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The creep deformation behavior of a single-crystal Co–Al–W-base superalloy at 900 °C

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ABSTRACT

The creep deformation behavior of a single-crystal Co–Al–W–Ni–Cr–Ta alloy with low tungsten content has been studied at stresses between 275 and 310 MPa at 900 °C. The alloy exhibits comparable creep strength with that of Co–Al–W-base alloys containing more tungsten. The creep deformation consists of three stages, the primary stage, the steady-state stage and the tertiary stage, when described by the creep strain rate versus time curve. At 900 °C, γ′ precipitates tend to raft along the direction of applied tensile stress in the steady-state creep stage and a topologically inverted and rafting γ/γ′ microstructure is formed in the tertiary stage. The main deformation mechanism in the primary creep stage is dislocation shearing of γ′ precipitates, and in the following creep stages, the dominant deformation mechanism is dislocations bypassing γ′ precipitates.

1. Introduction

The discovery of a relatively stable γ′-Co_3 (Al,W) phase with $L1_2$ structure in the ternary Co–Al–W system by Sato et al. [1] provides a possibility to develop a new class of γ′-strengthened Co-base superalloys. Solidus and liquidus of Co–Al–W-base alloys are about 100–150 °C higher than those of advanced Ni-based single-crystal superalloys [2]. Compared to Ni-base single-crystal alloys, segregation of constituent elements in the dendritic structure in single-crystals Co-base alloys is very limited [3], resulting in a higher resistance to the formation of convective instabilities and freckles. A pronounced flow stress anomaly is observed in the quaternary or ternary Co–Al–W system and their strength is much higher than that of conventional Co-base superalloys strengthened by refractory elements and carbides [4]. These studies provide further motivation to investigate the high temperature creep properties of the new Co-base alloys [5,6,8–10], which is a dominant mode of deformation for single crystal turbine blades used in gas turbine engines. The primary researches on the effect of alloying elements on the creep strength suggest that the addition of Ta or Ti to the ternary Co–Al–W system is beneficial to the creep properties [5] while the creep strength of Co–7.8Al–7.8W–1.5Ta (at%) is decreased obviously with a 4.5 at% Cr addition [5]. It seems that the addition of Cr is detrimental and the addition of Ta may not overwhelm the negative Cr effect. In a ternary Co–Al–W alloys system, the creep strength is decreased obviously with decreasing the W content from 11 at% to 8 at% [6]. In a quaternary Co–Al–W–Ta alloy system, the addition of Ni results in the depression of creep life at 1000 °C and 137 MPa [7]. In our previous work [8], microstructures and tensile properties of a multinary Co–10Al–5W–17Ni–6Cr–2.7Ta (at%) alloy have been investigated and the results show that its γ′ solvus temperature is relatively high (about 1100 °C) and the yield strength is comparable with that of Co–Al–W-base alloys containing more tungsten. It seems that alloying with high Ni and high Ta can overwhelm the negative Cr effect as well as the negative low W effect in the case of tensile tests. Since creep deformation behaviors are much more complex compared to tension, it is uncertain whether such alloying constituent is still beneficial to creep properties. Besides, until now only few TEM investigations about the creep deformation mechanisms of this new class of γ′-strengthened Co-base superalloys are available [7,9,10,11]. Therefore, the aim of the present study is focused on the creep strength and deformation behavior of the multinary Co–10Al–5W–17Ni–6Cr–2.7Ta (at%) alloy at 900 °C.

2. Experimental procedure

The master alloy with a nominal composition of Co–10Al–5W–17Ni–6Cr–2.7Ta (at%, named as 5W) was melted in a vacuum induction furnace, and then directionally solidified into [001] single crystal rods with a length of about 220 mm by HRS (Conventional High-rate-solidification Bridgeman Process) at a constant withdraw rate of 6 mm/min. Before heat treatments, the single-crystal bars were sectioned and sealed in several evacuated quartz capsules back-filled with argon. The heat treatments were carried out as

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follows: 1310 °C/10 h, furnace cooling+1000 °C/36 h, air cooling+750 °C/24 h, air cooling. In order to reveal both the dendrites and γ/γ′ microstructure, heat-treated samples were electrolyzed in a solution of 42 ml H_3PO_4+34 ml H_2SO_4+24 ml H_2O at 10 V. The γ phase is etched by this etching method. A conventional Rigaku D/MAX 2500 X-ray diffractometer was used to measure the γ/γ′ lattice mismatch. The intensity profiles of {004} reflection were collected and the peak fitting was done by using the MDI Jade software (Materials Data Inc., Livermore, CA), Pearson VII functions being used. From the lattice constants of γ and γ′, a_γ and $a_{\gamma'}$, the lattice mismatch δ was calculated by the following formula:

$$\delta = 2(a_{\gamma'} - a_\gamma)/(a_{\gamma'} + a_\gamma)$$

Tensile creep specimens with a nominal 32 mm gage length and a diameter of 5 mm were machined from the heat-treated bars. The constant-load tensile-creep tests were conducted in air at stresses between 275 and 310 MPa at 900 °C. The crept microstructures were analyzed by using a scanning electron microscope (SEM). The volume fraction of γ′ precipitates and γ channel width were analyzed by image analyzer. Some samples were run to failure, and some were interrupted for dislocation observations. Specimens were cut transversely or at about 45° angles to the tensile axis by wire electrical discharge machining. These discs were polished down to 50 μm, and then subjected to twin-jet polishing in a solution of methanol with 5 vol% perchloric acid at −30 °C and 18–20 V. A JEM 2100 transmission electron microscope (TEM) was used for dislocations analysis.

3. Experimental results

3.1. Heat-treated microstructure and γ/γ′ lattice mismatch

Fig. 1a shows the TEM bright field image of 5W alloy after heat treatments. The γ′ precipitates exhibit cuboidal morphology, aligned along the ⟨100⟩ direction. The interfaces between γ′ and γ phases are coherent, and interfacial dislocations are not observed. Tiny γ′ precipitates can be observed in channels (Fig. 1b), which are formed during cooling after the aging treatments. The frequency size distribution of γ′ precipitates as well as the γ′ volume fraction was already published in a previous work [8].

The lattice mismatch of the heat-treated sample at room temperature was determined by X-ray diffraction. Because 5W alloy consists of pure γ′ and γ phases which have nearly equal lattice parameters, the diffraction peaks of γ′ and γ phases in XRD curve overlap and form a convoluted curve. The key point of the determination of δ is the peak fitting of strongly overlapping γ′ and γ sub-peaks. The fit and experimental data are shown in Fig. 2. The lattice parameters of $a_{\gamma'}$ and a_γ is 0.35956 nm and 0.35808 nm, respectively, and the lattice mismatch δ is +0.412%. The partition coefficient of W between γ and γ′ phase can be depressed by Ni additions, resulting in a decrease of γ/γ′ lattice mismatch [12]. In view of W occupying B sites of the $L1_2$ A_3B ordered phase, the decrease of W content can decrease the lattice parameter of γ′ phase, which may cause the decrease of the γ/γ′ lattice mismatch as well. As a result, the composition characteristics of 5W results in a relatively lower γ/γ′ lattice mismatch at room temperature compared to these ternary Co–Al–W alloys [6,10]. The different thermal expansion coefficients of γ and γ′ phases and the redistribution of alloying elements between γ and γ′ phases at higher temperatures result in a change of the γ/γ′ lattice mismatch with increasing temperature. It is reported that the γ/γ′ lattice mismatch of Co–9 at%Al–9 at%W–0.1 at%B at room temperature is positive with 0.8% and it is decreased to 0.35% at 850 °C and 0.1% at 900 °C [6,10]. As for Co–9 at%Al–7 at%W–2 at%Ta–20 at%Ni, the lattice misfit at 1000 °C is lower than the value at room temperature as well [7]. These results indicate that the γ/γ′ lattice mismatch of 5W alloy at high temperature may be lower than the value at room temperature.

3.2. Creep curves

Based on our previous work [8], the applied stress at 900 °C is 50–60% of the yield stress at same temperature which simulates the medium-temperature and medium-stress condition. The creep curves at 900 °C are illustrated in Fig. 3. It can be seen that the creep curves exhibit limited primary creep strains. Increasing the applied stress, the primary creep strain as well as creep strain rate is increased and the creep life is decreased. As shown in Fig. 3c, the plot of strain rate vs. strain, the strain rate is reduced initially and reaches a minimum rapidly. Thereafter the creep strain rate increases gradually until creep rupture eventually occurs. For the relationship between strain rate and creep time (Fig. 3d), it exhibits three stages. The creep strain rate initially decreases rapidly (the primary stage)

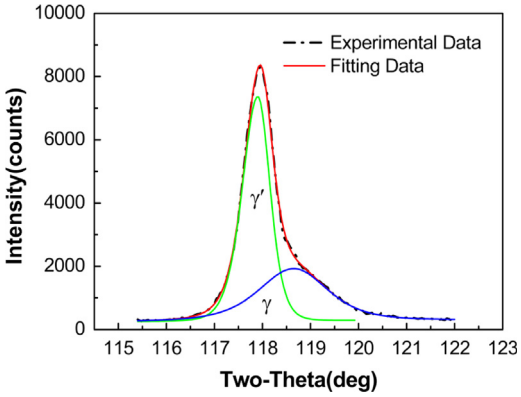

Fig. 2. The (004) diffraction peak of the heat-treated specimen recorded at room temperature.

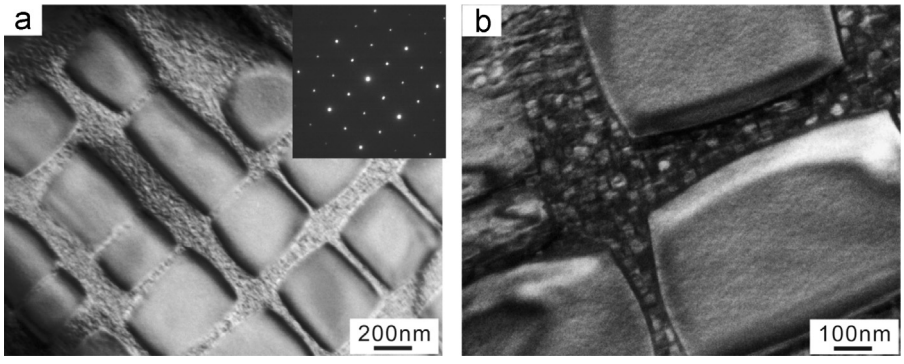

Fig. 1. (a) Bright-field TEM image of 5W alloy after heat treatments and (b) TEM dark-field image showing hyperfine γ′ precipitates. Images are taken with beam direction [001].