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1 Effect of Microstructural and Morphological 2 Parameters on the Formability of BCC Metal Sheets

3 Gérald Franz, Farid Abed-Meraim,* and Marcel Berveiller

4 The determination of forming limit strains in sheet metal forming industry is a useful way
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
8 expensive experimental process. Moreover, a combined theoretical-numerical model allows
9 investigating the impact of essential microstructural aspects (e.g., initial and induced
10 textures, dislocation density evolution, softening mechanisms, ...) and deformation
11 mechanisms on the ductility of polycrystalline aggregates. In this paper, the impact of
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.

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

industrial forming processes, the FLDs are nowadays 1
commonly determined at localization. 2

It should also be noted that in sheet metal forming 3
processes several failure modes may occur (buckling, 4
wrinkling, diffuse and localized necking), which can 5
sometimes be coupled with damage phenomena. A unified 6
approach considering all these mechanisms remains very 7
difficult to achieve and hence the present work only 8
focuses on the onset of strain localization due to 9
macroscopic shear band formation. 10

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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at which localization is said to occur. For these reasons, the bifurcation approach has been preferred in the current study, because it does not require any additional adjusting parameter and can be used within a fully three-dimensional framework. This bifurcation analysis is based on the formation of strain localization bands corresponding to jumps of some mechanical fields across discontinuity interfaces.

The main objective of the current paper is to investigate the impact of microstructural and morphological parameters on formability limits of BCC materials, which can provide guidelines for the design of new materials with enhanced ductility properties. For this purpose, the bifurcation approach will be coupled with an elastic-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 theory are presented in Section 2. Then, the main results obtained with the present model are presented in Section 3 in terms of strain localization analyses for a ferritic single-phase steel, where the impact of mean grain size is investigated for a 1000-grain polycrystalline aggregate similar to the ferritic single-phase steel IF-Ti. Finally, some concluding remarks are drawn in Section 4.

2. Theoretical Framework

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 transformation and crystallographic slip – generally result in irreversible deformation. The present model only focuses on the crystallographic slip mechanism, and the corresponding deformation of BCC metals will thus be considered. For this type of materials, 24 independent slip systems will be assumed potentially active, i.e., the slip planes {110} and {112} and the slip directions $\langle 111 \rangle$.

The elastic distortion of the lattice and the plastic flow due to slip on the crystallographic planes can be considered as the most important aspects of single crystal behavior. The adopted formulation is based on pioneering contributions.^[9–15] The single crystal elastic-plastic constitutive law is written within the large strain framework and is defined through the derivation of a tangent modulus \mathbf{l} relating the nominal stress rate $\dot{\mathbf{n}}$ to the velocity gradient \mathbf{g} :

$$\dot{\mathbf{n}} = \mathbf{l} : \mathbf{g} \quad (1)$$

An additive decomposition of the velocity gradient \mathbf{g} is commonly used:

$$\mathbf{g} = \mathbf{d} + \mathbf{w} \quad (2)$$

where the symmetric part \mathbf{d} designates the total strain rate and the anti-symmetric part \mathbf{w} corresponds to the total rotation rate.

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

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

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

$$\mathbf{g}^p = \mathbf{d}^p + \mathbf{w}^p = \sum_g \dot{\gamma}^g \bar{\mathbf{m}}^g \otimes \bar{\mathbf{n}}^g \quad (4)$$

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

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

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

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

$$\begin{cases} \tau^g < \tau_c^g \Rightarrow \dot{\gamma}^g = 0 \\ \tau^g = \tau_c^g \text{ and } \dot{\tau}^g \leq 0 \Rightarrow \dot{\gamma}^g = 0 \\ \tau^g = \tau_c^g \text{ and } \dot{\tau}^g > 0 \Rightarrow \dot{\gamma}^g \geq 0 \end{cases} \quad (5)$$

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

$$\tau^g = \boldsymbol{\sigma} : \mathbf{R}^g \quad (6)$$

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

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

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

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

4 Finally, combining the previous relations with the single
 5 crystal elastic-plastic law given by Equation (1), the
 6 expression of the tangent modulus \mathbf{L} is given in indicial
 7 notation by:

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

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.

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

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

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:

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

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 γ_c is the
 26 critical annihilation distance of dislocations.

27 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:

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

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

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

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 expand-
 ed 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...).

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

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

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 \mathbf{L} , such as:

$$\dot{\mathbf{N}} = \mathbf{L} : \mathbf{G} \quad (14)$$

The overall nominal stress rate $\dot{\mathbf{N}}$ and the overall
 velocity gradient \mathbf{G} for the aggregate are defined as the
 volume averages of their local counterparts $\dot{\mathbf{n}}$ and \mathbf{g} ,
 respectively:

$$\begin{cases} \mathbf{G} = \frac{1}{V} \int_V \mathbf{g}(\mathbf{x}) dV = \overline{\mathbf{g}(\mathbf{x})} \\ \dot{\mathbf{N}} = \frac{1}{V} \int_V \dot{\mathbf{n}}(\mathbf{x}) dV = \overline{\dot{\mathbf{n}}(\mathbf{x})} \end{cases} \quad (15)$$

In order to solve the averaging problem, the following
 fourth-order concentration tensor is commonly intro-
 duced:

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

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

$$\mathbf{L} = \frac{1}{V} \int_V \mathbf{l}(\mathbf{x}) : \mathbf{A}(\mathbf{x}) dV = \overline{\mathbf{l}(\mathbf{x}) : \mathbf{A}(\mathbf{x})} \quad (17)$$

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 \mathbf{A}^I related to grain I can then be written as:

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

where \mathbf{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 N_g grains with a respective volume fraction f^I , the one-site self-consistent expression corresponding to the self-consistent scheme in the sense of Hill^[21] can be finally obtained as:

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

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 \mathbf{A} equal to the fourth-order identity tensor, which leads to an effective modulus simply given by:

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

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

each grain during loading. The crystallographic and morphological orientation of each grain is denoted respectively by the Euler angles $\varphi_1, \phi, \varphi_2$ and $\varphi'_1, \phi', \varphi'_2$, according to Bunge's notation.^[22,23] A schematic representation defining these three Euler's angles is given in **Figure 1**. For the morphological orientation, these angles define the orientation of the principal coordinate system of the ellipsoid representing the grain relative to the coordinate system (RD, TD, ND) of the sample.

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

$$\begin{cases} \dot{\varphi}_1 = -w_{12}^e + \frac{\cos\phi}{\sin\phi} (w_{13}^e \cos\varphi_1 + w_{23}^e \sin\varphi_1) \\ \dot{\phi} = -w_{23}^e \cos\varphi_1 + w_{13}^e \sin\varphi_1 \\ \dot{\varphi}_2 = -\frac{1}{\sin\phi} (w_{13}^e \cos\varphi_1 + w_{23}^e \sin\varphi_1) \end{cases} \quad (21)$$

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

$$\begin{cases} \dot{\varphi}'_1 = -w_{12} + \frac{\cos\phi'}{\sin\phi'} (w_{13} \cos\varphi'_1 + w_{23} \sin\varphi'_1) \\ \dot{\phi}' = -w_{23} \cos\varphi'_1 + w_{13} \sin\varphi'_1 \\ \dot{\varphi}'_2 = -\frac{1}{\sin\phi'} (w_{13} \cos\varphi'_1 + w_{23} \sin\varphi'_1) \end{cases} \quad (22)$$

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

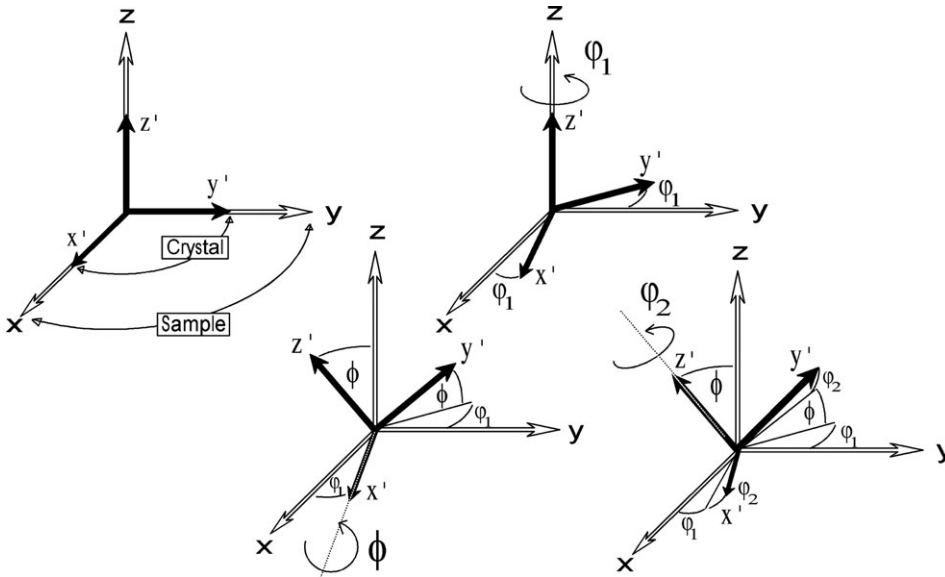


Figure 1. Schematic representation defining the three Euler angles $\varphi_1, \phi, \varphi_2$ according to Bunge's notation.

1 evolution is due to the total deformation rate as:

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

2 where \mathbf{g}^{el} is the projection of the velocity gradient onto the
3 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:

$$[\mathbf{G}] = \mathbf{G}^+ - \mathbf{G}^- = \boldsymbol{\kappa} \otimes \mathbf{v} \quad (24)$$

10 where $\boldsymbol{\kappa}$ denotes the jump amplitude and \mathbf{v} is the unit
11 normal to the shear band.

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

$$[\dot{\mathbf{N}}^T] \cdot \mathbf{v} = 0 \quad (25)$$

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

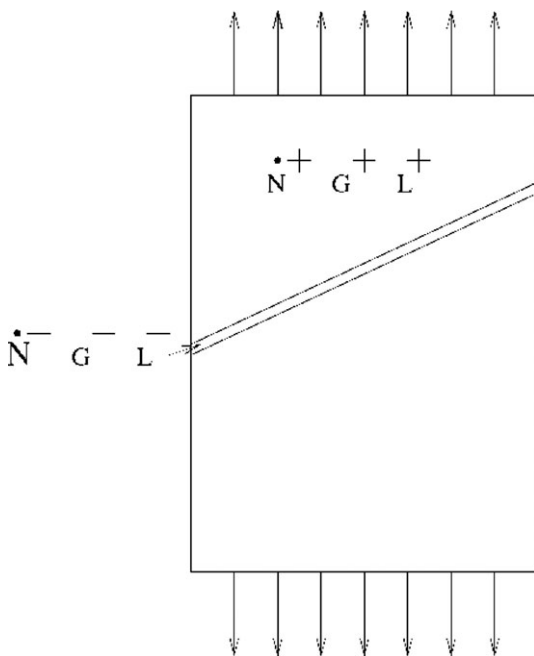


Figure 2. Localization of the deformation along a shear band.

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

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

4 A non-trivial solution for the linear system of Equation (26),
5 corresponding to the occurrence of bifurcation and thus to
6 the existence of at least one non-zero $\boldsymbol{\kappa}$, is obtained when the
7 following condition is satisfied:

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

8 The associated normal \mathbf{v} , in the three-dimensional
9 space, defines the localization band orientation,
10 while the amplitude of the jump cannot be calculated
11 directly.

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

3.1. Material Parameter Identification and Validation

12 The following results are obtained for a ferritic single-
13 phase steel, denoted IF-Ti, for which the identified
14 material parameters are reported in **Table 1**. An initially
15 random texture defined by 1000 crystallographic orienta-
16 tions is considered.

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

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

Parameters	τ_{c0} [110]	τ_{c0} [112]	g_0	y_c	D
Values	55 MPa	55 MPa	90	3.25 nm	20 μm

Table 1. Material parameters used for the IF-Ti single-phase steel.

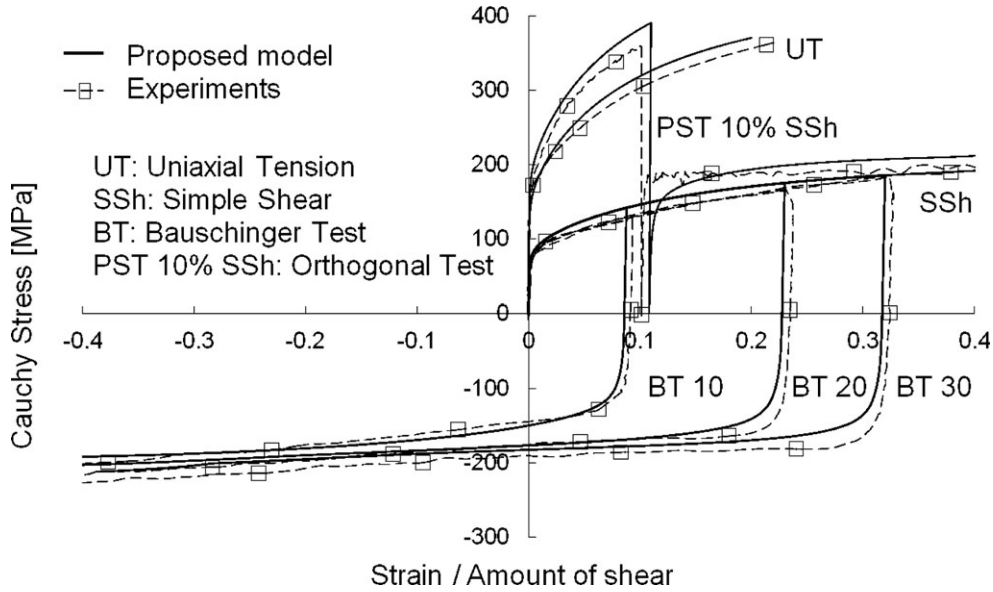


Figure 3. Comparison between the proposed model and the experiments for the studied IF-Ti single-phase steel for different linear and sequential loading paths performed perpendicular to the rolling direction (PST 10% SSh refers to a cross test consisting of 10% plane strain tension followed by simple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

3.2. Comparison with Reference Model

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.

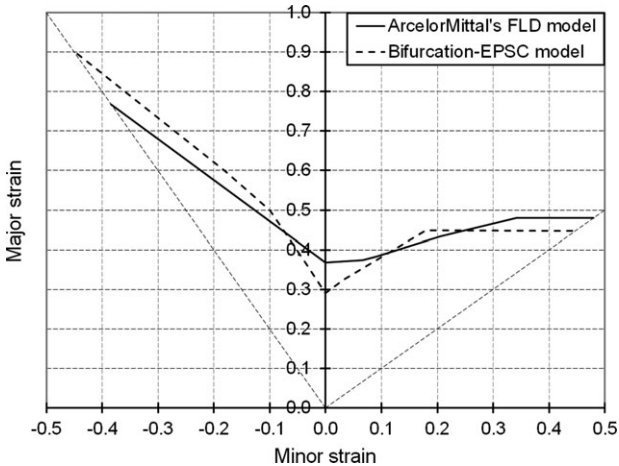


Figure 4. Simulated FLDs associated with linear strain paths for the IF-Ti single-phase steel obtained with bifurcation-EPSC and ArcelorMittal's models.

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

3.3. Simulation of the Strain Paths

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

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

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

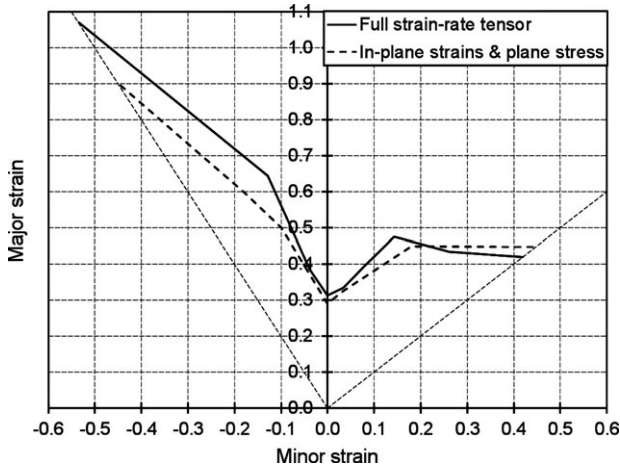


Figure 5. Simulated FLDs associated with linear loading paths for the IF-Ti single-phase steel obtained with the bifurcation-EPSC model: two different ways to prescribe the strain paths.

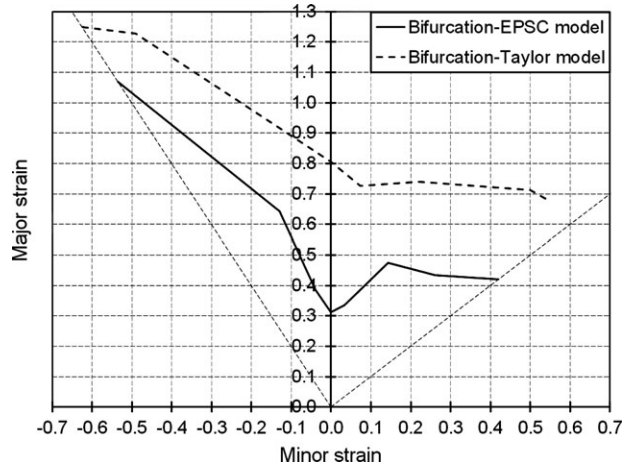


Figure 6. Simulated FLDs associated with linear loading paths for the IF-Ti single-phase steel as obtained with bifurcation-EPSC and bifurcation-Taylor models.

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).

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

8 **Figure 5** shows the limit strains as predicted by the two
 9 parameterization procedures. One can observe that the
 10 formability limits predicted with the second procedure
 11 (i.e., by imposing the in-plane strain components along
 12 with plane-stress conditions) are lower, in the whole, than
 13 those obtained with the first procedure (i.e., full strain
 14 components as given by Equation (28)). This trend is in
 15 general agreement with the fact that plane-stress con-
 16 ditions tend to precipitate strain localization (i.e., the
 17 consideration of the through-thickness stresses has been
 18 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.

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

plasticity models in conjunction with the M-K ap-
 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.

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 except
 the mean grain size D . It is known that a decrease in the
 mean grain size produces higher-strength materials but, in

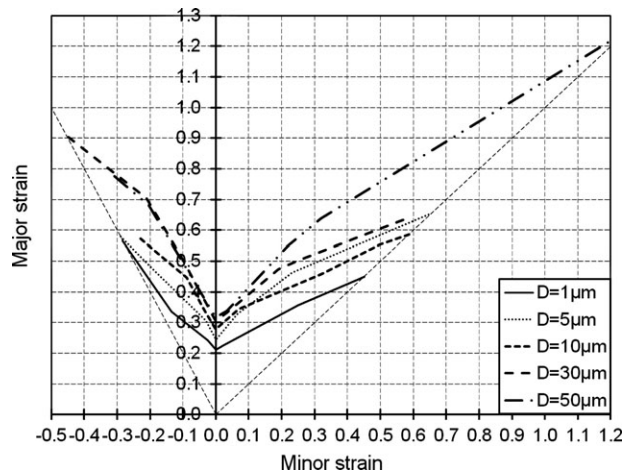


Figure 7. Simulated FLDs associated with linear loading paths for the IF-Ti single-phase steel as obtained with the bifurcation-EPSC model: Effect of the mean grain size.

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.

4. Conclusions

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

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 ductility prediction of new materials at the very early stage of the design of new grades of steel and thus provides a useful tool for steelmakers. Its main interest is to allow comparisons in terms of formability for various materials and to reveal the impact of microstructural effects on ductility. Therefore, it can be advantageously used to optimize the ductility of new steels or to design materials with desired formability.

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