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Modeling dynamic recrystallization of L-605 cobalt superalloy

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ABSTRACT

The evolution of the microstructure of L-605 cobalt superalloy during high-temperature deformation is investigated using mean-field models combined with a reverse-engineering method. This method provides the unknown material parameters such as mobility and nucleation frequency, and it helps to quantify the occurrence of recrystallization. The effect of alloying elements on the mobility is quantified by comparison with the pure metal. Despite the large concentration of elements, the solute-drag effect can still be properly described by the Cahn model. A comparison with experimental results sheds new light on the occurrence of joint mechanisms that are usually neglected, such as self-heating and meta-dynamic recrystallization. The inclusion of these phenomena is not significantly more complicated when compared to the basic model, but it significantly improves the predictability for large strain rate.

1. Introduction

During the hot-working process of metallic materials, a succession of metallurgical phenomena takes place, leading to a severe transformation in their microstructure. Among them, dynamic recrystallization is a very common process, and it can be turned into an extremely useful tool for microstructural design [1,2]. For metals with low stacking-fault energy, dynamic recrystallization occurs by the nucleation and growth of new grains from the grain boundaries [3,4]. This process is known as discontinuous recrystallization [5,6], and it can be predicted using a mean-field model [7,8]. For pure metals, modeling is possible after a thorough analysis of a large experimental dataset, including rheology tests and microstructure observations [9]. The process may be far more complex for alloys: for instance, the migration process of interfaces may be hindered by solute elements [10], which can affect both the resulting grain size and the flow stress of the material [11].

This paper reports an investigation of dynamic recrystallization in a cobalt-based superalloy (Co-20Cr-15W-10Ni). The modeling work is supported by a large amount of experimental data [12], and the material parameters are determined from experiments using a reverse-engineering method. In the second part of the investigation, the parameters determined from experiments and modeling are used to predict the evolution of the microstructure during deformation.

While this method provides satisfactory results for a large range of deformation conditions, it shows some limitations at high strain rates. This decrease of reliability can be used as a sensor for detecting the occurrence of additional mechanisms not yet included in the model. The predominance of meta-dynamic recrystallization for the high-strain-rate stage is quantified, and the effect of solute elements is investigated.

2. Experimental observation of dynamic recrystallization

Modeling of recrystallization requires a significant amount of experimental data. On one hand, the high-temperature rheology of the material must be determined in order to deduce the stress and to estimate the average dislocation density. On the other hand, the determination of interfaces migration must be supported by microstructural observations.

The mechanical behavior was studied by hot-compression tests. Cylindrical specimen of Co–20Cr–15W–10Ni (wt.%) (L-605 alloy, Carpenter Technology Corp.) with a diameter of 8 mm and a height of 12 mm was used. Specimen was placed between two anvils of Si_3N_4 in a hot-working simulation tester (ThermecMaster Z machine, Fuji Electronic Industrial Co., Ltd, Japan). Foils of graphite and mica were used to decrease the heat loss and friction between the specimen and anvils. The specimens were heated by

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Fig. 1. Inverse pole figure (IPF) map of L-605 after the heating step and before compression.

induction until the temperature reached 1200 °C. The heating rate was 5 °C s^{-1}, and temperature was held constant for 10 min to homogenize microstructure and to dissolve potential precipitates. Then temperature was set between 1000 and 1200 °C for the compression test, and it was held 5 min to minimize the thermal gradient. Compression was conducted with a strain rate in the range 0.001–10 s^{-1} until reaching the true strain $\varepsilon=0.9$. The specimens were finally gas-quenched at a cooling speed of −50 °C s^{-1} to preserve the high-temperature microstructure.

The initial material was a cold-swaged bar with a grain size of 53 µm. After the heating step and just before the compression, specimen was cooled down to observe the microstructure. Microstructure was equiaxial, with a mean grain size of 85 µm measured by EBSD [12]. Microstructure consists of large annealed grains with many $\Sigma 3$ twin boundaries (Fig. 1). This microstructure is the effective initial state before compression.

As the focus of the present paper is the development of a physical model to predict microstructural evolution, the experimental data already published in [12] will not be discussed in detail. The usual shape of a compression curve is as follows: stress increases during deformation because of strain hardening, and reaches a maximum; flow softening then occurs because of dynamic recrystallization (Fig. 2). For large deformations, the flow stress σ stabilizes to a steady-state value σ_{ss} depending only on the deformation conditions.

Strain-hardening can be described by the Voce equation for the deformation stages I to III [13]:

$$\sigma(t) = \sigma_y + (\sigma_V - \sigma_y)(1 - e^{-\dot{\varepsilon}\cdot t\cdot\theta_0/(\sigma_V-\sigma_y)}), \quad (1)$$

where σ_y is the yield stress, and σ_V is the steady-state extrapolated stress. The extrapolation of compression curves from the initial hardening stage provides an estimation of the saturation stress σ_V. The difference between the saturation stress σ_V and the steady

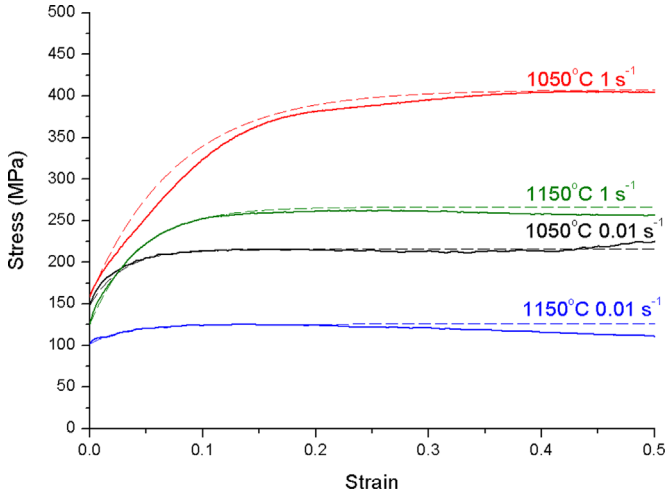

Fig. 2. Comparison of experimental stress data (solid lines) and data fitted with the Voce law (dashed lines).

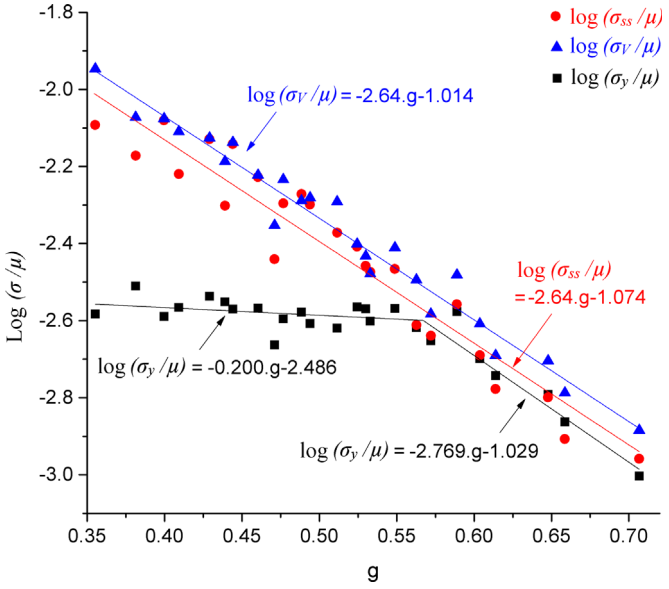

Fig. 3. Evolution of steady-state extrapolated stress σ_V, steady-state flow stress σ_{SS} and yield stress σ_y with the parameter g.

state stress σ_{ss} is attributed to the operation of softening mechanisms such as recrystallization.

The evolution of σ_V and σ_{ss} with changing deformation conditions can be estimated by the Kocks–Mecking model: flow stress depends on a parameter g that combines the effects of temperature and strain rate [13,14]. The following constitutive equation provides the steady-state stress as a function of g:

$$\log(\sigma/\mu) = k_1 \cdot g + k_2, \quad (2)$$

$$\text{with } g = \frac{k_B T}{\mu b^3}\cdot \ln\left(\frac{\dot{\varepsilon}_0}{\dot{\varepsilon}}\right), \quad (3)$$

In the above equation, k_1 and k_2 are material constants, k_B is the Boltzmann constant, μ is the shear modulus and $\dot{\varepsilon}_0$ is a floating parameter respectively.

Fig. 3 illustrates the experimental variation of $\log(\sigma_V/\mu)$ and $\log(\sigma_{ss}/\mu)$ with g for $\dot{\varepsilon}_0 = 10^9$ s^{-1}. This variation with g also holds for the yield stress. The yield stress decreases with g, and, therefore, it decreases with increasing temperature and decreasing strain rate. A sudden change in the slope occurs at $g=0.55$, which means that the mechanisms at stake suddenly change at this point. This phenomenon may be due to the pinning of dislocations by solute elements. At low temperature and low strain rate, the dislocation motion stops because of segregated solute atoms, until the stress exceeds the yield stress value. At high temperature and low strain rate, the solute atoms have sufficient time and mobility to diffuse, and they can move away from the dislocations. Therefore the dislocations slip more easily, leading to a decrease in the yield stress. This sudden unpinning of dislocations may be the reason for the change of the slope at $g=0.55$.

Fig. 2 represents extrapolated curves obtained from Eqs. (1) and (3) (dashed lines). The agreement with experimental curves (solid lines) is fair for strain below 0.3, and then a discrepancy occurs owing to the activation of softening mechanisms.

This estimation of the mechanical behavior must be combined with a study of the microstructural evolution to quantify the mechanisms at the origin of recrystallization. Fig. 4 represents the evolution of the size of recrystallized grain, recrystallized-area fraction, and flow softening with deformation parameters. The main conclusion of our previous work [12] is the occurrence of