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Microstructural evolution of a nickel-based superalloy during hot deformation

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ABSTRACT

Hot compressive tests of a nickel-based superalloy are performed under the strain rate range of 0.001–1 s^{-1} and deformation temperature range of 920–1040 °C. Optical microscopy (OM) and transmission electron microscopy (TEM) are employed to investigate the evolution of dynamic recrystallized (DRX) grain and dislocation substructure. It is found that the effects of deformation degree, strain rate and deformation temperature on DRX grain are significant. When the deformation degree or temperature is increased, the number of DRX grains rapidly increases. But, the increase of strain rate reduces the number of DRX grains. The dislocation substructure is also very sensitive to the deformation degree, strain rate and deformation temperature. With the increase of deformation degree, the evolution of dislocation substructure can be characterized as: high dislocation density → dislocation network → subgrain → DRX grain. Under high deformation temperatures or low strain rates, the dislocation substructure can be easily annihilated and rearranged because of the occurrence of DRX. Based on the evaluated DRX volume fractions, the contour map is constructed to optimize the hot deformation parameters.

1. Introduction

Generally, metals or alloys undergo the large plastic deformation in hot forming processes [1,2]. The complex microstructural evolutions [3–5] are often induced by the multiplicate hot deformation mechanisms, such as work hardening (WH), dynamic recovery (DRV) and dynamic recrystallization (DRX) [6–8]. It is well known that the dynamic recrystallization is an effective means to refine grains [9–11]. In recent years, some researches have been reported on the microstructural evolution and dynamic recrystallization behaviors of various alloys. For example, the dynamic recrystallization behaviors of typical nickel-based superalloy [7], 30Cr2Ni4MoV steel [12,13], 7075 aluminum alloy [14], forged 42CrMo steel [15,16], as-extruded 42CrMo steel [17], Fe–30Mn–3Si–3Al TWIP steel [18], low-carbon bainite steel [19], AISI 4140 alloy steel [20], Fe–30Mn–5Al steel [21], AISI 304L and 316L stainless steels [22,23], commercial purity titanium [24], high manganese austenitic stainless steel [25], 4130 and 4340 steels [26,27], and a medium carbon microalloyed steel [28] were studied, and the suitable DRX kinetics models were established to predict the volume fractions of DRX. It is found that the microstructural evolutions during dynamic recrystallization are sensitive to the processing parameters, such as strain rate, deformation temperature and strain. Therefore, understanding the microstructural evolution and dynamic recrystallization behaviors of metals or alloys is very important.

Nickel-based superalloys, which have excellent high-temperature mechanical properties, are widely used in modern aero engines and gas turbines. Over the last decades, some investigations on the hot deformation behaviors of nickel-based superalloys have been reported [29–40]. Wen et al. [29] identified the optimum processing parameters for an aged GH4169 superalloy by processing maps. Shore et al. [30] found that the wrought Incoloy 901 alloy has the better hot workability, compared to the cast one. Lin et al. [31] investigated the effects δ phase on the fracture characteristics of an aged nickel-based superalloy, and found that δ phase is the potential nucleus sites for microvoids. Ning et al. [32] established the processing maps for GH4169 superalloy, and the safe

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hot working domains were determined. Lin et al. [33] found that the fracture of nickel-based superalloy is attributed to the coupling effect of microvoid and localized necking. Etaati et al. [34,35] investigated the hot deformation behaviors of Ni–42.5Ti–7.5Cu and Ni–42.5Ti–3Cu alloys, and established the suitable constitutive equations to predict the flow stress. Also, Ning et al. [36] established phenomenological constitutive equations to describe the hot deformation behaviors of the post-cogging FGH4096 superalloy. Wu et al. [37] found that DRX has significant effects on the activation energy, as well as the power dissipation maps of a Ni–Cr–Co based P/M superalloy. Zhang et al. [38] found that the hot deformation behaviors of a powder metallurgy nickel-based superalloy FGH96 are significantly affected by the initial states. Wang et al. [39] discussed the effects of dynamic recrystallization mechanism on the hot deformation characteristics of a nickel-based corrosion resistant alloy by processing maps. Antonov et al. [40] studied the precipitate phase stability and compositional dependence on alloying additions in γ–γ'–δ–η Ni-base superalloys.

Although a number of studies have been carried out to investigate the hot workability and microstructural evolution of different superalloys, the research on the evolutions of grain microstructure and dislocation substructure are still limited for nickel-based superalloys. The objective of this study is to systematically discuss the effects of hot deformation parameters on the evolutions of grain microstructure and dislocation substructure, as well as the DRX degree. Meanwhile, the contour map is established based on the area fractions of DRX.

2. Material and experiments

A commercial nickel-based superalloy (GH4169) was selected in this study, and its chemical composition (wt.%) is: 52.82Ni–18.96Cr–5.23Nb–3.01Mo–1.00Ti–0.59Al–0.01Co–0.03C–(bal.)Fe. Cylindrical specimens with a diameter of 8 mm and a height of 12 mm were cut down from the wrought billet. The specimens were solution treated at 1040 °C for 45 min followed by water quenching. Hot compressive tests were performed on a Gleeble-3500 thermo mechanical simulator under the strain rate range of 0.001–1 s^{-1} and deformation temperature range of 920–1040 °C. In order to reduce the friction and avoid the adhesion, tantalum foil with the thickness of 0.1 mm was placed between the specimen and dies. Before loading, each specimen was heated to the designed deformation temperature at a rate of 10 °C/s, and then soaked for 300 s to reduce thermal gradient. Then, the specimens were compressed, and the height reductions are between 11% and 70%. i.e., true strains are between 0.12 and 1.2. After hot compressive deformation, the specimens were immediately quenched by water at a cooling rate of 500 °C/s.

In addition, grain microstructure was observed by a Leica DMI5000M optical microscope (OM). Samples for OM analysis were sliced along the compression axis section, mechanically polished, and then chemical etched for 3–5 min with a solution consist of 100 ml HCl + 100 ml CH_3CH_2OH + 5g $CuCl_2$ at room temperature. Fig. 1 shows the microstructure of the studied superalloy before hot deformation. From Fig. 1, it can be observed that the grains were nearly equiaxed, and the mean grain size was evaluated as 75 μm by the linear intercept method (ASTM: E112-12). In order to analyze the dislocation substructures in the deformed specimens, the transmission electron microscope (TEM) observations were performed on the JEM-2100F microscope. Samples for TEM observations were firstly machined from the deformed specimens, and then grinded into 70–80 μm thick foils. Subsequently, some disks with 3 mm in diameter were punched out from the grinded thin foils, and electro-polished using a solution of perchloric acid and ethanol (1:9 in volume).

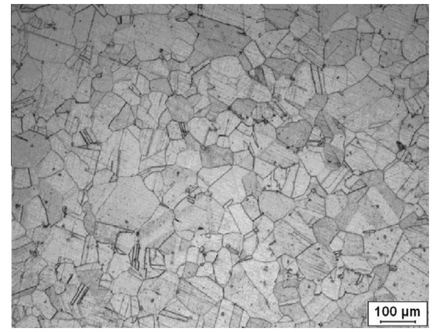

Fig. 1. Optical micrograph of the solution-treated superalloy before hot deformation.

3. Results and discussions

3.1. Hot compressive deformation behaviors

Generally, the interfacial friction between the dies and specimen is unavoidable during hot deformation. In order to exclude the effects of friction on flow stress, all the stress-true curves are corrected based on the methods introduced in Ref. [41]. Fig. 2 illustrates the typical true stress–strain curves of the studied superalloy. From Fig. 2, it is observed that the flow stress first rapidly increases to a peak, and then gradually decreases until a relatively steady state reaches under given deformation conditions. The characteristics of flow stress are related to the competing phenomenon between work hardening and dynamic softening (induced by DRV and DRX) mechanisms during hot deformation. At the initial stage of hot deformation, the main deformation mechanisms are the work hardening and dynamic recovery. The dislocations rapidly multiply and accumulate, resulting in the obvious work hardening. Meanwhile, because the stacking faults energy of the studied superalloy is relatively low, the dynamic recovery is too weak to overcome the work hardening at the initial stage of hot deformation. Thus, the flow stress remarkably increases. With the further straining, the flow stress gradually decreases. This is because the dynamic recrystallization results in the distinct dynamic softening. It is well known that the dislocation is swept away by grain boundaries in the course of new grain formation. So, the work hardening rate is decreased [7]. Consequently, the flow stress is gradually reduced with the further straining. When the work hardening and dynamic softening reach a new balance, the steady flow stress appears.

In addition, Fig. 2 indicates that the flow stress is significantly affected by strain rate and deformation temperature under the tested conditions. When the strain rate is decreased or the deformation temperature is increased, the flow stress obviously decreases. The drop in flow stress is related to DRX behavior. It is well known that the nucleation and growth of DRX are thermal activation process. One the one hand, although low strain rate decreases the rate of DRX nuclei, it provides sufficient time for DRX grains to grow up. On the other hand, high deformation temperature increases the grain boundary mobility, and thus also promotes the growth of DRX grains. So, the dynamic softening is increased, leading to the decreased flow stress [42,43]. Meanwhile, it is observed that the flow stress shows a plateau shape characteristic around the peak stress when the deformation temperature is low or the strain rate is high, as shown in Fig. 2. The