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A continuum model for the creep of single crystal nickel-base superalloys

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Abstract

In this paper, we develop a constitutive theory within a thermodynamic setting to describe the creep of single crystal superalloys that gainfully exploits the fact that the configuration that the body would attain on the removal of the external stimuli, referred to as the "natural configuration", evolves, with the response of the body being elastic from these evolving natural configurations. The evolution of the natural configurations is determined by the tendency of the body to undergo a process that maximizes the rate of dissipation. Here, the elastic response is assumed to be linearly elastic with cubic symmetry associated with the body which remains the same as the configuration evolves. A form for the inelastic stored energy (the energy that is 'trapped' within dislocation networks) is utilized based on simple ideas related to the motion of the dislocations. The rate of dissipation is assumed to be proportional to the density of mobile dislocations and another term that takes into account the damage accumulation due to creep. The model developed herein is used to simulate uniaxial creep of ⟨0 0 1⟩ oriented single crystal nickel-base superalloys. The predictions of the theory agree well with the available experimental data for CMSX-4.

Keywords: Nickel alloys; Creep; Constitutive equations; Theory and modeling; Natural configuration

1. Introduction

The demand for increased efficiency of gas turbines used in power generation and aircraft applications has fueled research into advanced materials for gas turbine blades. Higher efficiencies are possible if turbine blades can be designed to withstand inlet temperature of the order of 2200 °C or more. At such high temperatures, it is critical to use materials that have excellent resistance to creep. We shall start with a discussion of the salient features of the response of nickel-base superalloys in order to delineate the characteristics that need to be captured by the model.

Single crystal nickel-base superalloys have been developed for gas turbine blade applications. These alloys have superior thermal, fatigue and creep properties compared to conventional cast alloys because grain boundaries have been eliminated. Typical first generation alloys include CMSX-2, second generation alloys include CMSX-4, MC-2, TMS-63, and third generation alloys include CMSX-10 with high Re content.

A typical modern superalloy (e.g. CMSX-4) for turbine blades is a single crystal, which contains particles, based on the ordered γ' $L1_2$ structure, lying in a matrix based on a disordered face-centered cubic Ni_3Al. The γ' phase forms remarkably regular cubes packed in a rather regular cubic array and it occupies 65–70% of the volume. The two-phase structure of a superalloy contributes essentially to its excellent creep strength at high temperatures, the phase boundaries providing

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obstacles to dislocation motion. The volume fraction of the γ' phase is an important factor in optimizing superalloy composition to get the best creep strength. Usually a maximum in the creep strength is reached between 70% and 80% volume fraction of γ' phase with further increase leading to a significant drop in strength (see [1]).

It has been observed by Pollock and Argon [2] that in the primary stage of creep in modern superalloys, and during most of the secondary creep, plastic deformation is confined to the γ channels. The γ' particles act as impenetrable obstacles. The γ' phase has another very remarkable property. Whereas most metals and alloys, including the γ matrix, have flow stresses that decrease steadily with increasing temperature, alloys related to Ni_3Al, and many other alloys with the $L1_2$ structure, show flow stresses that can increase by a factor of 5 as the temperature increases from room temperature to about 650 °C [3]. The high strength of γ' is especially valuable at high temperatures.

The lattice parameters of the γ matrix and the γ' precipitate are very similar, but not identical. The creep deformed microstructure and many mechanical properties depend on the lattice misfit. The presence of various alloying elements strongly affects the value of misfit (see [4]). The misfit could be positive or negative depending on the particular composition of the superalloy. Moreover, the misfit changes with the kind of heat treatment the alloy is subject to and it also varies with temperature [5]. The sign of the misfit plays an important role in the evolution of microstructure as the material creeps (see [6]).

Another important microstructural property of superalloys is the ability of cubic γ' phase to transform into flat plates ("rafts") under the influence of stress and temperature. This directional coarsening is especially important in nickel base superalloys because the morphological changes in the two phase microstructure alter the creep resistance of the material in the stress and temperature range where these alloys are used in applications such as turbine blades. It has been shown by Nabarro [7] that in the elastic regime, the thermodynamic driving force for rafting is proportional to the applied stress, to the lattice misfit and to the difference of elastic constant of the γ and γ' phases. It has also been found that the direction of rafting depends upon the direction of loading and the sign of lattice misfit. Two types of rafting behavior in $\langle 0\,0\,1\rangle$ oriented nickel-base single crystals have been identified [8].

Type N-rafts develop transverse to the direction of the externally applied stress.

Type P-rafts develop parallel to the direction of the externally applied stress.

Type N behavior is usually associated with negative misfit alloys stressed in tension, or positive misfit alloys stressed in compression. Conversely, type P behavior is associated with positive misfit alloys stressed in tension, and negative misfit alloys stressed in compression. The differences in the microstructural evolution associated with a change from positive to negative misfit are an indication that the rafting is primarily dominated by internal stresses developed due to the misfit. In fact, the γ–γ' interface plays an important role in the creep property of superalloys [9]. The evolution of rafts with creep depends on the applied stress and operating temperature. At temperatures beyond 950 °C, experiments (Reed et al. [10]) suggest that rafting is complete during very early stages of creep deformation. Thus, from the modeling point of view, we can take into account, only the effect which a fully rafted microstructure confers. However, at lower temperatures, it is likely that the rafts evolve at a rate comparable with the rate of the evolution of the strain, in which case a suitable criterion for the evolution of the microstructure is needed.

The creep behavior of single crystal superalloys is highly anisotropic. The inherent crystallography of the single crystals leads to orientation dependent creep behavior. From a design point of view, it is imperative to use an orientation, which utilizes the maximum strength of the superalloy. In fact it is known that the creep strength of single crystal superalloys along $\langle 0\,0\,1\rangle$, which is also the preferred grain growth direction is favorable compared to $\langle 0\,1\,1\rangle$ or $\langle 1\,1\,1\rangle$ directions. There have been numerous experimental investigations into the creep behavior and the related microstructural aspects of $\langle 0\,0\,1\rangle$ oriented single crystal nickel base superalloys. Apart from this, experiments have also focussed on characterizing the behavior of single crystal turbine blades with centrifugal loading away from the exact $\langle 0\,0\,1\rangle$ orientation. Such situations are common in actual practice where misalignment of up to 15° [11] could occur due to a variety of reasons.

A number of studies have been devoted to studying the creep performance of $\langle 0\,0\,1\rangle$ oriented superalloy single crystals. At lower temperatures, particularly in the vicinity of 750 °C, a considerable amount of primary creep can occur (see [12,13]). At temperatures between 850 and 1000 °C, loading along $\langle 0\,0\,1\rangle$ yields a creep strain rate which increases monotonically with creep strain (i.e. tertiary creep is dominant), there being no evidence of a steady state regime (see [14–16]). At temperatures beyond 1000 °C, Reed et al., [10] reported that rafting of γ' phase occurs very rapidly and is complete in the very initial stages of creep deformation. After this stage, strain rate decreases with increasing strain for a considerable amount of time. Reed et al., [10] concluded that this strain hardening effect arises as a consequence of rafting of γ' phase. The strain rate in this temperature range keeps decreasing with increasing strain until a critical strain is reached. After the critical strain is reached, the creep strain rate increase sharply with strain with