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# Theoretical modelling of ductile damage in duplex stainless steels – Comparison between two micro-mechanical elasto-plastic approaches

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ABSTRACT

Ductile damage is a consequence of large strains more or less localized inside bands. Taking into account damage in constitutive behaviour of metallic materials is necessary to model various engineering problems involved in forming processes (stamping, punching, shearing...). Damage can be described at macroscopic level with continuum mechanics theories but introducing microstructural features can lead to more accurate predictions. In the present study, two polycrystalline plasticity models including damage effects in the framework of scale transition methods are investigated. These models are based on different approaches with direct application to duplex stainless steel. The first one is a variant of the Lipinski–Berveiller model in which ductile damage effects have been introduced. The second one is a generalized Cailletaud model taking into account ductile damage. Because of the microstructural complexity of the chosen materials, some particular developments of the micro-mechanical approaches are considered. Moreover, continuous damage mechanics is used at grains scale including its coupling with plastic or elastic–plastic flow. The modelling is justified from previous experimental results obtained by neutrons diffraction on duplex stainless steels. The developed models allow then deducing from the grains behaviour the macroscopic behaviour of the aggregate with damage effects.



## 1. Introduction

Nowadays, forming and manufacturing processes of metallic materials are currently simulated numerically using FEA and assuming small elastic strains but large inelastic (plastic or visco-plastic) deformations. In order to improve the productivity of material models used with FEA, it is necessary to provide accurate constitutive equations, accounting for different kinds of hardenings weakly or strongly coupled with ductile damage. Such a methodology is now classical using various constitutive equations [1–4]. Due to more or less localized but large inelastic strains, the influence of ductile damage on material behaviour is required in the modelling from their microstructural features. This can be done using either a macroscopic approach where various state variables associated to the mechanical phenomena under concern are introduced at each quadrature point taken as homogenous RVE; or a micro-mechanical approach of polycrystalline inelasticity where the microstructural composition of the RVE has to be accounted for.

The present paper is specifically dedicated to the comparison between damage modelling from two scale transition methods in the framework of polycrystalline plasticity. Only ductile damage occurring in polycrystalline metals under large plastic deformation is considered in this work. The modelling of ductile damage and its effects on the mechanical behaviour of metallic materials can be achieved following two basic approaches. The Gurson based approach simply modifies the yield criterion by the voids volume fraction [5–7]. The CDM approach is based on thermodynamics of irreversible processes [8–10] with state variables and state and dissipation potentials. It can be directly adapted to the grains scale. Eventually, the easiest method to develop a multiscale modelling scheme including damage effect consists in using existing micro-mechanical models applied to plasticity [11–13] and then incorporating damage [1,2,4,8–10,12]. Among these useful models one can find the Cailletaud model [14] and the Berveiller–Zaoui model [15]. In the present work, both variant models are considered assuming large plastic deformation and applied to DSS, while neglecting any phase transformation phenomena, but taking into account ductile damage occurrence at appropriated scales. Although presented in small deformation to emphasize our damage coupling, the large deformation approach is eventually considered through frame indifference principle for each model. The first one is discussed down to numerical aspects, whereas the second one is studied only from a theoretical point of view. This second model is introduced to improve the first one because of effective difficulties in the modelling of damage. The present paper targets

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**Nomenclature**

| | |
|---|---|
| RVE | representative volume element |
| DSS | duplex stainless steel(s) |
| CDM | continuous damage mechanics |
| FEA | finite element analysis |
| CRSS | critical resolved shear stress |
| ODF | crystalline orientations distribution function |
| $X$ | is a scalar variable (with no line under the quantity $X$) |
| $\underline{X}$ | is a second rank tensor (with one line under the quantity $X$) |
| $\underline{\underline{X}}$ | is a fourth rank tensor (with two lines under the quantity $X$) |
| $A{\cdot}B$ | corresponds to a simple contraction operation between two tensors $A$ and $B$ |
| $A{:}B$ | corresponds to a double contraction operation between two tensors $A$ and $B$ |
| $A \otimes B$ | corresponds to a tensorial product between two tensors $A$ and $B$ |
| $\langle X \rangle$ | are the Mc-Cauley brackets which means the positive part of a scalar $X$ |
| $(\underline{X})^T or (\underline{\underline{X}})^T$ | is the transpose of the quantity $X$ (second or fourth rank tensor) |
| $\|\underline{X}\|$ | is the euclidean norm of a second rank tensor $\underline{X}$ |

advantages of both approaches for theoretical comparison, but not necessary numerical ones. The present work is wholly based on CDM framework as described in details in [16,17], whatever the considered model.

The studied material is an UR45N austeno-ferritic stainless steel, containing approximately 50% austenite and 50% ferrite. Its chemical composition was given in a previous publication [18]. It was obtained by continuous casting, and then hot rolled down to 15 mm sheet thickness. It was then annealed during 1000 h at a temperature of 400 °C and next cooled in air. The characteristic microstructure of such steel consists of austenitic elongated islands along the rolling direction and embedded in a ferritic matrix. EBSD method showed that all crystallites of ferritic phase have almost the same orientation $\{0\,0\,1\}\langle 1\,1\,0\rangle$, while austenitic islands are divided into smaller grains with different orientations of which the majority is concentrated around $\{0\,0\,1\}\langle 1\,0\,0\rangle$ and $\{0\,1\,1\}\langle 2\,1\,-1\rangle$, see in Fig. 1 [19].

## 2. Variant of the Lipinski–Berveiller model

### *2.1. Damage at mesoscopic scale*

A possible formulation for elasto-plastic models was initially proposed by Berveiller and Zaoui [15] and generalized by Lipinski and Berveiller [20,21]. Indeed, the Berveiller–Zaoui model is written in small strains and tangent modulus tensor, corresponding to the Hill's self-consistent equations, is supposed constant and isotropic. Whereas the Lipinski–Berveiller self-consistent scheme is the application of the original Hill's self-consistent idea at finite elasto-plastic strains in the mathematical framework proposed by Iwakuma and Nemat Nasser and using kinematic integral equation to determine the strain rate concentration tensors. So that such model describes the behaviour of polycrystalline materials for arbitrary large strains taking into account rotations of crystal lattice [20,21]. The latter model was used by many authors to predict elasto-plastic or elasto-viscoplastic deformation and texture evolution in polycrystalline materials [3,21–24]. It was also applied for interpretation of diffraction experiments for two-phase duplex steels [18]. Three scales are thus considered namely: macroscopic (aggregate), mesoscopic (grain) and microscopic (slip system) ones. In the present paper, this model has been modified in order to take into account ductile damage thanks to the total energy equivalence assumption [1] used at mesoscopic scale. It leads to introduce the effective total strain tensor $\tilde{\underline{\varepsilon}}^g$ and the effective stress tensor $\tilde{\underline{\sigma}}^g$ for each grain (g) of the aggregate, whatever the phase (austenite or ferrite quoted indifferently as "g"):

$$\tilde{\underline{\varepsilon}}^g = \underline{\varepsilon}^g\sqrt{1-d^g} \text{ and } \tilde{\underline{\sigma}}^g = \frac{\underline{\sigma}^g}{\sqrt{1-d^g}} \tag{1}$$

The damage parameter $d^g$ is taken as a scalar measure of ductile damage at mesoscopic scale (grain scale). In grains of the equivalent fictive undamaged RVE, the stress – total strain relation can be described using rates formulation as:

$$\dot{\tilde{\underline{\sigma}}}^g = \underline{\underline{l}}^g : \dot{\tilde{\underline{\varepsilon}}}^g \Longleftrightarrow \dot{\underline{\sigma}}^g = \tilde{\underline{\underline{l}}}^g : \dot{\underline{\varepsilon}}^g \tag{2}$$

where $\underline{\underline{l}}^g$ and $\tilde{\underline{\underline{l}}}^g$ are the tangent modulus tensors defined for undamaged and damaged grains, respectively. Substituting Eq. (1) in Eq. (2), the stress – total strain relation at the grain scale in each phase ("g") is then given by:

$$\dot{\underline{\sigma}}^g = (1-d^g)\underline{\underline{l}}^g : \dot{\underline{\varepsilon}}^g - \frac{1}{2}\left(\underline{\underline{l}}^g : \underline{\varepsilon}^g + \frac{\underline{\sigma}^g}{1-d^g}\right)\dot{d}^g \tag{3}$$

Assuming that $d^g$ can be function of the total strain and stress tensors $d^g = f(\underline{\varepsilon}^g, \underline{\sigma}^g)$, we obtain directly:

$$\dot{\underline{\sigma}}^g = \left((1-d^g)\underline{\underline{l}}^g - \frac{1}{2}\left(\underline{\underline{l}}^g : \underline{\varepsilon}^g + \frac{\underline{\sigma}^g}{1-d^g}\right)\otimes\frac{\partial d^g}{\partial \underline{\varepsilon}^g}\right) : \dot{\underline{\varepsilon}}^g - \frac{1}{2}\left(\underline{\underline{l}}^g : \underline{\varepsilon}^g + \frac{\underline{\sigma}^g}{1-d^g}\right)\left(\frac{\partial d^g}{\partial \underline{\sigma}^g} : \dot{\underline{\sigma}}^g\right) \tag{4}$$

While comparing with Eq. (2), the elasto-plastic tangent modulus for damaged grains $\tilde{\underline{\underline{l}}}^g$ can be then explicitly written, indifferently in ferritic phase or austenitic phase:

$$\tilde{\underline{\underline{l}}}^g = \left(\underline{\underline{1}} + \frac{1}{2}\left(\underline{\underline{l}}^g : \underline{\varepsilon}^g + \frac{\underline{\sigma}^g}{1-d^g}\right)\otimes\frac{\partial d^g}{\partial \underline{\sigma}^g}\right)^{-1} : \left((1-d^g)\underline{\underline{l}}^g - \frac{1}{2}\left(\underline{\underline{l}}^g : \underline{\varepsilon}^g + \frac{\underline{\sigma}^g}{1-d^g}\right)\otimes\frac{\partial d^g}{\partial \underline{\varepsilon}^g}\right) \tag{5}$$

where $\underline{\underline{1}}$ is the unit fourth rank tensor.

### *2.2. Damage at microscopic scale*

At microscopic scale, the yielding criterion describing plasticity has also to be modified to include damage in each phase, for example with a classical linear relation:

$$f^s_{Linear} = \tilde{\tau}^s_C - \left(\tau^s_{C0} + \sum_t H^{st}\tilde{\gamma}^t\right) \leqslant 0 \tag{6}$$

where $\tau^s_{C0}$ is the initial CRSS on the slip system "s", and $H^{st}$ is the hardening matrix [18] in the undamaged material. However, to describe an evolution of slip systems for non-linear behaviour, Voce law [25] has been frequently used especially with self-consistent calculations to interpret diffraction data (for examples see [26–28]). According to this approach, the CRSS for a given slip system is now related to plastic shear strain $\tilde{\vartheta}^g$ accumulated on all systems in the considered grain "g" in each phase, i.e.: