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Microstructure evolution at the onset of discontinuous dynamic recrystallization: A physics-based model of subgrain critical size

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ABSTRACT

A physical model based on the evolution of subgrains size is proposed to describe the nucleation and growth processes during discontinuous dynamic recrystallization. The evolution of subgrains to viable recrystallization nuclei was found possible at very low strains. Afterwards, the number of stable nuclei considerably increased on a sigmoidal trend with strain and reached a saturated state at about 0.6 times the peak strain. The dependence of nucleation rate on strain was modeled using an Avrami-type equation and the driving force for the growth of recrystallized nuclei was similarly modeled in terms of strain. It is also shown that "site saturation" is the governing mechanism for the initiation of the discontinues dynamic recrystallization at the grain boundaries. The flow stress of the material was calculated using the law of mixture of recrystallized and unrecrystallized regions with fractional softening as the stress-partitioning factor. Satisfactory agreement between predicted and experimental results was obtained, thereby confirming the validity of the proposed model.

1. Introduction

Dynamic recrystallization (DRX) is termed as the recrystallization that occurs during high temperature deformation. This topic is of great fundamental and practical interest for metallurgists and has been the subject of a large number of investigations for many years [1–5]. Recrystallization during hot deformation is often subdivided into continuous dynamic recrystallization (CDRX) and discontinuous dynamic recrystallization (DDRX) [1]. The development of CDRX is associated with the formation of subgrains with low angle grain boundaries, which are then transformed to grains with high angle boundaries during straining [6]. For the DDRX process, which is the focus of the present paper, grain boundary bulging is considered as the initial step for the nucleation of recrystallized grains. After the nucleation step, the new DDRX grains grow to use up the deformed matrix. However, for bulging to take place, subgrains near the pre-existing grain boundaries should be well developed in order to provide the required driving force for local migration of the grain boundary [7–10]. Such critical condition for bulging is satisfied during deformation process and its determination with relation to DDRX has been the subject of extensive experimental and modeling investigations on flow stress and kinetics of recrystallization [11–16].

The dynamically recrystallized nuclei, defined as dislocation free volumes, form on pre-existing grain boundaries and grow inside the deformed surrounding matrix due to the dislocation density difference. However, new dislocations are generated due to the ongoing deformation within the DDRX grains and decrease the driving force for further growth. Therefore, the growth of nuclei decelerates the progress of recrystallization, and a new cycle of nucleation and growth is required for recrystallization to proceed. Microstructural observations have confirmed that the first layer of DDRX grains form a "necklace" structure of new grains at prior grain boundaries [17,18]. It has been observed that, at about the peak strain of a typical DDRX flow curve, the necklace structure is formed along most of the grain boundaries [19–21]. Under these conditions, for DDRX to progress further, new grains must be formed at the interface of the necklace structure and move inside the deformed matrix. The second, third and more nucleation cascades form consecutively, until the entire deformed grains are eventually consumed.

Microstructure evolution during DDRX has been analyzed experimentally by Ponge and Gottstein [22], Liu et al. [23], and Momeni et al. [24]. In parallel to the experimental investigations, others have employed physical-based models for this purpose [25–27].

In most of the published works, two types of methods are used to predict the critical strain or critical stored energy for the nucleation of dynamically recrystallized grains. Baily et al. used the

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critical dislocation density criteria for the formation of nuclei and associated it with the critical strain for DRX [28]. Poliak and Jonas, interpreted the critical condition according to the irreversible thermodynamics [12]. In their model, they assumed that the initiation of DRX is accompanied by flow localization, which is ultimately controlled by the stored energy and critical conditions. Gottstein et al. [13] predicted this critical strain based on a flow curve and dislocation work hardening model.

Although, evolution of the dislocation substructure occurs in all the above models; however, most of them do not describe the corresponding microstructural changes such as the formation of a necklace structure. In an early work, Cram and co-workers [27] described the nucleation condition of DDRX in terms of a physical-based model, which was originally developed to describe the initiation of static recrystallization. Their published data was, however, limited only to the prediction of flow stress at low strain rates in pure Cu. Montheillet et al. [29] predicted the flow stress of 304L stainless steel under steady state conditions by considering a grain scale semi-analytical approach that was including dynamic recovery, strain hardening, nucleation and growth of new grains. The model used different forms of dislocation strain hardening equations such as power law, Yoshie–Laasroui–Jonas, and Kocks–Mecking. However, the authors predicted the steady state flow stress and average grain size in 304L stainless steel without considering the nucleation and growth mechanisms at the early stages of recrystallization. In a similar manner, Damamme et al. [30] interpreted the steady state behavior and predicted dynamic recrystallization grain size in pure Ni.

Bernard et al. developed a new model of DDRX (called two-site mean field) by considering several basic physical processes such as dynamic recovery, dislocation density, work hardening, and grain boundary nucleation [31]. In their model, the competitive growth of grains and the effect of surrounding nucleus or grains were considered using two-site mean field model. Predicted flow stress curves, recrystallization kinetics, and average grain sizes were found in good agreement with the experimental data obtained for 304 stainless steel. Solas et al. [32] predicted the nucleation of recrystallized grain and migration of boundaries using a cellular automaton code, which was obtained based on analytical methods taking into account the misorientation and the local variation in stored energy and topological effects. Later on de Jaeger et al. [33] used the parameters proposed by Solas and determined the onset of DRX at low strains as low as 0.05 in Ni base superalloys. Their proposed mechanism for DRX was based on necklace formation at the initiation of DRX. The above described models cannot, however, be used to describe the initiation of recrystallization as well as the incorporation of nucleation and growth during DDRX.

The present paper expands the Cram's model [27] and provides a physics-based model to describe the nucleation and growth rates of nuclei at the early stages of DDRX. In addition, the model includes a microstructure evolution component incorporating subgrain size and dislocation density. Specifically, the model predicts the formation of a necklace structure at the incipience of the DDRX that actually depends on the deformation condition. Predictions made on nucleation and growth rates using the proposed model were validated by applying Cahn's model [34] and the law of mixture and compared with experimental results.

2. The model

2.1. Nucleation of recrystallized grains

During high-temperature deformation, subgrains are subjected to two opposite effects. The first one is the increasing strain that expands them and the second is the increasing stress that shrinks them. Indeed, subgrains grow during hot deformation until those with larger sizes, located adjacent to the boundaries, act as the sources of local bulging from which the nuclei of recrystallized grains are initiated [35]. The local bulging is due to the large difference between the dislocation densities on the two sides of the grain boundary. The required condition for bulging is met when a subgrain reaches a critical size. Due to the inhomogeneous deformation near the grain boundaries, subgrains have different potentials to undergo dynamic recovery (DRV). Some of them may be subjected to extensive DRV until the subgrain interior is swept away of dislocations. Upon successful bulging from the initial grain boundary these pioneering subgrains may become DRX nuclei. This can only occur if the driving force is high enough to overcome the grain boundary energy.

According to the classical nucleation theory, the critical subgrain size can be described using the following relation [7]:

$$r_c = \frac{2\gamma_{gb}}{\Delta E} \tag{1}$$

Here, γ_{gb} denotes the grain boundary specific energy and ΔE is the driving force for nucleation. This driving force originates from the difference in dislocation densities behind and ahead of the grain boundary. If the nucleus is considered strain-free, the driving force depends on the dislocation density in the matrix, and can be estimated as follows:

$$\Delta E = \alpha \rho G b^2 \tag{2}$$

where ρ is the dislocation density in the matrix, G is shear modulus of the material and b is the burgers vector.

The evolution of subgrain size with strain can be described using the following incremental functions:

$$\frac{dr}{d\varepsilon} = \left(\frac{dr}{d\varepsilon}\right)^{+} + \left(\frac{dr}{d\varepsilon}\right)^{-} \tag{3}$$

The terms on the right-hand side of Eq. (3) are attributed to the effects of plastic strain and stress on subgrain size, respectively.

2.1.1. Estimating the expansion of subgrains

In fact, subgrains migrate as dislocations attain them through straining. On the other hand, subgrain size depends on the applied stress and therefore shrinks as stress increases due to work hardening. Thus, the final subgrain size at a given strain and stress is the result of the reconciliation between the two terms on the right-hand side of Eq. (3).

The rate at which a subgrain coarsens, $v_{sb} = (dr/dt)^{+}$, can be described using the imposed driving force and the mobility of the subgrain boundaries (M_{sb}), given by:

$$V_{sb} = M_{sb} \cdot P_{sb} \tag{4}$$

Letting $P_{sb} = (2\gamma_{sb}/r)$ [7], then:

$$\left(\frac{dr}{dt}\right)^{+} = \frac{2\gamma_{sb} M_{sb}}{r} \tag{5}$$

Here, γ_{sb} is the specific energy of a subgrain boundary. The left-hand side of this equation can be written with respect to strain because $(dr/dt) = (dr/d\varepsilon)\cdot(d\varepsilon/dt)$. Therefore, Eq. (5) can be rewritten as:

$$\left(\frac{dr}{d\varepsilon}\right)^{+} = \frac{2\gamma_{sb} M_{sb}}{r\dot{\varepsilon}} \tag{6}$$