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Influence of primary and secondary orientations on creep rupture behavior of aluminized single crystal Ni-based superalloy

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ABSTRACT

Influence of primary and secondary orientations on the creep behavior of aluminized single crystal Ni-based superalloy was investigated at a high temperature with various tensile stress directions and side-surface orientations. The specimens were aluminized by pack aluminizing treatment at 1000 °C for 5 h under argon flow. The creep rupture tests were performed at 900 °C/392 MPa for thick specimens and at 900 °C/300 MPa for the thin specimens. The aluminizing treatment reduced the creep strength of the superalloy because of the crack initiation in hardened diffusion layers. Furthermore, it was found that the creep strength of thin aluminized specimens was affected by the crystallographic orientations, not only in the tensile direction, but also in the thickness direction. This is due to the difference in the crystallographic geometry of {111} planes, which is associated with the magnitude of ⟨112⟩ direction shear stress on the planes and the effective cross-section change by the crack propagation.

1. Introduction

Single crystal Ni-based superalloy is the material with superior high-temperature strength and environmental resistance. These superalloys are utilized as turbine blades of the industrial gas turbine and jet engine at high temperatures and high stress environments. Under high temperature circumstances, the fatigue, creep and hot corrosion induced the service lifetime of the turbine blade is limited [1]. To prolong its service lifetime, the operating temperature should be demoted by air cooling and/or by applying the thermal barrier coating. Alongside these advances in materials, significant improvements have been made to the cooling configurations. The latest generation blades use a multi-pass configuration, where the flow passes through a long passage before being exhausted. This maximizes the heat pick-up of the cooling air, increasing the main stream gas temperatures and reducing the required cooling flow. These advantages combine to improve the engine efficiency. The air has a relatively long path inside the blade before being exhausted through dust holes and film cooling holes, thereby increasing its potential to absorb heat through contact with the metal. Thinner walls increase the cooling efficiency and reduce the weight of the turbine blades. However, the improvements of the cooling configuration have brought about the decrease of wall thickness of the rotating blade. Nowadays, single-crystal blades designed with a wall thickness of 0.7 mm are being used [3]. In consequence, the effect of section thickness on creep deformation and rupture plays an important role in design and durability [2]. In addition, diffusion coating (i.e. aluminide coating) has been applied on the surface of the blades to protect them from oxidation and corrosion at high temperatures. When the aluminized specimens are exposed to high temperature, the microstructure is changed by the diffusion between the coating layer and the substrate. This microstructure change can cause the formation of topologically close-packed (TCP) phase [4], and secondary reaction zone diffusion zone (SRZ) in which TCP precipitation and recrystallization take place as a result of the interdiffusion between coating layer and substrate [5]. TCP phases formed during the coating process and in service result in decreasing creep strength [4]. Furthermore, the single crystal superalloys used in the turbine blade have remarkable anisotropic properties. The plastic anisotropy is determined by the arrangement of slip systems which is assigned by the crystallographic orientation of the specimen and the load stress direction. Although studies have been carried out on the creep properties of aluminized Ni-based single crystal superalloys [6,7], no study ever tried to discuss the effects of the primary and secondary crystallographic orientations using thin specimens. Therefore, the purpose of this study is to investigate the mechanism of anisotropy which is characteristic for the single crystal and its effect on creep strength using thin aluminized single crystals.

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2. Experimental procedures

A second generation single crystal Ni-based superalloy CMSX-4 was used as an experimental material in this study. The nominal composition of single crystal superalloy CMSX-4 in mass % is 9 Co, 6.5 Cr, 0.6 Mo, 1 Ti, 6W, 5.6 Al, 6.5 Ta, 3 Re, 0.1 Hf and balance Ni. The alloy was subjected to solution heat treatments (2 h at 1277 °C, 2 h at 1288 °C, 3 h at 1296 °C, 3 h at 1304 °C, 2 h at 1313 °C, 2 h at 1316 °C, 2 h at 1318 °C and 2 h at 1321 °C) and a two-step aging treatment (4 h at 1140 °C and 20 h at 870 °C). The initial crystallographic orientation of the superalloy CMSX-4 was determined by the X-ray Laue reflection method. The creep specimens were prepared by electric discharge machine (EDM) (Fig. 1). The creep specimens were cut with different cross-section areas of 2.8 mm × 2.8 mm (square cross-section) as thick specimen and 2.8 mm × 0.5 mm and as a thin specimen with a gauge length of 19.6 mm. The creep specimens were then mechanically polished down to 1200 mesh by SiC paper and ultrasonically cleaned in acetone bath for 10 min prior to the aluminizing process. In the present study, the High-Temperature Low-Activity (HTLA) process was used to prepare an aluminide coating with a mixture of 24.5 mass% Al, 24.5 mass% Cr, 49 mass% Al_2O_3 and 2 mass% NH_4Cl powders. The aluminizing treatment was conducted at 1000 °C for 5 h in flowing argon and then cooled to room temperature. The creep tests were then performed at two different conditions, that is, at 900 °C/320 MPa with orientations B and C for the thick specimens and at 900 °C/300 MPa with orientations A–D for the thin specimens. Each stress was applied for the cross-section area of the substrate before coating. The cross-section microstructures of the ruptured coated specimens were observed by a scanning electron microscopy (SEM). Microhardness measurement was applied from the coating to substrate regions at a load of 200 g for 20 s. The Vickers hardness number is based on the average diagonal length of an imprint made from the indenter. This measurement is useful to evaluate the gradient of the hardness in the coating-to-substrate regions.

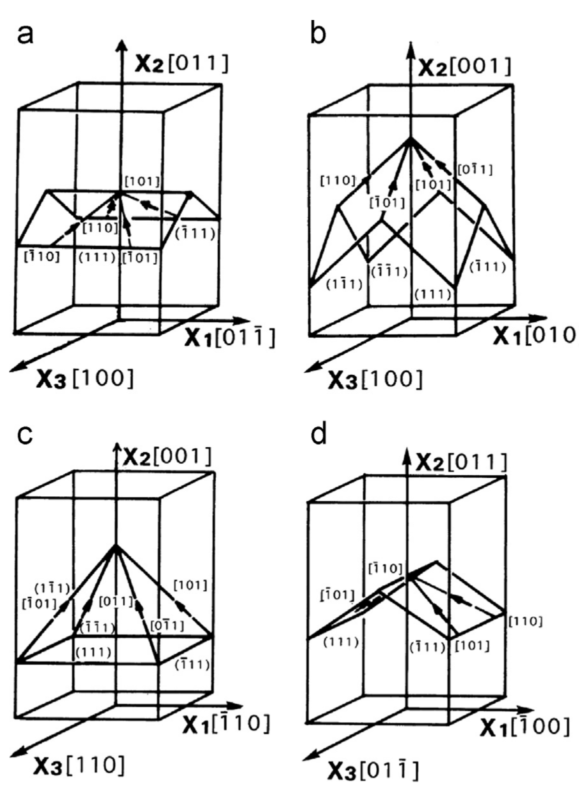

Fig. 1. Arrangement of {111}⟨101⟩ slip systems for four types of creep specimens. (a) Orientation A, (b) orientation B, (c) orientation C and (d) orientation D.

3. Results

3.1. Creep rupture test

3.1.1. Creep rupture life of thick aluminized specimen

The creep rupture lives of thick (square cross section) aluminized specimens at 900 °C/392 MPa are shown in Fig. 2. The creep rupture lives of the bare specimens were generally higher than those of the aluminized specimens in both orientations B and C. This indicated that the aluminizing treatment resulted in a significant decrease in creep rupture lives. In orientation B, the creep rupture lives were 360 h for the bare specimen and 250 h for the aluminized specimen. However, in orientation C, the creep rupture lives were 345 h for the bare specimen and 220 h for the aluminized specimen. With regard to the side-surface orientation, the specimens with orientation B had a slightly longer creep rupture life than those of the specimens with orientation C. But the difference in creep rupture lives between the two orientations was small in both bare and aluminized specimens.

3.1.2. Creep rupture life of thin aluminized specimen

The creep rupture lives at 900 °C/300 MPa of thin aluminized specimens are shown in Fig. 3. It was clear that the [001]-tensile stress direction (orientations B and C) showed a longer creep rupture life and higher ductility than the [011]-tensile stress direction (orientations A and D). In case of specimens with [001]-tensile stress direction, orientation B showed a lower creep strain rate and a longer creep rupture life than that of orientation C. The creep rupture lives were 491 h for orientation B and 186 h for orientation C. The creep rupture life for orientation B was about 2.6 times longer than that of orientation C. Concisely, the influence of crystallographic orientation on creep behavior was more pronounced in the thin specimen (Fig. 3) than that in the thick specimen (Fig. 2). Meanwhile the specimens with [011]-tensile stress direction, orientation A showed a larger creep strain rate and shorter rupture life compared with orientation D. The creep rupture life was 89 h for orientation D and 4 h for orientation A. The creep rupture life in orientation D was increased by a factor of 22 compared to orientation A.

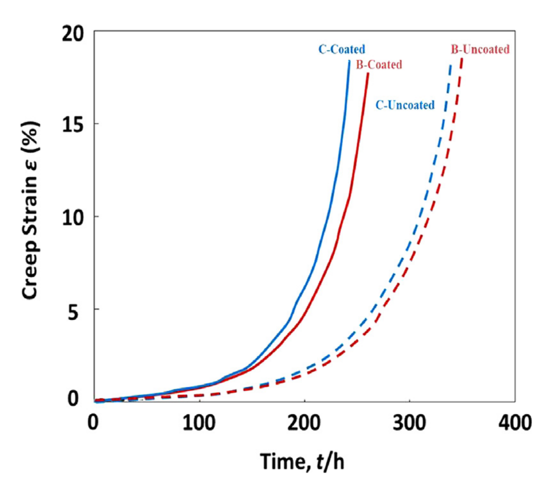

Fig. 2. Creep curve of thick aluminized single crystal Ni-base superalloy CMSX-4 at 900 °C/392 MPa.