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Effects of carbides and its evolution on creep properties of a directionally solidified nickel-based superalloy

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ABSTRACT

By means of heat treatment at different regimes, creep properties measurement and microstructure observation, an investigation has been made into the influence of solution temperatures on the configuration evolution of carbides and creep property of a directionally solidified nickel-based superalloy. Results show that the bulk-like MC type carbide in as-cast superalloy may be decomposed for transforming into the particle-like $M_{23}C_6$ carbides during heat treatment. As the solution temperature enhances and time prolongs, the probability of the decomposition and configuration evolution of the carbide increases. Moreover some fine $M_{23}C_6$ carbide particles are precipitated along the boundaries. Compared to the alloy solution treated at lower temperature, the alloy treated at higher temperature displays a better creep resistance at 780 °C. The fine cubiodal γ' precipitates are homogeneously distributed in the dendrite/inter-dendrite regions, the fine particle-like carbides are precipitated along boundaries to restrain the boundary sliding, which is thought to be the main reasons of alloy having better creep resistance. At the later stage of creep at 780 °C, the micro-cracks in alloy are firstly initiated and propagated along the boundaries at about 45° angles relative to the stress axis, due to bearing the bigger shearing stress during creep, up to fracture, which is thought to be the fracture mechanism of alloy during creep.

1. Introduction

During directional solidification, the crystals of the nickel-based superalloy are grown along [001] orientation, and the mis-orientation of the columnar crystal along the [100] and [0 1 0] orientation exists in the alloy. The mechanical and creep properties of the superalloys at high-temperature are enhanced to a great extent [1,2] because the transverse grain boundaries vertical to the stress axis in the alloy had been eliminated during solidification. Consequently, the superalloys had been widely used for manufacturing the hot end components in the advanced aero-engine.

The microstructure of the directional solidification (DS) nickel-based superalloys consists of γ' phase embedded coherent in the γ matrix, some carbides with various configurations are distributed in the grains and boundaries regions. Adding the refractory elements with higher melting point, such as W, Mo and Ta, may strengthen the γ' phase and γ matrix, the precipitated M_6C, $M_{23}C_6$ and MC carbides during heat treatment and service may improve the creep resistance of alloys [3,4]. But the morphology, size and distribution of the precipitated carbides have a great effect on the high temperature properties, creep life and work reliability of the alloy in service. As the size and edge-angle of the carbides increase, the stress concentration during creep of alloy is easily produced to promote the initiation and propagation of cracks along the carbides/matrix interfaces. The cracks are easily initiated along the regions near the MC type carbide with thick stripe-like and Chinese-like features [5]; the interface between the carbides and matrix is thought to be the propagating channels of the cracks [6]. Generally, the fine particle-like carbides are homogeneously distributed in the matrix of alloys, especially the fine carbides precipitating along boundaries may restrain the boundaries sliding to improve the creep resistance of alloys [7]. Therefore, the configuration evolution and control of carbides have been widely investigated due to the configuration and distribution of carbides having a great effect on the creep properties of alloy [8,9].

Significant amount of bulk-like MC type carbides are mainly precipitated during solidification of alloys [10]. While the particle-like $M_{23}C_6$ type carbides in alloys are mainly precipitated from the matrix and along the boundaries [11,12] during heat treatment and service due to the super-saturation of carbon in the matrix and degeneration of the MC type carbide [13,14]. Because the pinning effect of the sphere-like $M_{23}C_6$ carbides precipitated along

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boundaries may restrain the sliding of boundaries, the creep resistance of alloys may be improved [15,16]. Moreover, the granular $M_{23}C_6$ carbides distributed along boundaries may delay the propagation rate of cracks, so the surface of the fracture displays the feature of the toughness rupture [17].

Although the transverse boundaries in DS superalloys have been eliminated, the longitudinal boundaries parallel to the stress axis are still kept, and the creep damage is still the main failure mode of the alloys in service [18,19]. Moreover, the creep properties of alloys are related to the morphology, quantity and distribution of the carbides, and the ones of carbides may be improved by means of heat treatment regimes [20]. Therefore, the influence of heat treatment regimes on the carbides configuration evolution and creep properties of alloy had been widely investigated [21,22]. But the evolution of carbides during heat treatment and service as well as the effects of carbides configuration on the initiation and propagation of cracks is still unclear.

Hereby, in this paper, by means of the heat treatment at different regimes, creep properties measurement and microstructure observation, the influence of the heat treatment on the configuration, size and distribution of the carbides is explored to investigate the affecting regularity of the carbides configurations on the creep behaviors and fracture mechanisms of a directionally solidified nickel-based superalloy, which may provide the theoretical basis for the development and application of the alloy.

2. Experimental procedure

The bars of the directional solidification (DS) nickel-based superalloy with columnar crystal structure along [001] orientation had been produced in a vacuum directional solidification furnace. The actual chemical compositions of the DS superalloy are listed in Table 1. In order to investigate the effect of heat treatment regimes on the microstructure and creep properties of the alloy, the various heat treatment regimes are adopted. Compared to regime 1, the higher solution temperature of 1260 °C is chosen in regime 2. The heat treatment regimes of the DS nickel-based superalloy are given as follows:

1) 1180 °C × 2 h + 1230 °C × 4 h, A. C + 1100 °C × 4 h, A. C + 870 °C × 20 h, A. C,
2) 1180 °C × 2 h + 1260 °C × 4 h, A. C + 1100 °C × 4 h, A. C + 870 °C × 20 h, A. C.

After fully heat treated at various regimes, the bars of the DS nickel-based superalloy were cut into the plate-like tensile creep specimen with cross-section of 4.5 mm × 2.5 mm and the gauge length of 20.0 mm. After grinded and polished mechanically, the specimens of the alloy were put into a creep testing machine with GWT504 model, and the creep properties of the alloy were measured under the different conditions, some creep tests were interrupted at different stages, respectively. The microstructures of the alloy treated by different regimes and crept for different times are observed under Scanning electron microscope (SEM) and Transmission electron microscope (TEM), for investigating the effects of heat treatment regimes on the microstructure and creep properties of the alloy. The deformation and fracture mechanism of the alloy treated by different regimes during creep are briefly discussed.

Table 1
Chemical compositions of the superalloy (mass fraction, %).

Cr	Co	W	Mo	Al	Ti	Ta	Hf	B	C	Ni
8.68	9.80	7.08	2.12	5.24	0.94	3.68	1.52	0.012	0.09	Bal.

3. Experimental results and analysis

3.1. Effect of heat treatment on carbides morphology

The dendrite morphology of the as-cast directional solidification nickel-based superalloy on (001) cross-section is shown in Fig. 1(a); it is indicated that the dendrites on the cross-section are regularly arranged in the form of the regular "+" model, the average spacing of the primary dendrite is measured to be 150–160 μm, the dendrites in same grains are arranged along the same orientations. The mis-orientation of dendrites exists in between grains A and B; for example, the mis-orientation in between grain A and grain B is about 20°, as marked by the thicker straight lines in Fig. 1, the grain boundary is located in the region between grains A and B.

The morphology of the eutectic structure and carbides in the as-cast alloy are shown in Fig. 1(b), in which the eutectic structure with radial-like configuration is located in the inter-dendrite regions, the fine structure in the eutectic center region with radial configuration consists of fine γ' and γ phases, as shown in the region C of Fig. 1(b), and the coarser γ'/γ phases are located in the around region. The bulk-like carbides are located in both dendrite arm regions and inter-dendrite regions, as marked by the arrow in Fig. 1(b), and the fine γ' phase is located in the dendrite arm region, as shown in the region D in Fig. 1(b). It is indicated by SEM/EDS composition analysis that the elements about 17.2% C, 13.4% B, 9.93% Hf, 12.49% Ta and 10.34%W (atomic fraction, %) are enriched in the block-like carbides. Therefore, it may be deduced that the bulk-like carbide is thought to be the M(W, Ta, Hf)C(C,B) type carbide due to the ratio of the C, B and Hf, Ta, W (atomic fraction) being close to 1: 1.

After solution treated at 1230 °C and two-stage aging treatment, the microstructure of the alloy is shown in Fig. 2. After solution treated at 1230 °C, the fine γ' phase in size distributing in the dendrite arm regions may be dissolved to increase the saturation level of the solute elements in matrix. Then the homogeneous fine γ' phase may be precipitated in the dendrite arm regions during cooling and aging treatment, as shown in the region D of Fig. 2(a). But the coarser γ' phase distributing in inter-dendrite regions cannot be completely dissolved due to the lower dissolving rate of elements in low temperature, so that the γ' particles in the inter-dendrite regions may grow continually, during the latter heat treatment, to display the bigger size, as shown in the region E in Fig. 2 (a) and (b). The fine γ' phase in the dendrite arm regions is precipitated during cooling in air, and the one may be homogeneously grown up during aging treatment, as shown in the region D in Fig. 2(a) and (b). Moreover, it is indicated that the bulk-like carbides have transformed into the mesh-like morphology in some regions, as marked by the white arrow in Fig. 2(a), in which the grain boundaries are marked by the black arrow. While the coarser MC type carbide exists still in another region of the inter-dendrite, as marked by the white arrow in Fig. 2(b), some coarser γ' particles are distributed around the carbide.

The bulk-like carbides' configuration in as-cast alloy is marked by the arrow in Fig. 1(b) and some eutectic structure is distributed in the inter-dendrite regions. The morphology of some carbide particles in the alloy after heat treated at different stages of regime 2 is shown in Fig. 3. The microstructure of the alloy solution treated for 2 h at 1260 °C is shown in Fig. 3(a), which indicates that the edge of the original bulk-like carbides has been dissolved for transforming into the sphere-like morphology to distribute around the bulk-like carbides, as marked by the white arrow in Fig. 3(a).

After solution treated at 1260 °C for 4 h and aging treated, the microstructure in the located region of the alloy is shown in Fig. 3(b), which indicates that the majority of the original bulk-like carbides have transformed into the sphere-like morphology. In another located region, the morphology of the bulk-like carbides in alloy is shown in Fig. 3(c), which indicates that the bulk-like carbides in alloy have transformed into the sphere-like morphology for distributing within the grains and along the boundaries. It is indicated by SEM/EDS composition analysis that the elements about 5.91% C, 7.4% B,