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To cite this article: Frédéric Barlat 2018 *J. Phys.: Conf. Ser.* **1063** 012002

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Advanced Constitutive Modeling for Application to Sheet Forming

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Abstract. Continuum constitutive descriptions of plasticity suitable for finite element simulations of sheet forming processes are succinctly discussed. Although multi-scale approaches allow for a more explicit representation of the physical deformation mechanisms occurring at microscopic scales, they are usually not suitable for industrial applications because of the quick turnaround time needed for process design simulations. Therefore, advances in classical concepts such as plastic anisotropy and strain hardening are still very much in demand. This article describes possible ways to make use of multi-scale results for application to sheet metal forming simulations.

1. Background

The results of finite element (FE) simulations involving large plastic deformation such as in sheet forming depend on a large number of parameters. Beside numerical parameters, physical input such as boundary condition, contact and interface, and material behavior are playing a key role. This article deals only with the latter, more precisely, the influence of the constitutive description on forming simulation results. For instance, an accurate prediction of springback in U-draw bending for a simple rectangular blank made of advanced high strength steel is difficult to achieve [1]. Better results are usually obtained if an advanced constitutive model is employed. First, this requires the use of an elastic cord modulus that is a function of the accumulated plastic strain. Second, the plasticity model should consist of a non-quadratic anisotropic yield condition and an anisotropic hardening approach. The former is necessary because of plastic anisotropy and an accurate prediction of the plane strain flow stress, and the latter because of the forward-reverse loading occurring when the material flows over a die corner. It is well known that strain hardening with a large transient effect occurs during non-linear loading. Of course, this type of constitutive description requires the measurement of mechanical properties in different directions and for various stress states. Moreover, it should include a few cycles in a forward-reversal mode of deformation, as well as two-step type of experiments near cross-loading as, for instance, tension in the rolling direction (RD) followed by plane strain tension in the transverse direction (TD). All these tests require proper equipment with well-defined operation procedures.

Other simple but challenging simulations include cup drawing of a circular blank, indentation of a pre-stretched panel and expansion of a circular hole. In the case of cup drawing, the prediction of the earing profile, that is, the strain field in the product, requires a precise description of the material



behavior in stress states that are close to those encountered in the flange of a cup [2]. These states fluctuate between pure shear and uniaxial compression, in which the thickness strain variation is limited. For this purpose, assuming tension-compression symmetry, uniaxial tension tests are the easiest to conduct but a large number of loading directions should be considered, typically, every 15° from rolling. Even with an advanced anisotropic yield condition, some discrepancies with experimental results are expected if hardening is assumed to be isotropic because the material flowing near a die radius experiences a forward-reverse deformations cycle. In addition, if the material already possesses a dislocation structure from prior thermomechanical processing of the sheet, as usual for rigid container materials, the cup drawing process itself can be considered as a second step and involve additional anisotropic hardening effects. The indentation of a pre-stretched panel requires the use of an advanced constitutive model as well because the pre-strain and indentation correspond usually to two different stress states [3]. In addition, the variation of the elastic modulus must be characterized for a balanced biaxial stress state. Finally, for the flat hole expansion (HE) test, the radial thickness strain profile from the hole radius towards the specimen edge is highly sensitive to the selected constitutive model and the corresponding coefficients [4].

2. Advanced constitutive descriptions

For metal forming simulations at an industrial scale, the constitutive models must be continuum with coefficients measurable with mechanical testing equipment. Lower scale models such as, for instance, crystal plasticity allow the introduction of microstructural features. However, these models are too slow in large scale problems and the forming analysts who are performing the numerical forming simulations are usually not trained for this type of approaches. In addition, a microstructure is very complex and would require too many variables and coefficients to achieve a sufficient accuracy. For instance, crystal plasticity account for crystal structure, crystallographic texture and, more recently, dislocation mechanics. Yet, a material contains precipitates, dispersoids or other types of particles, and other features that are difficult to account for. Moreover, these features are very often not distributed homogeneously in the material, which adds to the complexity of lower scale descriptions. The advantage of continuum approaches is that the associated coefficients are obtained from the results of mechanical tests, which provide an average response of the entire microstructure.

Lower scale models can be very convenient to generate virtual experimental data that can be employed for the identification of a continuum model. This is particularly interesting when the experiments are difficult to conduct but this still requires the characterization of the microstructure. Perhaps, the best method to include microstructure information is to build features in the continuum model that closely or more loosely reproduce the general trends obtained with lower scale modeling. For instance, crystal plasticity demonstrates that, at first order, the yield condition for metal should be convex and that flow is associative, two features that are already widely accepted. Crystal plasticity also indicates that yield functions should be non-quadratic with a degree that depends on the crystal structure. Another example is provided in the next section.

2.1. Lower scale modeling

Classical crystal plasticity is concerned only with the infinitesimal shear produced by slip and the flow stress increases after dislocations glide and get trapped in the microstructure. The strength increase is usually provided through an evolution of the critical resolved shear stress in an analytical form or based on dislocation density evolution such as Kocks [5] and Mecking [6], which is strain path independent. The influence of strain path was considered only recently in crystal plasticity, based on the dislocation density evolution proposed by Rauch et al. [7]. In this approach called RGVB, the dislocation density is divided in two populations, forward and reverse, one that get trapped during forward loading and the other that leads to reverse slip. The forward population evolves according to the classical Kocks-Mecking model with an accumulation term and a recovery term while the second population tends to disappear due dislocation interactions and annihilation. The net effect is that, during load reversal, a fluctuating strain hardening is predicted. This approach was implemented by

Kitayama et al. [8] in the visco-plastic self-consistent crystal plasticity (VPSC) model initially developed by Lebensohn and Tomé [9]. Thus, it is possible to capture the Bauschinger effect and transient strain hardening effects during load reversal. Rauch et al. [10] also introduced in the RGVB model a third population of dislocation density, so-called latent, to describe the stress-strain behavior when the strain path change is close to cross-loading. While load reversal is characterized as a change in which all the slip systems remain identical but slip is reversed, cross-loading correspond to a change in which dislocation glide occurs on newly activated systems only. However, this extension of the RGVB model has not been implemented in crystal plasticity yet.

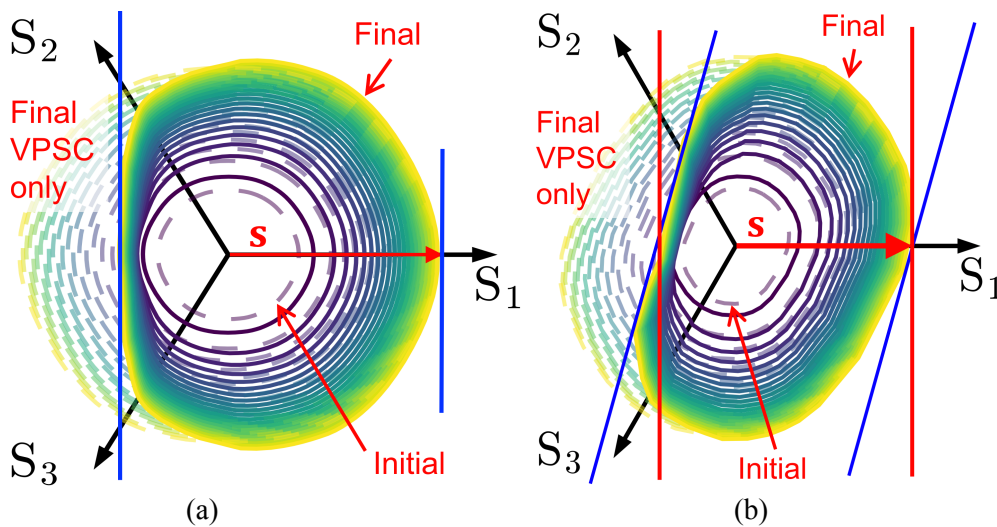


Fig. 1: Yield surface evolution observed in π –plane during uniaxial tension in RD; (a) isotropic texture and; (b) low carbon steel texture

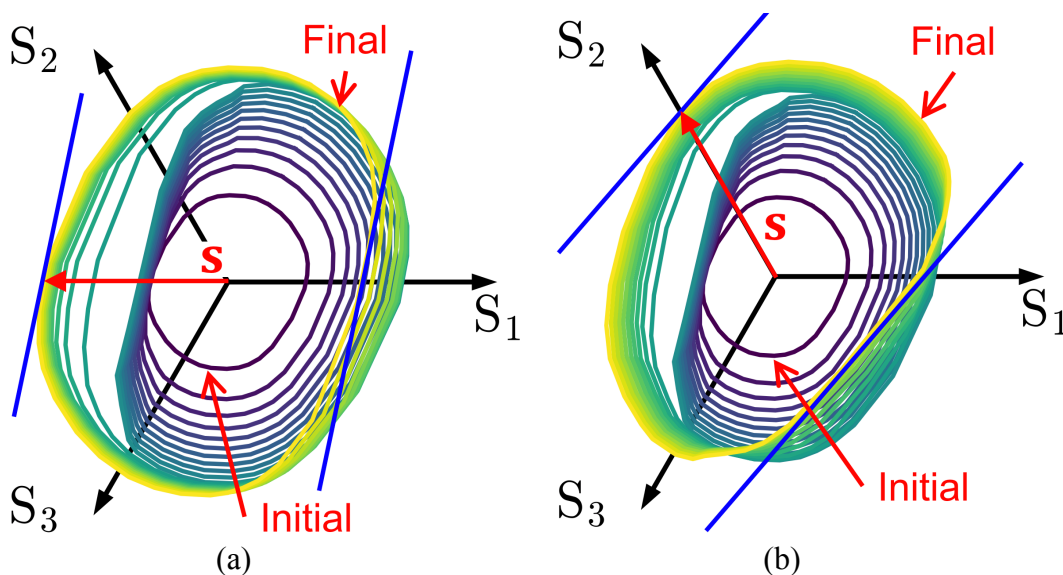


Fig. 2: VPSC-RGVB yield surface evolution for initial low carbon steel texture observed in π –plane during uniaxial tension in RD followed by second strain path: (a) stress reversal (uniaxial compression); (b) uniaxial tension in TD

Jeong et al. [11], conducted a series of simulations on isotropic and textured materials with the VPSC-RGVB model and observed a number of effects, in particular that, although a back-stress is introduced at the slip system level, a distortion occurs at the macroscopic scale of the yield surface during monotonic loading (Fig. 1). This distortion does not occur when the VPSC model only is used. Another observation is that the yield surface flattens in a direction that is orthogonal to the strain increment direction (yield surface normal at loading), not to the stress deviator. For load reversal (Fig. 2a) or for orthogonal tension (tension in RD then in TD), the yield surface flattening tends to occur in a direction orthogonal to the strain increment of the last loading mode (TD tension) and the memory of the prestrain tends to disappear, as observed by microscopic observations.

2.2. Continuum scale modeling

Under isotropic hardening, when a material is subjected to plastic deformation, the yield condition is described by the following relationship

$$\phi(\mathbf{s}) = \sigma_r(\bar{\varepsilon}) \quad (1)$$

where \mathbf{s} is the stress deviator. The yield function $\phi(\mathbf{s})$, isotropic or anisotropic, is a homogeneous function of first degree with respect to the stress. This means that in Eq. (1), $\phi(\mathbf{s})$ is an effective (or equivalent) stress and $\sigma_r(\bar{\varepsilon})$ is the reference stress-strain curve, in which $\bar{\varepsilon}$ is the effective strain based on the equivalence of plastic work $\phi d\bar{\varepsilon} = \boldsymbol{\sigma} : d\boldsymbol{\varepsilon}$. Anisotropic yield conditions under isotropic hardening have been widely used in the simulations of sheet metal forming. Many studies have shown that a proper choice of ϕ leads to better accuracy in the simulation results. Therefore, it is assumed in this work that the description of plasticity in Eq. (1) should remain the basis for an extension to anisotropic hardening.

A distortional plasticity model, so-called HAH, provides such a framework, extending Eq. (1) and allowing the modeling of the Bauschinger and other anisotropic hardening effects [12, 13]. The yield condition is expressed as

$$\bar{\sigma}(\mathbf{s}, f_-, f_+, \hat{\mathbf{h}}) = \{\xi(\mathbf{s})^q + \phi_h(\mathbf{s}, f_-, f_+, \hat{\mathbf{h}})\}^{\frac{1}{q}} = \sigma_r(\bar{\varepsilon}) \quad (2)$$

In the simplest case, $\xi(\mathbf{s})$ is equal to $\phi(\mathbf{s})$ in Eq. (1), the effective stress associated with the selected anisotropic yield function and, if $\phi_h = 0$, Eq. (2) reduces to Eq. (1). However, more generally, $\xi(\mathbf{s})$ is a slight modification of $\phi(\mathbf{s})$ allowing the introduction of cross-loading and latent hardening effects. Because both ξ and ϕ_h are homogeneous functions of first degree with respect to the stress, $\bar{\sigma}$ is an effective stress. ϕ_h distorts the anisotropic yield condition $\xi(\mathbf{s}) = \sigma_r(\bar{\varepsilon})$ in order to provide a description of the Bauschinger effect. In the original model [12], denoted HAH_σ , it is assumed that $\phi_h = \phi_{h1}$ with

$$\phi_{h1}(\mathbf{s}, f_-, f_+, \hat{\mathbf{h}}) = f_-^q \left| \hat{\mathbf{h}} : \mathbf{s} - |\hat{\mathbf{h}} : \mathbf{s}| \right|^q + f_+^q \left| \hat{\mathbf{h}} : \mathbf{s} + |\hat{\mathbf{h}} : \mathbf{s}| \right|^q \quad (\text{HAH}_\sigma \text{ model}) \quad (3)$$

f_- and f_+ are two state variables that control the Bauschinger effect. The so-called microstructure deviator $\hat{\mathbf{h}}$ is a normalized stress-like deviator variable that controls the location of the distortion. The initial microstructure deviator $\hat{\mathbf{h}}$ takes the value of the normalized stress deviator $\hat{\mathbf{S}}$ corresponding to the first strain increment. During deformation, $\hat{\mathbf{h}}$ always tends to align itself with the current stress deviator to mimic the microstructure evolution. The state variables f_+ and f_- can be expressed as a function of another set of state variables g_+ and g_- (with $\omega \equiv +$ or $-$), namely

$$f_\omega = \sqrt{\frac{3}{8} \left\{ \frac{1}{g_\omega^q} - 1 \right\}^{\frac{1}{q}}} \quad (4)$$

g_ω represents two state variables that vary between 0 and 1. The most important in this expression is that $0 \leq g_\omega \leq 1$ and $f_\omega \rightarrow 0$ when $g_\omega \rightarrow 1$. Other state variables are defined in the model as discussed in [12] and [13].

In the original HAH model in Eq. (3), the flattening of the yield surface tends to be is orthogonal to $\hat{\mathbf{h}}$. However, the crystal plasticity calculations discussed in Section 2 indicate that flattening of the yield surface should be orthogonal to the strain increment. Therefore, the following modification of the original Eq. (3) suggested in [11] was assumed by setting $\phi_h = \phi_{h2}$ with

$$\phi_{h2}(\mathbf{s}, f_-, f_+, \hat{\mathbf{h}}) = f_-^q \left| \frac{\hat{\boldsymbol{\epsilon}}_h : \mathbf{s} - |\hat{\boldsymbol{\epsilon}}_h : \mathbf{s}|}{\hat{\boldsymbol{\epsilon}}_h : \hat{\mathbf{h}}} \right|^q + f_+^q \left| \frac{\hat{\boldsymbol{\epsilon}}_h : \mathbf{s} + |\hat{\boldsymbol{\epsilon}}_h : \mathbf{s}|}{\hat{\boldsymbol{\epsilon}}_h : \hat{\mathbf{h}}} \right|^q \quad (\text{HAH}_\epsilon \text{ model}) \quad (5)$$

resulting in a model called HAH_ϵ . Essentially, this change produces a flattening of the yield surface in a direction orthogonal to $\hat{\boldsymbol{\epsilon}}_h$, the normal to the corresponding yield surface defined in Eq. (1). Note that for proportional loading, the material response for both formulations, HAH_σ and HAH_ϵ , is identical to that corresponding to isotropic hardening in Eq. (1), even if reverse loading, cross-loading and latent hardening effects are activated.

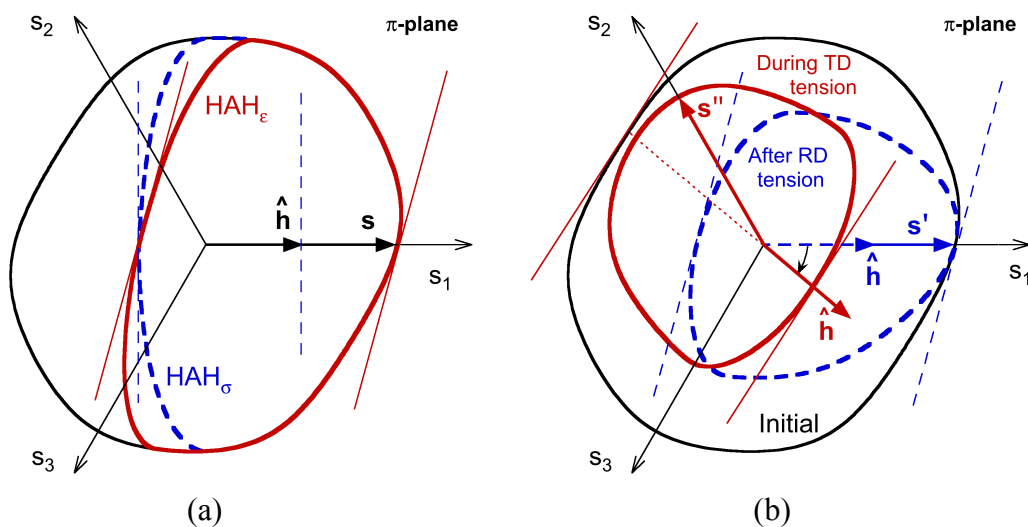


Fig. 3: HAH yield surface evolution observed in π –plane for textured material similar to that in Fig. 1(b); (a) during uniaxial tension in RD with HAH_σ and HAH_ϵ models; (b) during uniaxial TD tension after pre-strain in RD tension with HAH_ϵ model

As an illustration, Fig. 3a represents, in the π –plane, an example of yield surface evolution of a textured material similar to that of Figs. 1 and 2. These yield surfaces were predicted with HAH_σ and HAH_ϵ after RD uniaxial tension for a true strain of 0.08 and subsequent reloading up to a total accumulated strain of 0.16. In addition to plastic anisotropy, only the parameters controlling the Bauschinger effect were activated in this example (Fig. 3a). Fig. 3b represents the yield surface evolution predicted with the HAH_ϵ model corresponding to RD uniaxial tension for 0.08 followed by

uniaxial tension in the TD for the same amount of strain. In this example, the options for permanent softening and cross-loading contraction were also activated for illustration purpose but the trends regarding the flattening of the yield surface are consistent with those obtained with crystal plasticity in Fig. 2.

3. Discussion

The previous section indicates that the results of crystal plasticity combined with a dislocation density evolution model is a good guide to develop continuum approaches. It is likely that other features will emerge from multi-scale simulations and will be transferred to the continuum scale in similar ways. However, lower-scale models can be very useful for other purposes. In the HAH_σ and HAH_ε models, the identification of all the coefficients can be done sequentially for groups of parameters at a time, i.e., 1) isotropic hardening; 2) anisotropic yield function; 3) reverse loading and; 4) cross-loading. Five coefficients must be optimized to describe the case of reverse loading with Bauschinger effect and permanent softening. This number increases to 10 coefficients when other HAH features are turned on, i.e., the cross-loading effects. However, even the optimization of five coefficients can be a challenge. The optimization is solely done on the mathematical basis of an objective function, which is built with a limited number of stress-strain curves measured on a limited strain range for non-proportional loading. Moreover, the minimization of the objective function usually requires bounds for the coefficients within which the solution is assumed to occur. Therefore, the extrapolation space is limitless and a proper solution has a sense only if guided by physical considerations, hence by multi-scale modeling. The HAH distortional plasticity approach described in this article was selected to facilitate and illustrate the present discussion. However, these remarks are applicable to any other continuum plasticity model.

Acknowledgments

The author is grateful to POSCO for generous support and Dr. Youngung Jeong, Changwon National University, Korea, for his contribution to the figures.

References

- [1] Choi J, Lee J, Bong HJ, Lee MG and Barlat F 2018 Advanced constitutive modeling of advanced high strength steel sheets for springback prediction after double stage U-draw bending *Preprint* 10.1016/j.ijsolstr.2017.09.030
- [2] Yoon JW, Barlat F, Dick RE and Karabin ME 2006 *Int. J. Plasticity* 22, 174-193.
- [3] Lee JY, Lee M-G, Barlat F, Chung KH and Kim DJ 2016 *Int. J. Solids Structures* 87, 254-266.
- [4] Nakano H, Hakoyama T and Kuwabara T 2017 *Proc. Jap. Spring. Conf. Technology Plasticity* (Tokyo, JSTP) p 231-232
- [5] Kocks UF 1966 *Phil. Mag.* 13, 541-566.
- [6] Mecking H 1977 *Work-Hardening in Tension and Fatigue* ed AW Thompson (AIME, New-York).
- [7] Rauch EF, Gracio JJ and Barlat F 2007 *Acta Mater.* 55, 2939-2948.
- [8] Kitayama K, Tomé CN, Rauch EF, Gracio JJ and Barlat F 2013 *Int. J. Plasticity* 46, 54-69.
- [9] Lebensohn RA and Tomé CN 1993 *Acta Metall. Mater.* 41, 2611-2624.
- [10] Rauch EF, Gracio JJ, Barlat F and Vincze G 2011 *Model. Simul. Mater. Sci. Eng.* 19, 035009.
- [11] Jeong Y, Barlat F, Tomé C and Wen W 2017 *Int. J. Plasticity* 93, 212-228.
- [12] Barlat F, Gracio JJ, Lee MG, Rauch EF and Vincze G 2011 *Int. J. Plasticity* 27, 1309-1327.
- [13] Barlat F, Vincze G, Grácio JJ, Lee MG, Rauch EF and Tomé C 2014 *Int. J. Plasticity* 58, 201-218.