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Effect of Microstructural and Morphological 1 Parameters on the Formability of BCC Metal Sheets

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The determination of forming limit strains in sheet metal forming industry is a useful way 4 5 for quantifying metals in terms of formability. However, such forming limit diagrams (FLDs) 6 remain very difficult to obtain experimentally. Therefore, the numerical prediction of forming 7 limit strains represents a convenient alternative to replace this time consuming and expensive experimental process. Moreover, a combined theoretical-numerical model allows 8 investigating the impact of essential microstructural aspects (e.g., initial and induced 9 10 textures, dislocation density evolution, softening mechanisms, ...) and deformation mechanisms on the ductility of polycrystalline aggregates. In this paper, the impact of 11 12 microstructural and morphological parameters, particularly the mean grain size, on the 13 formability limit of BCC materials is investigated. To this end, an elastic-plastic self-14 consistent (EPSC) polycrystalline model, coupled with a bifurcation-based localization 15 criterion, is adopted to numerically simulate FLDs. The FLDs thus determined using the 16 bifurcation-EPSC model for an IF-Ti single-phase steel are compared to the FLDs given by 17 ArcelorMittal, demonstrating the predictive capability of the proposed approach in 18 investigations of sheet metal formability. The role of the averaging scheme is also shown to 19 be significant by comparing the critical limit strains predicted with the self-consistent scale-20 transition scheme to those obtained with the more classical full-constraint Taylor model. 21 Finally, numerical simulations for different values of mean grain size are provided in order 22 to analyze the impact of mean grain size on the formability of BCC metal sheets.

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24 1. Introduction

25 The concept of forming limit diagram (FLD) was first 26 introduced by Keeler^[1] and Goodwin^[2] in order to display 27 the critical limit strains leading to material failure for 28 different strain paths, varying from uniaxial to biaxial 29 tension conditions. The obtained curve gives a represen-30 tation of the in-plane components of the limit strains, in 31 which the major strain is plotted as a function of the minor 32 strain in the sheet plane. This conventional tool has long 33 been used to characterize the formability of sheet metals. 34 Because it is now widely recognized that the onset of 35 localized necking represents the main limitation of

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industrial forming processes, the FLDs are nowadays commonly determined at localization. 2

It should also be noted that in sheet metal forming processes several failure modes may occur (buckling, wrinkling, diffuse and localized necking), which can sometimes be coupled with damage phenomena. A unified approach considering all these mechanisms remains very difficult to achieve and hence the present work only focuses on the onset of strain localization due to macroscopic shear band formation.

Because experimental FLD measurements are a com-11 plex task and notably because of the strong influence of 12 several physical factors, such as microstructural and 13 textural anisotropy, various theoretical models have been 14 developed to predict localized necking. These are mainly 15 based either on imperfection theories, in particular the 16 Marciniak-Kuczynski (M-K) model^[3] and its generaliza-17 tion by Hutchinson and Neale,^[4] or on bifurcation 18 analysis, such as Rice's localization model.^[5,6] Note that 19 the M-K approach is widely adopted to simulate FLDs, due 20 to its flexibility and simple use for industrial applications; 21 however, the main drawback of this theory is the high 22 sensitivity of the results to some parameters such as the 23 initial thickness defect or the critical value of the threshold 24

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1 at which localization is said to occur. For these reasons, 2 the bifurcation approach has been preferred in the 3 current study, because it does not require any additional adjusting parameter and can be used within a fully three-4 5 dimensional framework. This bifurcation analysis is based 6 on the formation of strain localization bands correspond-7 ing to jumps of some mechanical fields across discontinu-8 ity interfaces.

9 The main objective of the current paper is to investigate 10 the impact of microstructural and morphological param-11 eters on formability limits of BCC materials, which can 12 provide guidelines for the design of new materials with 13 enhanced ductility properties. For this purpose, the 14 bifurcation approach will be coupled with an elastic-15 plastic self-consistent (EPSC) polycrystalline formulation.

The ability of the present multiscale model to accurately predict the macroscopic behavior of single-phase polycrystalline steels during monotonic and sequential loading paths has been shown in Franz et al.^[7,8]

First, the EPSC multiscale model and the bifurcation 20 theory are presented in Section 2. Then, the main results 21 22 obtained with the present model are presented in Section 3 23 in terms of strain localization analyses for a ferritic singlephase steel, where the impact of mean grain size is 24 investigated for a 1000-grain polycrystalline aggregate 25 similar to the ferritic single-phase steel IF-Ti. Finally, 26 27 some concluding remarks are drawn in Section 4.

28 2. Theoretical Framework

29 2.1. Local Elastic-Plastic Constitutive Modeling

A detailed presentation of the EPSC multiscale model can
be found in Franz et al.^[7,8] Only the main equations are
outlined here.

Three different essential mechanisms - twinning, phase 33 transformation and crystallographic slip - generally result 34 in irreversible deformation. The present model only 35 focuses on the crystallographic slip mechanism, and the 36 corresponding deformation of BCC metals will thus be 37 considered. For this type of materials, 24 independent slip 38 systems will be assumed potentially active, i.e., the slip 39 planes $\{110\}$ and $\{112\}$ and the slip directions $\langle 111 \rangle$. 40

The elastic distortion of the lattice and the plastic 41 flow due to slip on the crystallographic planes can be 42 considered as the most important aspects of single crystal 43 behavior. The adopted formulation is based on pioneering 44 contributions.^[9-15] The single crystal elastic-plastic con-45 stitutive law is written within the large strain framework 46 47 and is defined through the derivation of a tangent 48 modulus l relating the nominal stress rate n to the velocity 49 gradient g:

$$\dot{\mathbf{n}} = \mathbf{l} : \mathbf{g} \tag{1}$$

An additive decomposition of the velocity gradient **g** is 1 commonly used: 2

$$\mathbf{g} = \mathbf{d} + \mathbf{w} \tag{2}$$

where the symmetric part **d** designates the total strain rate and the anti-symmetric part **w** corresponds to the total 3 rotation rate. 4

These two parts can additionally be split into a plastic 5 and an elastic part as: 6

$$\mathbf{d} = \mathbf{d}^{\mathbf{e}} + \mathbf{d}^{\mathbf{p}}, \quad \mathbf{w} = \mathbf{w}^{\mathbf{e}} + \mathbf{w}^{\mathbf{p}}$$
(3)

The plastic part of the velocity gradient is related to the 7 slip rates $\dot{\gamma}^{g}$ by: 8

$$\mathbf{g}^{\mathbf{p}} = \mathbf{d}^{\mathbf{p}} + \mathbf{w}^{\mathbf{p}} = \sum_{g} \dot{\gamma}^{g} \vec{m}^{g} \otimes \vec{n}^{g}$$

$$\tag{4}$$

where \vec{m}^{g} is the vector parallel to the slip direction of the slip plane g with normal \vec{n}^{g} , $\dot{\gamma}^{g}$ is the associated slip rate. 9

From Equation (4), the plastic strain rate d^p and plastic 10 spin w^p can easily be written in terms of the Schmid 11 tensors \mathbf{R}^g and \mathbf{S}^g , defined as the symmetric and anti- 12 symmetric parts, respectively, of the tensor product 13 $\vec{m}^g \otimes \vec{n}^g$.

It is necessary to know the slip rates of active slip 15 systems in order to determine the expression of the local 16 tangent modulus I. The adopted procedure for the active 17 slip system selection will be briefly described here; more 18 details on this method and its validation can be found in 19 Franz et al.^[7] 20

For plastic behavior, the plastic flow rule for a given slip 21 system g is commonly expressed by distinguishing the 22 effectively active slip systems from those potentially active, 23 leading to the existence of several possible subsets of active 24 systems: 25

$$\begin{cases} \tau^{g} < \tau^{g}_{c} \Rightarrow \dot{\gamma}^{g} = 0\\ \tau^{g} = \tau^{g}_{c} \text{ and } \dot{\tau}^{g} \le 0 \Rightarrow \dot{\gamma}^{g} = 0\\ \tau^{g} = \tau^{g}_{c} \text{ and } \dot{\tau}^{g} > 0 \Rightarrow \dot{\gamma}^{g} \ge 0 \end{cases}$$

$$(5)$$

where the resolved shear stress acting on a given slip system g is given by: 26

$$\tau^g = \boldsymbol{\sigma} : \mathbf{R}^g \tag{6}$$

and its evolution can be expressed using the co-rotational derivative of the Cauchy stress tensor $\sigma^{\bigtriangledown}$: 27

$$\dot{\tau}^g = \boldsymbol{\sigma}^{\nabla} : \mathbf{R}^g \tag{7}$$

In order to select the active slip systems and derive their 28 slip rates within an elastic–plastic modeling framework, a 29

1 new approach, inspired by viscoplastic formulations, 2 allows replacing relation (5) with a rate-independent 3 regularization technique $\dot{\gamma}^g = k^g (\tau^g, \tau^g_c, \dot{\tau}^g) \dot{\tau}^g$

Finally, combining the previous relations with the single crystal elastic–plastic law given by Equation (1), the expression of the tangent modulus **l** is given in indicial notation by:

$$l_{ijkl} = C_{ijkl} - \frac{1}{2} \left(\sigma_{ik} \delta_{jl} - \sigma_{il} \delta_{jk} \right) - \frac{1}{2} \left(\delta_{ik} \sigma_{jl} + \delta_{il} \sigma_{jk} \right) - \sum_{g,h} \left(C_{ijmn} R_{mn}^{g} + S_{im}^{g} \sigma_{mj} - \sigma_{im} S_{mj}^{g} \right) M^{gh} \times \left(R_{pq}^{h} C_{pqkl} - R_{pq}^{h} \sigma_{pq} \delta_{kl} \right)$$
(8)

8 It can be observed that this local tangent modulus 9 exhibits elastic and plastic parts, which contain several 10 additional terms due to the large strain framework.

In the present model, it is assumed that the single crystal hardening law is given by the expression of the evolution of critical resolved shear stress rate $\dot{\tau}_c$ with slip rate $\dot{\gamma}$ for the whole system:

$$\dot{\tau}_c^g = \sum_{h=1}^{n_{gl}} H^{gh} \dot{\gamma}^h \tag{9}$$

in which summation is over the active slip systems, whose 15 number is n_{gl} , and where H^{gh} is the hardening interaction 16 matrix defining self-hardening and latent hardening.

17 Hardening is mainly due to mobile dislocation interactions 18 with lattice and pinned obstacles. During plastic strain, 19 dislocations are first created, stored and then annihilated 20 when their densities become sufficiently large. Kocks^[16] first 21 proposed a law describing the evolution of the dislocation 22 densities without a specified annihilation mechanism. This 23 law has been improved later by Essmann and Mughrabi^[17] 24 considering the annihilation of close dislocations:

$$\dot{\rho}^g = \frac{1}{b} \left(\frac{1}{L^g} - 2y_c \rho^g \right) \dot{\gamma}^g \tag{10}$$

where *b* is the magnitude of the Burgers vector, L^g is the mean 25 free path of dislocations on the slip system *g*, and y_c is the 26 critical annihilation distance of dislocations.

The mean free path can be related to the mean grain size 28 D and expressed thanks to a parameter g_0 specific to the 29 dislocation storage:

$$\frac{1}{L^g} = \frac{1}{D} + \frac{\sqrt{\sum_{h=1,h\neq g}^{n_{gl}} \rho^h}}{g_0}$$
(11)

The relationship between critical resolved shear stress and dislocation densities has been extended to the multislip case^[18] as:

$$\tau_c^g = \tau_0^g + \alpha \mu b \sqrt{\sum_{h=1}^{n_{\rm gl}} a^{\rm gh} \rho^{\rm h}}$$
(12)

where τ_0^g is the initial critical resolved shear stress, α is a constant related to the stability of the dislocation configurations, μ is the shear modulus, a^{gh} is the anisotropy interaction matrix, and ρ^{h} is the mean dislocation density for slip system *h*. The anisotropy interaction matrix introduced by Franciosi^[18] and expanded by Hoc^[19] will be used, in which the different components are defined by nine parameters depending on the nature of the dislocation interactions (e.g., coplanar or collinear systems...).

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Differentiating Equation (12) and using Equation (9) and (10), the hardening matrix is expressed as:

$$H^{\rm gh} = \frac{\alpha\mu}{2\sqrt{\sum_{k=1}^{n_{\rm gl}} a^{\rm gh}\rho^k}} a^{\rm gh} \left(\frac{1}{L^{\rm h}} - 2y_c \rho^{\rm h}\right) \tag{13}$$

2.2. One-Site Self-Consistent Approximation

In order to simulate the overall macroscopic response of polycrystalline aggregates, thanks to knowledge of the behavior of their individual constituents, a self-consistent scheme is employed. A detailed presentation of this averaging approach is given by Lipinski and Berveiller,^[14] Lipinski et al.^[15] and Franz et al.^[7] Only the main equations are outlined hereafter.

The incremental form of the single crystal constitutive law given by Equation (1) can still be used to express the macroscopic behavior law, making use of the macroscopic tangent modulus L, such as:

$$\mathbf{N} = \mathbf{L} : \mathbf{G} \tag{14}$$

The overall nominal stress rate \dot{N} and the overall24velocity gradient G for the aggregate are defined as the25volume averages of their local counterparts \dot{n} and g,26respectively:27

$$\mathbf{G} = \frac{1}{V} \int_{V} \mathbf{g}(\mathbf{x}) \, dv = \overline{\mathbf{g}(\mathbf{x})}$$

$$\dot{\mathbf{N}} = \frac{1}{V} \int_{V} \dot{\mathbf{n}}(\mathbf{x}) \, dv = \overline{\dot{\mathbf{n}}(\mathbf{x})}$$
(15)

In order to solve the averaging problem, the following 28 fourth-order concentration tensor is commonly introduced: 30

$$\mathbf{g}(\mathbf{x}) = \mathbf{A}(\mathbf{x}) : \mathbf{G} \tag{16}$$

A systematic expression for the macroscopic tangent 31 modulus is easily obtained by combining the local 32 behavior law, Equation (1), with Equation (14)–(16): 33

$$\mathbf{L} = \frac{1}{V} \int_{V} \mathbf{l}(\mathbf{x}) : \mathbf{A}(\mathbf{x}) \, dv = \overline{\mathbf{l}(\mathbf{x}) : \mathbf{A}(\mathbf{x})}$$
(17)

It is also assumed that the polycrystalline aggregate is composed of ellipsoidal grains with different crystallographic orientations and that for each individual grain *I*, the behavior and mechanical fields are homogeneous. The expression of the concentration tensor **A**^{*I*} related to grain *I* can then be written as:

$$\mathbf{A}^{I} = \left(\mathbf{I} - \mathbf{T}^{II} : \left(\mathbf{l}^{I} - \mathbf{L}\right)\right)^{-1} : \overline{\left(\mathbf{I} - \mathbf{T}^{II} : \left(\mathbf{l}^{I} - \mathbf{L}\right)\right)^{-1}}^{-1}$$
(18)

where T^{II} denotes the interaction tensor for grain *I*, which is related to Eshelby's tensor^[20] for an ellipsoidal inhomogeneity. For a polycrystalline aggregate comprising *Ng* grains with a respective volume fraction f^{f} , the one-site self-consistent expression corresponding to the selfconsistent scheme in the sense of Hill^[21] can be finally obtained as:

$$\mathbf{L} = \sum_{I=1}^{Ng} f^{I} \mathbf{l}^{I} : \mathbf{A}^{I}$$
(19)

Note that the classical full-constraint Taylor model can
be derived as a special case of the self-consistent scheme
by considering that the deformation within each grain is
equal to the macroscopic deformation. This amounts to
taking the fourth-order concentration tensor A equal to the
fourth-order identity tensor, which leads to an effective
modulus simply given by:

$$\mathbf{L} = \frac{1}{V} \int_{V} \mathbf{l}(\mathbf{x}) dv = \overline{\mathbf{l}(\mathbf{x})}$$
(20)

A last point concerns the integration within the model of the morphological and crystallographic evolutions for

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each grain during loading. The crystallographic and 1 morphological orientation of each grain is denoted 2 respectively by the Euler angles φ_1 , ϕ , φ_2 and φ'_1 , ϕ' , φ'_2 , 3 according to Bunge's notation.^[22,23] A schematic repre-4 sentation defining these three Euler's angles is given in 5 **Figure 1**. For the morphological orientation, these angles 6 define the orientation of the principal coordinate system 7 of the ellipsoid representing the grain relative to the 8 coordinate system (RD, TD, ND) of the sample. 9

The change of crystallographic orientation is due to 10 elastic rotation rate $w^{e,[13,24]}$ The evolution of Euler's 11 angles with the lattice rotation is governed by the following 12 equations: 13

$$\begin{cases} \dot{\varphi}_{1} = -w_{12}^{e} + \frac{\cos\phi}{\sin\phi} \left(w_{13}^{e} \cos\varphi_{1} + w_{23}^{e} \sin\varphi_{1} \right) \\ \dot{\phi} = -w_{23}^{e} \cos\varphi_{1} + w_{13}^{e} \sin\varphi_{1} \\ \dot{\varphi}_{2} = -\frac{1}{\sin\phi} \left(w_{13}^{e} \cos\varphi_{1} + w_{23}^{e} \sin\varphi_{1} \right) \end{cases}$$
(21)

With regard to the morphology of the grain, the 14 morphological orientation evolves with the total rotation 15 rate as: 16

$$\begin{cases} \dot{\varphi}'_{1} = -w_{12} + \frac{\cos\phi'}{\sin\phi'} (w_{13}\cos\varphi'_{1} + w_{23}\sin\varphi'_{1}) \\ \dot{\varphi}' = -w_{23}\cos\varphi'_{1} + w_{13}\sin\varphi'_{1} \\ \dot{\varphi}'_{2} = -\frac{1}{\sin\phi'} (w_{13}\cos\varphi'_{1} + w_{23}\sin\varphi'_{1}) \end{cases}$$
(22)

The shape of the grains is taken as an ellipsoid, which is 17 represented by its half-axes *a*, *b* and *c*. The morphological 18



Figure 1. Schematic representation defining the three Euler angles φ_1 , ϕ , φ_2 according to Bunge's notation.

1 evolution is due to the total deformation rate as:

$$\begin{pmatrix}
\dot{a} = a g_{11}^{\text{ell}} \\
\dot{b} = b g_{22}^{\text{ell}} \\
\dot{c} = c g_{33}^{\text{ell}}
\end{cases}$$
(23)

where \mathbf{g}^{ell} is the projection of the velocity gradient onto the 2 morphological frame.

3 2.3. Localization Bifurcation Criterion

4 The so-called Rice's localization criterion^[5,6] corresponds 5 to a bifurcation of the governing equations, which 6 is associated with admissible jumps for strain and stress 7 rates across a shear band as illustrated in **Figure 2**.

8 Because field equations have to be satisfied, the 9 kinematic condition for the strain rate jump is:

$$[\mathbf{G}] = \mathbf{G}^+ - \mathbf{G}^- = \mathbf{\kappa} \otimes \mathbf{v} \tag{24}$$

where κ denotes the jump amplitude and υ is the unit 10 normal to the shear band.

11 On the other hand, the continuity of the stress rate 12 vector has to be verified for the forces along the interface 13 created by the localization band:

$$\begin{bmatrix} \dot{\mathbf{N}}^{\mathrm{T}} \end{bmatrix} \cdot \mathbf{v} = \mathbf{0} \tag{25}$$

14 At the incipience of bifurcation, it is commonly 15 considered that the tangent modulus is the same on each



Figure 2. Localization of the deformation along a shear band.

part of the band. By using the constitutive law (14) along with Equation (24) and (25), the following condition is derived:

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$$(\mathbf{v} \cdot \mathbf{L} \cdot \mathbf{v}) \cdot \mathbf{\kappa} = \mathbf{0} \tag{26}$$

A non-trivial solution for the linear system of Equation (26), , corresponding to the occurrence of bifurcation and thus to the existence of at least one non-zero κ , is obtained when the following condition is satisfied:

$$det(\mathbf{v}\cdot\mathbf{L}\cdot\mathbf{v}) = 0 \tag{27}$$

The associated normal v, in the three-dimensional8space, defines the localization band orientation,9while the amplitude of the jump cannot be calculated10directly.11

3. Numerical Results – Prediction of Ductility Limits for BCC Materials

3.1. Material Parameter Identification and Validation

The following results are obtained for a ferritic singlephase steel, denoted IF–Ti, for which the identified material parameters are reported in **Table 1**. An initially random texture defined by 1000 crystallographic orientations is considered.

It is necessary to identify four parameters relative to the single crystal modeling. The three first parameters: initial critical resolved shear stress τ_0^g , parameter g_0 , which is related to the mean free path of dislocations, and critical annihilation distance of dislocations y_c are determined using only two mechanical tests, i.e., a uniaxial tensile test (or a simple shear test) and a reverse shear test. The mean grain size *D* can be easily identified using optical micrography.

The model is then validated by comparison with experimental stress-strain responses corresponding to various mechanical tests (uniaxial tensile test, simple shear test, plane strain tensile test, biaxial tensile test, balanced biaxial tensile test, Bauschinger test, orthogonal test) at different orientations with respect to the rolling direction. As depicted in **Figure 3**, modeling results are in agreement with the experimental ones for the IF-Ti steel.

Parameters	τ_{c0} [110]	τ_{c0} [112]	g 0	Ус	D
Values	55 MPa	55 MPa	90	3.25 nm	$20\mu{ m m}$

 $\label{eq:table_$



Figure 3. Comparison between the proposed model and the experiments for the studied IF-Ti single-phase steel for different linear and sequential loading paths performed perpendicular to the rolling direction (PST 10% SSh refers to a cross test consisting of 10% plane strain tension followed by simple shear, while BT refers to reverse shear tests at 10%, 20%, and 30% of shear prestrain).

1 3.2. Comparison with Reference Model

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In this Section, the Bifurcation-EPSC approach is applied to the IF-Ti steel for FLD prediction and the obtained diagram is compared to the FLD provided by ArcelorMittal, as reported in Figure 4. The ArcelorMittal FLD model^[25,26] can be considered as reference for comparison because it has proven its reliability in predicting formability for linear strain paths for a wide range of grades of sheet metals, for which experimental FLDs have been simultaneously measured and compared. 10



Figure 4. Simulated FLDs associated with linear strain paths for the IF-Ti single-phase steel obtained with bifurcation-EPSC and ArcelorMittal's models.

As shown in Figure 4, the FLD obtained with the 1 Bifurcation-EPSC model for the studied ferritic single- 2 phase steel is close to the ArcelorMittal FLD. Having 3 assessed the predictive capability of the present model in 4 the determination of forming limit strains, attention will be 5 directed now towards the investigation of the impact of 6 microstructural parameters on formability limits of BCC 7 materials. 8

3.3. Simulation of the Strain Paths

The FLD depicts the limit strain values determined for 10 different strain paths, covering uniaxial tension, through 11 plane strain tension, to equibiaxial tension. Different ways 12 are possible to define these strain paths, which correspond 13 to different choices for the applied loading or prescribed 14 boundary conditions. We propose here to investigate the 15 impact of such choices on the determination of FLDs for the 16 previously studied IF-Ti steel. The same set of parameters, 17 as reported in Table 1, is used for these simulations. 18

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In this Section, the different strain paths can be 19 obtained by prescribing a complementary set of compo- 20 nents for two parameterization tensors, in order to set the 21 associated boundary value problem. The first choice 22 consists of prescribing all strain-rate tensor components 23 as follows: 24

$$\mathbf{d} = \mathbf{d}_{11} \begin{bmatrix} 1 & 0 & 0 \\ 0 & \rho & 0 \\ 0 & 0 & -1 - \rho \end{bmatrix}$$
(28)



Figure 5. Simulated FLDs associated with linear loading paths for the IF–Ti single-phase steel obtained with the bifurcation–EPSC model: two different ways to prescribe the strain paths.

1 and the strain paths are defined by varying the in-plane 2 strain ratio ρ between -0.5 (uniaxial tensile test) and 1 3 (equibiaxial tensile test).

Another way to simulate the strain paths is by prescribing the in-plane strain components, with the same definition for the in-plane strain ratio ρ , in conjunction with the plane-stress conditions.

Figure 5 shows the limit strains as predicted by the two parameterization procedures. One can observe that the formability limits predicted with the second procedure (i.e., by imposing the in-plane strain components along with plane-stress conditions) are lower, in the whole, than those obtained with the first procedure (i.e., full strain components as given by Equation (28)). This trend is in general agreement with the fact that plane-stress conditions tend to precipitate strain localization (i.e., the consideration of the through-thickness stresses has been shown to delay the critical limit strains).

19 3.4. Influence of the Scale-Transition Scheme

20 The influence of the averaging scheme on the predicted
21 limit strains is investigated in this Section. Numerical FLDs
22 obtained with two different scale-transition schemes (i.e.,
23 the self-consistent model (bifurcation–EPSC model) and
24 Taylor's model (bifurcation–Taylor model)) for the studied
25 IF–Ti steel are reported in Figure 6 and compared.

One can observe that, for the whole range of loading paths, the limit strains predicted with the full-constraint Taylor model are found to be particularly overestimated in comparison with those predicted with the self-consistent model. The impact of the adopted homogenization rule on formability limit prediction for BCC and FCC materials has been recently investigated using rate-sensitive crystal



Figure 6. Simulated FLDs associated with linear loading paths for the IF–Ti single-phase steel as obtained with bifurcation–EPSC and bifurcation–Taylor models.

plasticity models in conjunction with the M–K approach.^[27,28] These works have shown that, in the biaxial-stretching domain, the full-constraint Taylor model predicts higher limit strains than those obtained with the self-consistent scheme for BCC materials.

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3.5. Effect of Mean Grain Size

Figure 7 illustrates the effect of the mean grain size on the
ductility limit of the studied IF–Ti steel. In this investiga-
tion, all the parameters in Table 1 are held constant except89the mean grain size D. It is known that a decrease in the
mean grain size produces higher-strength materials but, in11



Figure 7. Simulated FLDs associated with linear loading paths for the IF–Ti single-phase steel as obtained with the bifurcation–EPSC model: Effect of the mean grain size.

turn, induces a drop in ductility. This experimental
observation is well reproduced by the proposed model,
and the results shown in Figure 7 are in general agreement
with the above statement.

5 4. Conclusions

In this paper, an elastic-plastic self-consistent (EPSC) 6 polycrystalline model has been combined with Rice's 7 localization criterion to investigate the influence of 8 microstructural and morphological parameters on the 9 formability of BCC materials in sheet forming processes. 10 Numerical FLDs have been determined for ferritic single-11 phase steel, denoted IF-Ti, and compared to reference 12 FLDs provided by ArcelorMittal. To this end, the macro-13 scopic behavior law has been first accurately modeled in 14 order to take into account the most important softening 15 mechanisms. It has been shown that the EPSC multiscale 16 model correctly reproduces the stress-strain responses for 17 various mechanical tests (linear and non-linear loading 18 19 paths). The adopted bifurcation-based ductility criterion allows the prediction of FLDs, which are found to be close 20 to reference FLDs. It has also been shown that the self-21 consistent scheme predicts more realistic forming limit 22 strains than the full-constraint Taylor model does. 23

Finally, the influence of the mean grain size on formability has been investigated and the model predictions led to higher limit strains for larger mean grain size values, in agreement with experimental observations.

The proposed theoretical and numerical tool allows 28 ductility prediction of new materials at the very early stage 29 of the design of new grades of steel and thus provides a 30 useful tool for steelmakers. Its main interest is to allow 31 comparisons in terms of formability for various materials 32 and to reveal the impact of microstructural effects on 33 ductility. Therefore, it can be advantageously used to 34 optimize the ductility of new steels or to design materials 35 with desired formability. 36

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- 47 materials
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