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Flow stress model for IN718 accounting for evolution of strengthening precipitates during thermal treatment

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ABSTRACT

A flow stress model describing precipitate hardening in the nickel based alloy Inconel® 718 following thermal treatment is presented. The interactions between precipitates and dislocations are included in a dislocation density based material model. Compression tests have been performed using solution annealed, fully-aged and half-aged material. Models were calibrated using data for solution annealed and fully-aged material, and validated using data from half-aged material. Agreement between experimental data and model predictions is good.

1. Introduction

A dislocation density based material model has been coupled with a model for growth and coarsening of γ'' precipitates in the nickel based superalloy Inconel® 718 (hereafter referred to as IN718). The material model can be used in thermo-mechanical simulations of repair welding followed by heat treatment to predict residual stresses and final microstructure in aeroengine components. It is assumed that motion and immobilisation of dislocations are the most important mechanisms for plastic behaviour under these conditions. Modelling must account for interactions between dislocations and solutes as well as precipitates. The model has been calibrated and validated, using Gleeble compression tests at various temperatures for three variants of IN718. One is solution annealed, one is half-aged and the last is fully-aged. Half-aged material is so-named because its flow stress lies half way between the annealed and fully-aged materials. Test results from half-aged material were used for validation of the model. Agreement between model predictions and experimental data is good.

A brief overview of precipitate hardening of IN718 is given in the next section. This is followed by a description of the dislocation density based flow stress model. The various contributions to the flow stress are described with particular emphasis on the strengthening effects of precipitates. Models describing the evolution of mean particle size and particle volume fraction, and their relationship with flow stress, are presented. The results and a short discussion follow, leading to the conclusions.

2. Background

IN718 is a precipitate hardening alloy commonly used in aircraft engines, power plant and gas turbines. The most important strengthening mechanism is precipitation hardening that results from around 13 vol.% of the coherent ordered disc-shaped body-centred tetragonal (bct) γ'' phase, which comprises nickel and niobium (Ni_3Nb). A small hardening contribution from ordered fcc γ' precipitates (approx. 4 vol.%) is also present, but we ignore this in the current work since it is not the dominant strengthening precipitate [1]. The size of the precipitates depends on the ageing temperature and time of heat treatment, and has been measured by a number of authors, for example Camus and Engberg [2], Chaturvedi and Han [1] and Slama and Abdellaoui [3].

The various contributions to flow stress resulting from dislocation glide are well established. The work of Bailey and Hirsch [4] and Livingston [5] are early examples. The review by Gil Sevillano et al. [6] describes other studies. Bergström [7] integrated these contributions into a dislocation density based flow stress model that was later summarised in 1983 [8]. This paper and the work of Frost and Ashby [9] provide essential background to the model

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summarised below. It is assumed that the motion of dislocations is the most important contribution to plastic strain in a material as shown by Estrin [10].

3. Dislocation density flow stress model with precipitate hardening

Plastic deformation results from the introduction and movement of dislocations in the crystal lattice, and depends on their interaction with the material structure (notably immobile dislocations, solutes, precipitates, defects, etc.) that hinder their movement. It is common to assume that these obstacles provide contributions to the macroscopic flow stress that are additive, and described by expressions of the form:

$$\sigma_y = \sigma_i + \sigma_{HP} + \sigma_G + \sigma^* + \sigma_p \ldots \qquad (1)$$

where σ_y is the flow (yield) stress; σ_i is an internal friction stress caused by the movement of a dislocation through a perfect lattice (the Peierls–Nabarro stress); σ_{HP} is the grain size dependency (Hall–Petch effect) that includes the effect of stress concentrations at grain boundaries and the additional stress required to transfer plastic deformation across grains [11]; σ_G is an athermal stress resulting from long-range disturbances of the lattice caused by immobile dislocations; σ^* is a short-range interaction (the stress needed to move dislocations past short-range obstacles); σ_p results from the additional stress required to move dislocations around or through precipitates and solutes. There may also be other contributions, which are not considered significant in IN718. We ignore the internal friction stress and the Hall–Petch effect in the current model; they are taken into account in the initial dislocation density that is assumed in the long-range contribution that defines the virgin yield limit. This approach has similar features to the models of Fan and Yang [11] and Shenoy et al. [12]. The yield stress can then be written:

$$\sigma_y = \sigma_G + \sigma^* + \sigma_p \qquad (2)$$

Eq. (2) is easily implemented in a user subroutine in a finite element code, e.g. MSC.Marc. The radial-return algorithm can be applied in a straightforward manner in the same way as in plasticity models based on the concept that the stress state must remain on a yield surface during plastic straining. However, the yield surface is dependent on the plastic strain rate introduced by recovery terms described later. It is also possible to formulate the relations in a visco-plastic framework. Then various deformation mechanisms contribute to the plastic strain rate. Frost and Ashby [9] present a theory using this approach. The details of the dislocation density model developed here are summarised below.

3.1. Long range contribution

The long-range term σ_G in Eq. (2) is an athermal stress contribution that is independent of temperature, i.e. thermal vibrations cannot assist dislocations in overcoming disturbances in the lattice. It is written:

$$\sigma_G = \alpha M G b \sqrt{\rho_i} \qquad (3)$$

where ρ_i is the effective immobile (forest) dislocation density within the microstructure of the material during deformation, b is the Burgers vector, G is the shear modulus (which is temperature dependent) and M is the Taylor factor, which translates the effect of the resolved shear stress in different slip systems into effective stress and strain quantities. It is assumed to be constant and is taken to be 3.06 for fcc crystals [8]. The term α is a proportionality factor measuring the efficiency of dislocation strengthening, which lies in the range 0.2–0.8 [7,13]. It is assumed not to depend significantly on the dislocation substructure, although there is evidence that it may to a small extent [14,15]. We also ignore the temperature dependency of the Burgers vector.

The evolution of the immobile dislocation density model comprises three terms, one hardening (+) and two recovery (−) terms. The fist recovery term originates from glide, the second from climb:

$$\dot{\rho}_i = \dot{\rho}_i^{(+)} - \dot{\rho}_{i(glide)}^{(-)} - \dot{\rho}_{i(climb)}^{(-)} \qquad (4)$$

In this model, we do not consider different types of dislocations; all dislocations are immobile. The evolution of immobile dislocations is described in the following sections.

3.1.1. Hardening

It is assumed that immobilisation of dislocations is proportional to the density of mobile dislocations and the distance between each type of obstacle for their motion. Furthermore, the immobilisation of dislocations is assumed to be additive contribution from grain boundaries, cell walls and precipitates [15–17]. According to Orowan equation, see Eq. (11) below, the density of mobile dislocations and there average velocity is proportional to the plastic strain rate. This leads to:

$$\dot{\rho}_i^{(+)} = \frac{M}{b}\frac{1}{\Lambda}\dot{\bar{\varepsilon}}^p \qquad (5)$$

The introduced mean free path Λ of a moving dislocation is thus obtain from summation of the inverses of grain size g, the dislocation cell or subgrain diameter s, and the average distance between precipitates in the glide plane l_p:

$$\frac{1}{\Lambda} = \left(\frac{1}{g} + \frac{1}{s} + \frac{1}{l_p}\right) \qquad (6)$$

The initial grain size g_0 is assumed to be constant in the current model, i.e. $g = g_0$ before grain growth or recrystallization are taken into account. The cell (or subcell) diameter s, which is the prestage of forming a subgrain during the process of recovery [17], is assumed to be a characteristic length and is applicable to other Low Energy Dislocation Substructures (LEDS). In general, s is proportional to $1/\sqrt{\rho_i}$ [10]. We also assume that the mean free path is proportional to the particle spacing, l_p. It should be noted that this term is only important when the distance between two particles is smaller than the subcell diameter. If the distance between two subcells is smaller or equal to the distance between two particles, the prior term will dominate the mean free path, and precipitates will have a small influence on Λ. Combining Eq. (5) and the prior statements with Eq. (5), the rate of increase in immobile dislocations can be written as:

$$\dot{\rho}_i^{(+)} = \frac{M}{b}\left(\frac{\sqrt{\rho_i}}{K_{cell}} + \frac{1}{g_0} + \frac{K_{prec}}{l_p}\right)\dot{\bar{\varepsilon}}^p \qquad (7)$$

where the size of the subcell is related to the immobile dislocation density by the parameter K_{cell}. The parameter K_{prec} is related to how effective a precipitate is in immobilisation of a dislocation. Thus, the different mechanisms provide additive contributions to the increase in the number of immobile dislocations.

There are various approaches for estimating the effective distance between particles in the glide plane. For small particles, the dislocation line shears the precipitates and remains relatively straight. This assumption (among eight others) is used in Friedel's statistics [18]. It should be noted that this model is an approximate valid for infinitesimal particles. A model valid for finite particles, on the other hand, is the so-called square lattice spacing model. If there are n_s obstacles per unit area in the glide plane, the spacing between them is [19]:

$$l_p = \sqrt{\frac{1}{n_s}} = \sqrt{\frac{2\pi}{3f_p}}\bar{r}_p \qquad (8)$$