Materials and Design 99 (2016) 107–114

Contents lists available at ScienceDirect

Materials and Design

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

Study of static recrystallization behavior in hot deformed Ni-based superalloy using cellular automaton model

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ARTICLE INFO

Article history:
Received 30 November 2015
Received in revised form 3 March 2016
Accepted 10 March 2016
Available online 12 March 2016

Keywords:
Ni-based superalloy
Static recrystallization
Cellular automaton
Grain boundary

ABSTRACT

A cellular automaton model (CA) with probabilistic state switches is developed to simulate the microstructural evolution in a Ni-based superalloy during static recrystallization (SRX). The kinetics of SRX is formulated on a mesoscale level. SRX nuclei grow under the driving pressure resulting from the reduction of stored energy and grain boundary energy. It is found that SRX grain size can be well predicted by the developed CA model. Solute elements of the studied superalloy have a drag effects on grain boundary motion. But, this solute drag effect is not obvious at high deformation temperatures. Nucleation rate of the SRX grain increases with the increase of strain, strain rate, and deformation temperature. Besides, the inhomogeneity of grain size first increases, then decreases with the increase of SRX fraction. The changes of inhomogeneity mainly result from the evolutions of the primary and SRX grain sizes.

1. Introduction

Generally, metals or alloys undergo multi-pass deformation in hot forming processes [1,2]. During the inter-passes, microstructures are modified by static recovery (SRV), static recrystallization (SRX), and metadynamic recrystallization (MDRX) [3–6]. The microstructural modification during inter-passes has a significant influence on the subsequent hot deformation behaviors. Therefore, understandings of microstructural change during inter-passes are of great importance for the improvements of products' properties [1,7–11].

Cellular automaton (CA) method is a powerful tool to study static, dynamic and metadynamic recrystallizations in hot deformed metals and alloys. In past, a number of researches have been carried out to study the recrystallization behaviors of some typical alloys [13–24]. Reyes et al. [12] successfully predicted the grain size evolution of a Ni-based superalloy during dynamic recrystallization by CA method. Shojaeefard et al. [13] studied the microstructural evolution of AA1100 alloy during dynamic recrystallization by coupling a CA method with a modified Laasraoui-Jonas model. Chen and Cui [14] quantitatively predicted the microstructural evolution of 30Cr2Ni4MoV alloy in a four-hit forging process by CA method. Chen et al. [15], Jin et al. [16], Liu et al. [17], and Liu et al. [18] developed the suitable CA models to describe the dynamic recrystallization behaviors of 30Cr2Ni4MoV rotor steel, AZ31 magnesium alloy, GH4169 superalloy, and Q235 steel, respectively. Han et al. [19,20] proposed a CA model to simulate the grain growth in the hot deformed low carbon steel with and without the second phase particles. Sieradzki and Madej [21] compared the CA and Monte Carlo methods to model static recrystallization behaviors for a C—Mn carbon steel. Salehi and Serajzadeh [22] assessed the microstructural evolution of the cold deformed carbon steels during non-isothermal heat treatment by a CA model. Also, Afshari and Serajzadeh [23] simulated the static recrystallization behavior of the cold side-pressed low carbon steels using CA method.

Owing to their favorable service performance, Ni-based superalloys are broadly used in aerospace and power generation components. In recent years, some investigations have been carried out on hot deformation behaviors of Ni-based superalloys [25–39]. Chen et al. [24] found that the recrystallized grain possesses a low dislocation density due to the sufficient release of stored energy. Yang et al. [25] found that the uniform temperature distribution is beneficial for uniform microstructures of a Ni-based superalloy weldment. In order to optimize processing parameters, Wen et al. [26,27], Zhang et al. [28], Jiang et al. [29], He et al. [30] and Liu et al. [31] developed the processing maps of some typical nickel-based superalloys. Besides, the microstructural evolution of some nickel-based superalloys were investigated by Lin et al. [32,33], Momeni et al. [34], Zhang et al. [35,36], and Ning et al. [37].

In this study, the hot compressed deformation experiments are carried out to study the static recrystallization behavior of a Ni-based

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http://dx.doi.org/10.1016/j.matdes.2016.03.050
0264-1275/

superalloy. A cellular automaton model is developed to predict the microstructural evolution of the studied Ni-based superalloy during static recrystallization. In the developed CA model, the suitable functions are developed to predict the grain boundary mobility and nucleation rate under static recrystallization conditions. Finally, the developed CA model is validated by the experimental results.

2. Material and experiments

A commercial Ni-based superalloy with chemical compositions (wt.%) of 52.82Ni–18.96Cr–5.23Nb–3.01Mo–1.00Ti–0.59Al–0.01Co–0.03C–(bal.) Fe was used. Hot compressive deformation tests were carried out on cylindrical specimens (Φ8 mm × 12 mm) in order to study the static recrystallization behavior of the studied superalloy. Fig. 1 shows the detailed experimental procedure of hot compression tests. All the specimens were homogenized at 1313 K for 45 min, and then rapidly quenched by water. Then, each specimen was heated to deformation temperature by a heating rate of 10 K/s, and the specimen was held for 5 min to create a uniform temperature field prior to hot compression. Hot compressive deformation tests were carried out on a Gleeble-3500 simulator at the deformation temperatures of 1223–1313 K, and the strain rates of 0.01–0.1 s^{-1}. The specimens were compressed to designed strains (0.08–0.17), and the compressed specimens were hold at the deformation temperature for 60–600 s before water quenching. Tantalum foils (0.1 mm) were placed between the specimen and dies for reducing friction.

The initial and deformed microstructures of specimens were examined by DMI5000M optical microscope (OM). The specimens were firstly portioned parallel to the longitudinal compression axis. Then, the exposed sections were subjected to mechanical polish and chemical etching at room temperature. A chemical mixture consisting of HCL (100 ml) + CH_3CH_2OH (100 ml) + $CuCL_2$ (5 g) was used for etching, and the etching time was 3–5 min. The average grain sizes were measured following the standard ASTM: E112–12, and the average initial grain size is evaluated as 75 μm.

3. Development of CA model

In order to simulate the microstructural evolution during static recrystallization (SRX), a mesoscale model is developed by coupling the SRX theory with CA method. In the following subsections, the theory on the nucleation and growth of SRX grains are firstly presented. Analytical equations of the developed CA model are described in detail. Then, the procedures of simulation are proposed.

3.1. Nucleation of SRX grains

SRX is a dynamic process in which the grain nuclei form and grow up [38,39,40]. The grain nuclei closely relate with the growth of subgrains. Pre-deformation may induce a large amount of small subgrains. In the subsequent holding period, subgrains in the vicinity of primary grain boundaries may grow. When the size of subgrain is large enough for a capillary pressure to be overcome, the subgrain turns into a recrystallized nucleus [41]. Since the cell size of the present CA is larger than that of subgrain, the formation of SRX grain nuclei cannot be explicitly simulated [42]. Thus, a constant nucleation rate is employed in the CA model, i.e., grain nuclei are continuously provided at a constant rate. The nucleation rates under different conditions are identified according to the experimental results.

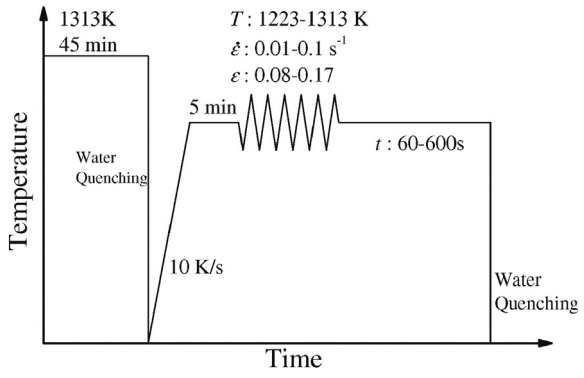

Fig. 1. Experimental procedure of hot compression tests.

3.2. Growth of SRX grains

Generally, grain nuclei grow by means of grain boundary motion. The boundary moves under a driving pressure (P) resulting from a reduction of free energy. The relationship between the moving velocity of grain boundary (V) and the driving pressure (P) can always be described as [43],

$$V = MP \tag{1}$$

where M is the so-called mobility of grain boundary which is dependent on grain boundary misorientation [44].

$$M = M_0\left(1-\exp\left[-5\left(\frac{\theta}{\theta_0}\right)^4\right]\right) \tag{2}$$

where M_0 is the mobility of high-angle boundaries. The identification of M_0 will be discussed in detail in Section 4.2. θ is the grain boundary misorientation. θ_0 is the critical misorientation value between the high and low angle grain boundaries, and is generally taken as 15°.

It is commonly assumed that the driving pressure (P) mainly arises from the reduction of stored energy and grain boundary energy. Both the stored energy and grain boundary energy are related with dislocations. The former refers to the energy of dislocation accumulated in the vicinity of grain boundary, while the latter refers to the energy of dislocation constituting grain boundaries [45].

The driving pressure (P_S) provided by the stored energy are related with the density difference of dislocations located at the two sides of grain boundary [46]. i.e.,

$$P_S = \tau\Delta\rho \tag{3}$$

where $\Delta\rho$ is the density difference of dislocations inside and outside of the moving grain boundary. τ is the dislocation line energy, and $\tau = 0.5\ \mu b^2$. μ and b are the shear modulus and the Burger's vector, respectively.

The driving pressure (P_G) provided by grain boundary energy is usually much lower than P_S. P_G can be expressed as [47],

$$P_G = \gamma\kappa \tag{4}$$

where κ is the grain boundary curvature. γ is the grain boundary energy, and can be calculated by [48],

$$\gamma = \begin{cases} \gamma_0 & \theta \geq \theta_0 \\ \gamma_0\dfrac{\theta}{\theta_0}\left(1-\ln\left(\dfrac{\theta}{\theta_0}\right)\right) & \theta < \theta_0 \end{cases} \tag{5}$$

where γ_0 is the high-angle boundary energy which can be calculated by [49],

$$\gamma_0 = \frac{\mu b\theta_0}{4\pi(1-\nu)} \tag{6}$$

where ν is the Poisson ratio, μ the shear modulus.