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Damage induced anisotropy of polycrystals under complex cyclic loadings

T.B. Mounounga, A. Abdul-Latif *, D. Razafindramary

Laboratoire d'Ingénierie des Systèmes Mécaniques et des Matériaux (LISMMA), Supméca, 3, rue Fernand Hainaut, 93407 St. Ouen Cedex, France

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ABSTRACT

Motivated by a low-cyclic fatigue micromechanical model proposed recently [1], qualitative and quantitative studies are performed emphasizing the concept of damage induced anisotropy. In this model, the plastic strain and local damage variables are examined at the crystallographic slip system scale for FCC metallic polycrystals. Determined at the macroscopic scale, the elastic behavior is initially assumed to be compressible and isotropic. The anisotropic damaged behavior, caused by activation/deactivation concept, is adopted using a fourth-order damage tensor at the overall scale. Accordingly, the overall behavior, notably the deactivation phase due to the microcracks closure under complex cyclic loadings, is of particular interest in the study.

A host of plastic damaged behaviors of metallic polycrystals is predicted underlining the damage activation/deactivation effects on the multiaxial low-cyclic fatigue (LCF) behavior. Actually, the model is tested under strain- and stress-controlled conditions describing the effects of the loading complexity and the mean stress on the polycrystal LCF behavior. Finally, the model can successfully describe the LCF behavior of the Waspaloy at room temperature.

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1. Introduction

Currently, the elasto-inelastic behavior of metals shows an increasing maturity especially in the engineering theory of plasticity. Taking into account the damage mechanics notably under cyclic loading with the self consistent modeling is a hard task especially, due to the non-linearity of materials response. Some attempts have been made to describe the damaged-elasto-inelastic behavior of the material under different loading paths (e.g., [2–6]). Various damage categories have been described in the literature, such as creep damage, low cycle fatigue, high cycle fatigue and brittle damage [7–11], and many others.

The nonlinearity of material behavior is generally induced by plasticity and damage mechanics. Ductile polycrystalline materials usually fail as a result of nucleation, growth and coalescence of microdamages. Experimental observations show that the accumulation of microdamages has a tendency to form a localized damage, due to plastic strain localization up to the final structure failure. In fact, in several metallic materials, the kinematic strengthening is related to the creation of slip bands. The setting of these bands in the material induces undoubtedly an internal back stress in grains leading accordingly to an anisotropic behavior. Besides, TEM observations reveal strain localized in slip bands during cycling leading to an important dislocations density in these bands [12]. Microstructural observations related to specimen outer surfaces show that crack initiation occurs in some slip bands as in Waspaloy [13]. Thus, these slip bands together with microcracks seem to be important factors leading to an anisotropic behavior concerning the elastic and plastic strains.

Modeling of the spatial localization of the cyclic fatigue damage on a given structure is not a trivial task. In fact, modeling the fatigue damage localization within the specimen free surface via the finite elements methodology can be suitably performed in the framework of the non-local mechanics [14–17]. Note that localization analyses using a classic continuum model without incorporating the internal length concept is not adequate for modeling the intrinsic failure process. To remedy the problem of mesh dependency, different approaches have been developed; among these, the non-local models (e.g., [18–21]) are largely used. From a computational viewpoint, the non-local models appear to be straightforward to implement. Until now, the non-local formulation is separately used either in brittle damage models or in ductile plastic damage models [17].

An induced-oriented anisotropy phenomenon can experimentally be observed, in fatigue. In fact, microcracks may open or close depending on the applied loading direction. Thus, different responses can experimentally be observed for compression and tension loads, which lead to damage deactivation behavior as in [22] for an aluminum alloy. Theoretically, many approaches have been proposed since the last decade (e.g., [23–34]). Recently, a micromechanical model of damaged-elasto-inelastic behavior

* Correspondence to: Université Paris 8, IUT de Tremblay, 93290 Tremblay-en-France, France. Tel.: +33 1 41 51 12 34; fax: +33 1 48 61 38 17.
E-mail address: aabdul@iu2t.univ-paris8.fr (A. Abdul-Latif).

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for FCC polycrystals has been developed with small strain assumption, describing the damage activation/deactivation effect in an LCF [1]. The condition for damage activation/deactivation has been formulated and treated only at the macroscopic scale, using the mathematical projection operators. Through these operators, a fourth-order damage tensor is determined. This tensor can naturally describe the damaged behavior under multiaxial cyclic loadings and the associated phenomenon of the induced-oriented anisotropy.

The aim of the present work is to elucidate the developed model ability to describe the deactivation phenomenon, due to microcracks closure and its effect on polycrystals behavior. A host of plastic damaged behavior of metallic polycrystals is consequently predicted. The damage activation/deactivation and loading path effects on the multiaxial LCF behavior are examined out under different strain- and stress-controlled conditions. The corresponding non-linearity is appropriately described by the model under these loading situations. Note that special emphasis is laid to the biaxial cyclic loading paths, especially the non-proportional ones showing the additional hardening and damage evolution. Quantitatively, the model can successfully describe the LCF behavior of the Waspaloy at room temperature.

2. Modeling

The theoretical formulation of the developed model is presented in detail in [1]. Hence, a short description of the main features of the model is now given. For further discussion about the theoretical aspect, the reader is referred to the reference given above. Summarized in Table 1, this model adopts the small strain hypothesis. It uses three different levels namely, macroscopic (aggregate of grains), granular and microscopic (slip system). The elastic behavior is assumed to be initially isotropic and compressible and determined at the macroscale. The inelastic strain is based on the slip theory. An internal state variable of intragranular isotropic hardening (q^s, R^s) is defined at the slip system level. The slip rate can be determined as long as the shear stress (τ^s) and the hardening variables are known. It is also assumed that the intragranular damage variable (d^s, Y^s) initiates and then evolves at the slip system level, where the slip is highly localized. As a result, for each slip system, when the slip attains a certain value (threshold), the intragranular damage initiates ($d^s > 0$) and its final fracture takes place when d^s attains a critical value d^s_{cr}.

We begin by recalling the employed notations. In Table 1, the index $s \in \{1,2, \ldots, n\}$ is associated to the system rank, with n being the maximum number of octahedral systems in the grain ($n=12$ for an FCC). Similarly, the index $g \in \{1,2, \ldots, N_g\}$ describes the grain rank, with N_g being the maximum number of grains contained in the aggregate.

After determining the granular stress ($\underline{\underline{\sigma}}^g$) through the interaction law (Eq. (9)), the resolved shear stresses τ^s for all slip systems are calculated by the twice-contracted tensorial product between $\underline{\underline{\sigma}}^g$ and Schmid orientation second order tensor $\underline{\underline{m}}^s$ (Eq. (1)). The intragranular isotropic hardening coupled with damage, which is defined by Eqs. (2) and (3), describes the elastic domain expansion on the system (s). Q^s represents its modulus and b^s is a coefficient characterizing its non-linearity. The hardening interaction matrix H_{rs} is supposed to describe dislocation–dislocation interactions, i.e., allowing the introduction of the cross influence of the slip of the system (s) on the hardening of the system (r). A simple 12×12 matrix is chosen for FCC polycrystals. The value of pseudo-multiplier $\dot{\lambda}^s$ for each slip system coupled with damage is a power function of the distance to the yield point defined by Eq. (4). K^s and z^s are two parameters characterizing the viscous sensitivity of the material. Note that the local plastic flow

Table 1
Complete set of the model equations coupled with damage deactivation effect [1].

At the slip system level

$$\tau^s = \underline{\underline{\sigma}}^g : \underline{\underline{m}}^s \tag{1}$$

$$R^s = Q^s \sqrt{1-d^s} \sum_{r=1}^{n} H_{rs} q^r \sqrt{1-d^r} \tag{2}$$

$$\dot{q}^s = \frac{\dot{\lambda}^s}{\sqrt{1-d^s}}(1-b^s q^s) \tag{3}$$

$$\dot{\lambda}^s = \left(\frac{\tilde{f}^s}{K^s}\right)^{z^s} = \left(\frac{\frac{|\tau^s|}{\sqrt{1-d^s}} - \frac{R^s}{\sqrt{1-d^s}} - k_o^s}{K^s}\right)^{z^s} \tag{4}$$

$$\dot{\gamma}^s = \frac{\dot{\lambda}^s}{\sqrt{1-d^s}} \mathrm{sign}(\tau^s) \tag{5}$$

$$\dot{d}^s = \dot{\lambda}^s \left(\frac{Y^s}{S^s}\right)^{s_o^s} \frac{H(\lambda^s - \gamma_{th}^s)}{(1-d^s)^{\mathrm{ws}}} \sum_{r=1}^{n} \left[D_{\mathrm{rs}} \left(\frac{\overline{Y}^r}{S^r}\right)^{s_o^s+1} \frac{H(\lambda^r - \gamma_{th}^r)}{(1-d^r)^{\mathrm{ws}}} \right] \tag{6}$$

with

$$\gamma_{\mathrm{th}}^s = \left(1 - \frac{N_{\mathrm{psys}}}{N_{\mathrm{tsys}}}\right)^{-\alpha} \gamma_o^s \tag{7}$$

$$\overline{Y}^s = \frac{\frac{R^s}{\sqrt{1-d^s}} q^s \sqrt{1-d^s}}{2(1-d^s)} \tag{8}$$

At the granular level

$$\underline{\underline{\sigma}}^g = \underline{\underline{\Sigma}} + C^g \left\{ \sum_{h=1}^{\mathrm{Ng}} \upsilon^h \underline{\underline{\beta}}^h - \underline{\underline{\beta}}^g \right\} \tag{9}$$

$$\dot{\underline{\underline{\beta}}}^s = \dot{\underline{\underline{\varepsilon}}}^g_{\mathrm{in}} - a^g \underline{\underline{\beta}}^g \sum_{s=1}^{n} \dot{\lambda}^s \tag{10}$$

$$\dot{\underline{\underline{\varepsilon}}}^g_{\mathrm{in}} = \sum_{s=1}^{n} \frac{\dot{\gamma}^s}{\sqrt{1-d^s}} \underline{\underline{m}}^s \tag{11}$$

At the macroscopic level

$$\dot{\underline{\underline{E}}}_{\mathrm{in}} = \sum_{g=1}^{\mathrm{Ng}} \upsilon^g \dot{\underline{\underline{\varepsilon}}}^g_{\mathrm{in}} \tag{12}$$

$$\dot{D}^T = \sum_{g=1}^{N_D^{g'}} \upsilon_D^{g'} \sum_{s=1}^{n'} \frac{\dot{d}^s}{n'} \tag{13}$$

$$\underline{\underline{\Sigma}} = \sum_{i=1}^{3} \Sigma_i^* \underline{p}_i \otimes \underline{p}_i \tag{14}$$

$$\underline{\underline{Q}} = \sum_{i=1}^{3} \underline{p}_i \otimes \underline{p}_i \tag{15}$$

$$\underline{\underline{Q}}^+ = \sum_{i=1}^{3} H(\Sigma_i^*) \underline{p}_i \otimes \underline{p}_i \tag{16}$$

$$P_{ijkl}^+ = Q_{ia}^+ \, Q_{jb}^+ \, Q_{ka} \, Q_{lb} \tag{17}$$

$$\underline{\underline{\underline{\underline{D}}}} = D^T \underline{\underline{P}}^+ \tag{18}$$

$$\underline{\underline{\underline{\underline{R}}}}^d = \left(\underline{\underline{\underline{\underline{I}}}} - \underline{\underline{\underline{\underline{D}}}}^T \right) : \underline{\underline{\underline{\underline{R}}}}^o \tag{19}$$

$$\dot{\underline{\underline{\underline{\underline{R}}}}}^d = \left(\dot{D}^T \underline{\underline{\underline{P}}}^+ + D^T \dot{\underline{\underline{\underline{\underline{P}}}}}^+ \right) : \underline{\underline{\underline{\underline{R}}}}^o \tag{20}$$

$$\dot{\underline{\underline{\Sigma}}} = \dot{\underline{\underline{\underline{\underline{R}}}}}^d : \underline{\underline{E}}_e + \underline{\underline{\underline{\underline{R}}}}^d + \dot{\underline{\underline{E}}}_e + \dot{\underline{\underline{M}}} \tag{21}$$

$$\dot{\underline{\underline{M}}} = -\frac{1}{2} \dot{D}^T \frac{\partial \underline{\underline{\underline{\underline{P}}}}^+}{\partial \underline{\underline{E}}_e} : \underline{\underline{\underline{\underline{R}}}}^o : \underline{\underline{E}}_e : \underline{\underline{E}}_e - \frac{1}{2} D^T \left[\frac{\partial \dot{\underline{\underline{\underline{\underline{P}}}}}^+}{\partial \underline{\underline{E}}_e} \right] : \underline{\underline{\underline{\underline{R}}}}^o : \underline{\underline{E}}_e : \underline{\underline{E}}_e - D^T \frac{\partial \underline{\underline{\underline{\underline{P}}}}^+}{\partial \underline{\underline{E}}_e} : \underline{\underline{\underline{\underline{R}}}}^o : \dot{\underline{\underline{E}}}_e : \underline{\underline{E}}_e \tag{22}$$

on a slip system occurs when the absolute value of its resolved shear stress $|\tau^s/\sqrt{1-d^s}|$ is greater than the actual flow surface radius: $R^s/\sqrt{1-d^s}+k_o^s$ (Eq. (4)). k_o^s is the initial value of the critical resolved shear stress. Since the rate independent models (considering that plastic flow at the slip system level is a rate independent) do not possess the uniqueness in the numerical solutions. Thus, the rate dependent slip is adopted to resolve such numerical difficulties used previously by several researchers. Although, the rate independent case can almost be obtained by choosing a high value of viscous exponent z^s and a low value of the coefficient K^s, i.e., minimizing the viscosity effect. The intragranular damage evolution is expressed in Eq. (6) coupled with Eqs. (7) and (8). A new intragranular damage criterion (Eq. (7)) depends explicitly on the accumulated slip system as well as on

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