Materials Science and Engineering A 532 (2012) 579–584

Contents lists available at SciVerse ScienceDirect

Materials Science and Engineering A

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

Influence of the characteristic of recrystallization grain boundary on the formation of creep cracks in a directionally solidified Ni-base superalloy

G. Xie*, L.H. Lou

Superalloys Division, Institute of Metal Research, Chinese Academy of Sciences, 72 Wenhua Road, Shenyang 110016, China

ARTICLE INFO

Article history:
Received 31 May 2011
Received in revised form 30 August 2011
Accepted 26 October 2011
Available online 11 November 2011

Keywords:
Recrystallization
Superalloy
Cracks
Dislocation
Grain boundary

ABSTRACT

Creep tests of a directionally solidified (DS) Ni-base superalloy specimens containing local recrystallization (RX) were carried out at 980 °C/235 MPa and the influence of the characteristic of RX grain boundary on the formation of creep cracks was investigated. The RX grain boundary morphology, misorientation and dislocation pattern at high, low angle and twin boundaries during creep were examined. It was shown that only a few high angle grain boundaries (GBs), but not all of them cracked during creep and even after ruptured. This is associated with the dislocation mobility at GBs or the morphology of GBs. The possibility of crack formation was discussed based on the transmission behavior of dislocations at different GBs.

1. Introduction

Recrystallization (RX) in DS blades caused either by the residual stresses or by the plastic deformation generated during manufacturing is a well-known issue in the investment casting industry. The effects of applied stress [1], heat treatment parameters [2,3], heat treatment atmosphere [4], as well as the microstructural features such as residual eutectics [5] and carbides [6] on RX behaviors in several DS superalloys have been reported.

It is generally accepted that RX may reduce the mechanical properties of DS superalloys [7–10]. For example, less than half of the creep rupture life of DZ22 alloy can be achieved when there was a surface RX layer induced by shot-peening [7]. Similar data were also reported by Khan et al. [8]. Besides the RX layer induced by shot-peening, local RX showed similar detrimental effect on the creep rupture life of DS superalloys [9,10].

As is well known, the detrimental effect of RX was due to the formation of the transverse GBs in DS superalloys. In our previous work, the microstructural evolution during high temperature creep test of a DS alloy containing location RX was investigated [11]. However, the detailed studies on relationship between the characteristic of RX grains and failure of DS superalloys containing local RX during creep have not been conducted. On the other hand, the deformation and failure of crystals have been related to the characteristic of GBs in many simple systems. For example, Huang et al. [12,13] observed that the development of the deformation microstructure showed strong orientation dependence in tensile strained copper and aluminium. Cavitations were found on most random high angle boundaries, whereas coincident site lattice boundaries tended to resist cavitation [14,15].

The aim of our present paper is to study the effect of the characteristic of RX GBs on the formation mechanism of cracks during high temperature creep of a DS superalloy, and try to reveal the microstructural evolution during creep.

2. Experimental

The nominal compositions of the DS superalloy studied (DZ125L) are shown in Table 1. Alloy was directionally solidified into plate with the size of $220 \times 70 \times 16$ mm using a Bridgman furnace. Details of the DS process were reported elsewhere [16]. The resulted DS slab has an average grain size of 2–3 mm and a primary dendrite arm spacing around 300 μm.

DS slabs were cut into plates by electron discharge machining (EDM) and indented using Brinell hardness tester on one side. After heat treatment (1220 °C/2 h/AC + 1080 °C/4 h/AC + 900 °C/16 h/AC), the local RX occurred. The morphology of RX under the indentation after heat treatment was shown in Fig. 1a. These RX GBs were relatively smooth and only a few large size second phase formed. The indentation was then carefully removed by grinding. Dimension of the plate specimens for creep tests and location of RX were shown in Fig. 1b schematically.

Creep tests were performed at 980 °C/235 MPa with the stress axis parallel to the DS direction. Some of the creep tests were

* Corresponding author. Tel.: +86 24 23971712; fax: +86 24 23971712.

0921-5093/$ – see front matter © 2011 Elsevier B.V. All rights reserved.
doi:10.1016/j.msea.2011.10.124

Table 1
Nominal chemical composition in wt.% of alloy investigated.

Cr	Co	W	Mo	Al	Ti	Ta	C	B	Ni
9	10	7	2	5	3.5	4	0.1	0.01	Bal.

interrupted at different stages to investigate the microstructural evolution.

Specimens for optical microscopy observation were prepared by the standard metallographic procedure. Electron backscatter diffraction (EBSD) technique was employed to determine the orientation of RX grains and misorientation of GBs during creep. The dislocation pattern during creep was characterized by transmission electron microscopy (TEM, Philips TECNAI 20). After interrupted creep tests, small plate was cut from the RX site as well as the un-recrystallized region of the specimen. The plane of the small plate was parallel to the specimen surface. These small plates were first sliced into 50 μm discs and then ion polished into TEM foils.

3. Results

Creep curve at 980 °C/235 MPa of specimen containing local RX was shown in Fig. 2a. The creep rate of the specimen as a function of time was shown in Fig. 2b. The maximum depth of local RX in specimen is about 1000 μm. Three regimes were observed, namely, a very short primary creep, followed by secondary (steady) creep and tertiary (accelerated) creep. Circles in Fig. 2b indicate the interrupted points during creep test at 980 °C/235 MPa.

Microstructure of the specimen interrupted at the primary creep stage indicated that micro-cracks formed only at a few RX GBs, as showed in Fig. 3a. Cracks did not nucleate at all of transverse RX GBs, and no crack was found at the twin boundaries in Fig. 3a. Distribution of grain orientation indicated that the original DS materials had the [0 0 1] crystallographic direction, but the RX grains with various orientations obviously departed from above crystallographic direction (Fig. 3b). A lot of sub-grain boundaries which had the misorientation angle less than 5° were observed between RX grains (showed as white arrows). These sub-grain boundaries may be formed during pre-deformed treatment or creep test. RX grains with cracks were shown as black arrows. In most case, the orientation of RX grains with cracks was near the [1 1 1] (blue grains) or the [1 0 1] (green grains), that is, away from the [0 0 1].

According to the EBSD result in Fig. 3, the relationship between crack formation and the misorientation angle θ of RX GBs was shown in Fig. 4. It indicated that most of RX GBs had large misorientation angle and only a few low angle RX GBs formed. Moreover, these low angle RX GBs did not crack. When misorientation angle of RX GBs was beyond about 20°, cracks initiated, but not all of high angle RX GBs cracked. Statistically meaningful data indicated that about 30 percent high angle GBs ($\theta > 20°$) of RX grains partially cracked.

Fig. 5 shows the dislocation pattern near RX GBs with different misorientation angle after interrupted at the primary stage of creep test. Due to the coherence between γ' and γ matrix, the misorientation of two neighbor grains can be estimated from the orientation of γ' particles in the grains. The angle between RX grain and those of original DS grain is about 43° and 8°, respectively. From Fig. 5a, one can see that dislocations mainly generated and moved in the γ matrix. Near the RX grain boundary, a large amount of dislocations generated in the original DS material. On the contrary, low dislocation density in the γ matrix was observed and only a few dislocations generated near the grain boundary with small misorientation angle, as shown in Fig. 5b. Low dislocation density in the original DS material may be due to smaller strain at the primary

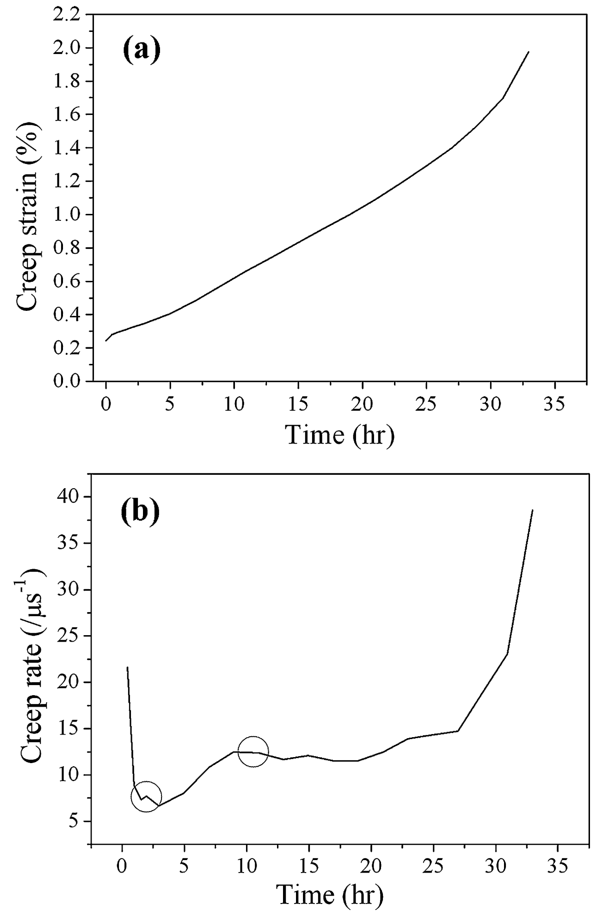

Fig. 2. Creep strain (a) and creep rate (b) curves of specimen containing local recrystallization at 980 °C/235 MPa.

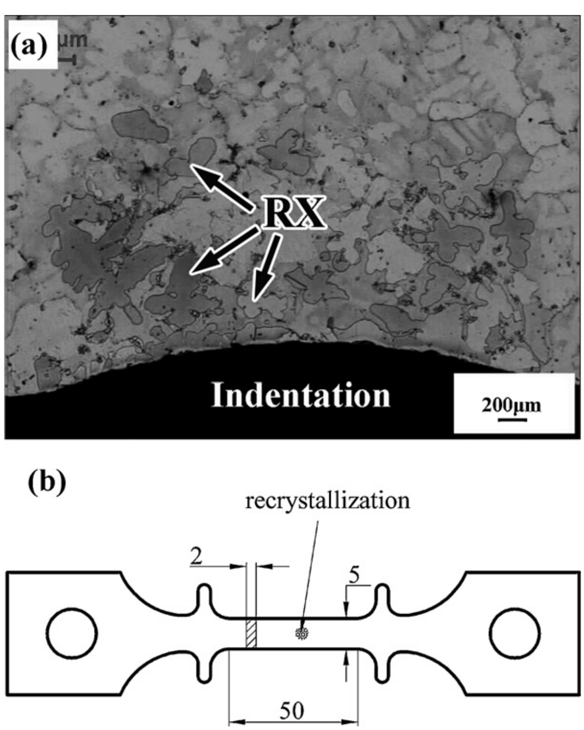

Fig. 1. (a) Morphology of RX after heat treatment. The local RX formed mainly in the dendritic core region, leaving the interdendritic region un-recrystallized. (b) Dimension of specimen and location of RX for creep test.