containing 4.2% Re

Sugui Tian *, Yong Su, Benjiang Qian, Xingfu Yu, Fushun Liang, Anan Li

School of Materials Science and Engineering, Shenyang University of Technology, Shenyang 110178, China

ARTICLE INFO

Article history:
Received 26 July 2011
Accepted 5 January 2012
Available online 14 January 2012

Keywords:
C. Rapid solidification
E. Creep
F. Microstructure

ABSTRACT

By means of the microstructure observation and measurement of creep properties, an investigation has been made into the high-temperature creep behavior of a single crystal nickel-based superalloy containing Re. Results show that, compared to the 2%-Re alloy, the 4.2%-Re superalloy possesses a better creep resistance in the temperature ranges of 1040–1100 °C. The apparent activation energy and stress exponent of the alloy during steady state creep are measured to be $Q = 483.5$ kJ/mol and $n = 4.7$ in the experimental temperature and stress ranges. The dislocation climbing over the rafted γ′ phase dominates the creep process of the alloy during steady state creep. In the later stage of creep, the deformation mechanism of the alloy is that the super-dislocations with $[0\bar{1}1]$ and $[011]$ trace directions shear into the rafted γ′ phase. The main/secondary slipping dislocations are alternately activated to twist the rafted γ′ phase up to the creep fracture, which is thought to be the fracture mechanism of the alloy during creep.

1. Introduction

Single crystal nickel-based superalloys have been widely used to make the blade parts of the advanced aeroengines for their high volume fraction of γ′ strengthening phase and the resulting good creep resistance at high temperatures [1–3]. With the increase of service performance, such as the aeroengine power and thermal efficiency, the mechanical and creep properties of superalloys at high temperatures need to be further improved [4–6]. Adding refractory element like Re, W, or Mo is expected to slow down the diffusion processes in superalloys due to the solid solution strengthening effects of these elements in the material, of which Re has been found to be particularly beneficial element [7]. However, it is still unclear how Re improves the mechanical performances of nickel-based superalloys. Pyczak et al. [8] have shown that at a given temperature, the element Re can impede the undirectional coarsening of γ′ phase in alloy in the absence of an applied external stress, but promote the process of directional coarsening during deformation. As for the distribution of Re in superalloys, which is thought to be of critical significance to the mechanical properties of them, some researchers believe that Re clusters are distributed in the γ matrix phase with a size of about 1 nm [9–12], while Mottura et al. reached a different conclusion, that no clusters of Re existed in nickel-based superalloys [13]. In the literature [14], Ge et al. found that in superalloys which had not undergone creep Re was randomly distributed in the γ phase, while after creep tests Re and W were enriched in the γ phase close to the γ/γ′ interface. This enrichment was not evenly distributed parallel to the interface, but in the form of clusters. Additionally, the aggregation of Re and Mo after creep test were observed in dislocation cores near the interfaces, which will exert a pinning force for stopping the motion of dislocations during creep. However, on the other hand, the addition of Re increases the density and cost of superalloys, and may promote the precipitation of TCP (Top Closed Phase) in superalloys [15–17]. Wollmer et al. have conducted the alloy investigation in order to minimize the Re concentration on the premise of keeping the creep strength [18]. As for the dislocation configuration of superalloys during creep, Rae et al. [19] propose for instance that the reduction of the matrix SFE, due to simultaneous additions of Co and Re in TMS 75, would lower the mobility of the perfect $a/2\langle 1\,1\,0\rangle$ dislocations in the γ channels during creep at 1023 K. But Diologent et al. show that the SFE (Stacking Fault Energy) in the γ matrix of the superalloys is almost the same in all the highly alloyed Ni-based solid solution, irrespective of their precise chemistry [20]. Wilson et al. point that Re additions may play a part in the primary creep behavior of second-generation and later alloys, but are not capable of producing large primary creep strains alone [21].

The mechanical and creep properties of single crystal superalloys have close relationships with their microstructure evolution and deformation mechanisms, which have been studied by many researchers. Fedelich et al. [22] have developed a constitutive model for the mechanical behavior of single crystal superalloys at high temperatures, which relies on the slip system theory and is able to predict rafting of the γ′ phase and its influence on plastic flow, and other researchers [23–27] have also proposed different models to describe the creep features of different single crystal superalloys. Tinga et al. [28] have proposed a phenomenological relation to describe the kinetics of the microstructure evolution. Epishin et al. [29] have developed a new experimental technique (repeated

* Corresponding author. Tel.: +86 24 25494089; fax: +86 24 25496768.
E-mail address: tiansg@sut.edu.cn (S. Tian).

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doi:10.1016/j.matdes.2012.01.008

load annealing of flat wedge shaped specimens) for charactering the microstructural degradation under creep conditions. Gunturi et al. [30] believes that the anisotropic creep of single crystal nickel-based superalloy is related to the deformation mechanism and lattice rotation, while the strain rate of the alloy increases when the ⟨110⟩ interfacial dislocations, shear into the γ′ phase [31].

Even so, few literatures have studied the microstructure evolution of Re-containing single crystal nickel-based superalloys at high temperatures, and the deformation mechanism of the alloy during creep is still an open question.

In this paper, the creep behavior of a single crystal nickel-based superalloy containing 4.2% Re is investigated by the microstructure observation and creep curve measurement in certain ranges of the applied stresses and temperatures, and the deformation mechanism of the alloy during creep is briefly discussed.

2. Experimental procedure

The single crystal nickel-based superalloy with [001] orientation has been produced by means of selecting crystal method in a vacuum directional solidification furnace under the condition of a high temperature gradient. The nominal composition of the superalloy is Ni–Cr–Co–W–Mo–Al–Ta–4.2%Re (wt.%), and all samples are within 7° deviating from the [001] orientation. The heat treatment regimes of the single crystal superalloy are given as follows: 1280 °C, 2 h + 1315 °C, 4 h, A.C + 1080 °C, 4 h, A.C + 870 °C, 24 h, A.C.

After fully heat treated, the bars of the alloy were machined, along the [001] orientation, into the tensile creep samples with the cross-section of 4.5 mm × 2.5 mm and the gauge length of 20 mm. The uniaxial constant loading tensile tests were conducted in a creep testing machine of GWT504 model. The creep curves of the alloy under the different conditions were measured. The microstructure of the alloy at the different states is observed by using Scanning electron microscope (SEM) and Transmission electron microscope (TEM). And the activation energy and stress exponent of the alloy during steady state creep are also calculated.

3. Experimental results and analysis

3.1. Creep features of the alloy

The creep curve of the single crystal nickel-based superalloys containing 4.2% Re (marked by the number 1) under the applied stress of 160 MPa at 1040 °C are shown in Fig. 1, and the creep curve (marked by the number 2) of a single crystal nickel-based superalloys containing 2% Re (Different Re contents were balanced by Ni) under the same conditions is also plotted in it for the comparative analysis. It can be calculated from the curve 1 that the strain rate of the 2%-Re alloy during steady state creep is 0.021%/h and creep lifetime is 223 h. From the curve 2 the strain rate of the 4.2%-Re alloy during steady state creep is calculated to be 0.0114%/h, the lasting time of steady state is about 200 h, and the creep lifetime is measured to be 326 h. It can be concluded by comparison with the 2%-Re alloy that the 4.2%-Re alloy displays a lower strain rate during steady state creep and a better creep resistance at the same conditions.

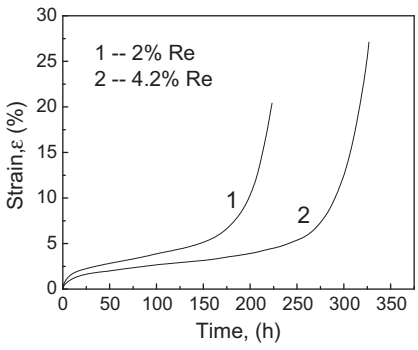

Fig. 1. Creep curves of the single crystal nickel-based superalloys containing 2% and 4.2% Re at 1040 °C/160 MPa.

The creep curves of the 4.2%-Re alloy under different conditions are measured, as shown in Fig. 2. Fig. 2a shows the creep curves of the alloy when applied different stresses at 1080 °C, from which it can be seen that the superalloy under the applied stress of 120 MPa displays a shorter initial creep stage and longer steady state creep stage of about 300 h, and the creep lifetime of the alloy is measured to be 450.9 h. As the applied stress is increased to 137 MPa, the strain rate of the alloy during steady state creep increases slightly, the creep lifetime of the alloy is measured to be about 200.6 h, and the decreased extent of the creep lifetime is about 55.4%. As the applied stress is further increased to 150 MPa, the creep lifetime of the alloy decreases remarkably to 93.3 h.

The creep curves of the alloy under the applied stress of 137 MPa at different temperatures are measured, as shown in Fig. 2b, from which it can be calculated that the strain rate of the alloy during the steady state creep at 1060 °C is 0.011%/h, the lasting time of the alloy during steady state creep is about 200 h, and the creep lifetime of the alloy is measured to be 311.7 h. As the temperature increases to 1080 °C and 1100 °C, the lifetimes of the alloy are measured to be 200.6 and 113 h respectively.

3.2. Constitutive equation and relative parameters

The transient strain of the alloy occurs when the load is applied at high temperatures. Dislocations multiply rapidly and fill in the matrix channels between the γ′ precipitates, which corresponds to the initial strain of creep. The deformation mechanism of the alloy in the initial stage of creep is only the slipping of dislocations activated in the channels of the γ matrix [32], and no dislocation shearing into the γ′ phase is detected in the alloy. As the creep goes on, the strain rate reduces as significant amount of dislocations are piled up in the matrix channels to produce the effect of the strain hardening. At the same time, the slipping, cross-slipping and climbing of dislocations occur in the action of thermal activation, which may release the stress concentration for maintaining the process of creep. Once the strain hardening and recovery softening get to equilibrium, the strain rate of the alloy during creep remains constant, and the creep enters the steady state stage. The strain rate of the alloy during steady state creep may be expressed by Dorn law given as follows [33]:

$$\dot{\varepsilon}_{ss} = A\sigma_A^n \exp\left(-\frac{Q}{RT}\right) \tag{1}$$

Where $\dot{\varepsilon}_{ss}$ is the strain rate during the steady state creep, A is the constant related to material structure, σ_A is the applied stress, n is the apparent stress exponent, R is the gas constant, T is the temperature, and Q is the apparent creep activation energy.

In the ranges of the applied temperatures and stresses, the dependences of the strain rates during steady state creep on the applied temperatures and stresses are shown in Fig. 3a and b respectively, from which the apparent creep activation energy and stress exponent of the alloy can be determined to be Q = 483.5 kJ/mol and n = 4.7 respectively. The results imply that the alloy has relatively good creep resistance in the ranges of the applied temperatures (1060–1100 °C) and stresses (120–150 MPa), and the deformation mechanism of the alloy during steady state creep is thought to be the dislocation climbing over the γ′ phase due to the stress exponent of 4.7.