Towards an Accurate Springback Prediction

Experiments and Modeling

Ali Torkabadi

**TOWARDS AN ACCURATE SPRINGBACK PREDICTION
EXPERIMENTS AND MODELING**

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EXPERIMENTS AND MODELING**

DISSERTATION

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Summary

Finite element (FE) simulations are used extensively during the tool design stage to predict springback; it can then be compensated for by adapting the tool's geometry. The tool is adapted in such a way that the formed product takes the intended shape after springback. In order to accurately predict springback and compensate for it correctly, constitutive models are needed to accurately describe the material mechanics.

In this thesis, specific attention has been paid to the unloading behavior of advanced high strength steels (AHSS). Springback is governed by the stress-strain behavior of the material during unloading when the forming forces are removed. Therefore, modeling the unloading behavior of the material is of great importance to springback prediction.

It is generally accepted that the springback of the deformed material is driven solely by the recovery of elastic strain upon unloading; however, experimental evidence has shown that this is an invalid assumption. It has widely been observed that a plastically deformed material shows a nonlinear unloading/reloading behavior. Considering that the springback is governed by the total recovered strain upon unloading of the deformed part, modeling the unloading behavior is essential for an accurate springback prediction.

In this research, the main mechanisms responsible for the observed nonlinear unloading/reloading behavior are studied. This is carried out by performing a combination of theoretical, experimental and numerical studies on DP600 and DP800 from the family of AHSS.

To understand the physics of the nonlinear unloading/reloading behavior, uniaxial tensile tests are conducted. It is observed that the unloading/reloading behavior of the material is complex, showing direction dependency, time dependent behavior and sensitivity to baking treatment.

Based on the experimental results, it is concluded that there are two potential mechanisms behind the nonlinear unloading/reloading behavior: 1. dislocation driven anelasticity and 2. inhomogeneous deformation at the microscale. According to the theory of dislocation driven anelasticity, the reversible motion of the dislocation bow-outs contributes to an additional strain on top of elastic strain during unloading and reloading. This additional strain, known as anelastic strain, results in the observed nonlinear unloading/reloading behavior. A mixed physical-phenomenological model is proposed to describe the observed nonlinearity for different levels of pre-strain. The proposed model is generalized to a 3D constitutive model incorporating elastic, anelastic and plastic strains. The model is shown to be capable of predicting the stress-strain response of a DP800 steel subjected to unloading/reloading cycles.

An alternative theory is established on the inhomogeneous deformation at the microstructure. To this end, the stress and strain partitioning in a dual phase microstructure is analyzed using the crystal plasticity finite element modeling (CPFEM) approach. The model shows that some fractions of the material re-yield in compression during unloading. Based on the insight obtained from CPFEM, a model based on the elasto-plastic self-consistent (EPSC) homogenization scheme is proposed. In this model, the material inhomogeneity is modeled by considering a distribution in yield stress of material fractions. The EPSC model is shown to capture the nonlinear unloading/reloading behavior and Bauschinger effect simultaneously.

Draw-bend experiments are used as a benchmark for evaluating the performance of the developed models in predicting the springback of DP800. The draw-bend setup was designed and built during this research and represents a realistic forming process. The draw-bend experiments are simulated using the newly developed models and the results are compared with the classical E-modulus degradation model and the case where the E-modulus is taken as a constant. The results show that modeling the nonlinear unloading/reloading behavior results in a more accurate springback prediction in comparison with the classical approaches.

Samenvatting

Eindige elementen (FE) simulaties worden op grote schaal gebruikt tijdens de ontwerpfase van gereedschappen om de terugvering van het product te voorspellen en ervoor te compenseren door de geometrie van het gereedschap aan te passen. Het gereedschap is zodanig aangepast dat het omgevormde product na de terugvering de beoogde vorm aanneemt. Om de terugvering nauwkeurig te voorspellen en er correct voor te compenseren, zijn er constitutieve modellen nodig om het materiaalgedrag nauwkeurig te beschrijven.

In dit proefschrift is specifieke aandacht besteed aan het terugveergedrag van geavanceerde hogesterkte staalsoorten (AHSS). De terugvering wordt bepaald door de spannings-rek-relatie van het materiaal tijdens het ontlasten wanneer de omvormkrachten worden verwijderd. Daarom is het modelleren van het ontlastingsgedrag van het materiaal van groot belang voor het nauwkeurig voorspellen van terugvering.

De traditionele aanname is dat de terugvering van het vervormde materiaal enkel en alleen wordt bepaald door de elastische rek bij het ontlasten; experimenteel bewijs heeft echter aangetoond dat dit een onjuiste aanname is. Er is op grote schaal waargenomen dat een plastisch vervormd materiaal niet-lineair gedrag vertoont tijdens ontlasten en opnieuw belasten. Aangezien de terugvering wordt bepaald door de totale rek tijdens het ontlasten van het vervormde deel, is het modelleren van het ontlastingsgedrag essentieel voor een nauwkeurige voorspelling van de terugvering.

In dit onderzoek worden de belangrijkste mechanismen onderzocht die verantwoordelijk zijn voor het waargenomen niet-lineaire gedrag tijdens het ontlasten en opnieuw belasten van een gedeformeerd product. Hiertoe wordt een combinatie van theoretische, experimentele en numerieke studies met DP600 en DP800 uitgevoerd.

Om de fysica van het niet-lineaire gedrag tijdens ontlasten en opnieuw belasten

te begrijpen, zijn uni-axiale trekproeven uitgevoerd. De conclusie is dat het gedrag tijdens ontlasten en opnieuw belasten van het materiaal complex is. Het materiaal vertoont richtingafhankelijkheid en tijdafhankelijkheid en wordt beïnvloedt door de ondergane warmtebehandeling.

Op basis van de experimentele resultaten kan geconcludeerd worden dat er twee mogelijke mechanismen zijn achter het niet-lineaire gedrag tijdens het ontlasten en opnieuw belasten: 1. dislocatie-gedreven anelasticiteit en 2. inhomogene vervorming op microschaal. Volgens de theorie van dislocatie-gedreven anelasticiteit draagt het omkeerbare uitbuigen van de dislocatie bij aan extra rek bovenop de elastische rek tijdens ontlasten en opnieuw belasten. Deze extra rek, bekend als anelastische rek, resulteert in het waargenomen niet-lineaire gedrag tijdens het ontlasten en opnieuw belasten. Een gemengd fysisch-fenomenologisch model is ontwikkeld om de waargenomen niet-lineariteit voor verschillende niveaus van initiële rek te beschrijven. Het model is gegeneraliseerd naar een 3D-constitutief model dat rekening houdt met elastische, anelastische en plastische rek. Het model blijkt in staat te zijn om adequaat de spannings-rek-relatie te voorspellen van een DP800-staal dat onderworpen is aan cycli van ontlasten en opnieuw belasten.

Een alternatieve theorie voor de inhomogene vervorming in de microstructuur is ontwikkeld. Hiertoe wordt de distributie van spanning en rek in een microstructuur met twee fasen geanalyseerd met behulp van een eindige elementen methode gebaseerd op kristal-plasticiteit (CPFEM). Het model laat zien dat sommige delen van het materiaal tijdens het ontlasten opnieuw plastisch vervormen tijdens compressie. Op basis van het verkregen inzicht is een model ontwikkeld dat gebaseerd is op een elasto-plastisch, zelf-consistent (EPSC) homogeniseringsschema. In dit model wordt de inhomogeniteit van het materiaal gemodelleerd door een distributie in vloeispanning van materiaalfracties. Het EPSC-model blijkt in staat het niet-lineaire gedrag tijdens ontlasten en opnieuw belasten en het Bauschinger-effect adequaat te voorspellen.

Draw-bend experimenten worden uitgevoerd voor het evalueren van nauwkeurigheid van de ontwikkelde modellen voor het voorspellen van de terugvering van DP800. Het draw-bend experiment is tijdens dit onderzoek ontworpen en gebouwd en representeert een realistisch vervormingsproces. De draw-bend experimenten zijn gesimuleerd met de nieuw ontwikkelde modellen, en vergeleken met de beschikbare klassieke modellen voor het voorspellen van terugvering. De resultaten tonen aan dat het modelleren van het niet-lineaire gedrag tijdens ontlasten en opnieuw belasten leidt tot een meer nauwkeurige voorspelling van de terugvering, ten opzichte van de klassieke benadering.

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Introduction

1.1 Springback in sheet metal forming

Sheet metal forming enables the creation of light parts with complicated geometries at low cost and therefore has been widely utilized in the automotive, aerospace and domestic appliance industry. However, the process inherently suffers from dimensional inaccuracy due to springback. Springback is a change of part geometry after removing the forming forces. It is driven by the internal stresses developed during the forming process. Dimensional inaccuracy leads to issues in the subsequent forming operations, assembly and the quality of the final product [93].

In practice, springback is controlled in two ways: 1. by increasing the sheet tension by applying a larger restraining force; 2. by adapting the forming tools to compensate for the springback. Increasing the sheet tension is not applicable to every forming process and can lead to a split in the sheet. Therefore, tool compensation is often used solely or in combination with sheet tensioning to control the springback.

Finite element simulations are commonly used during the tool design stage to estimate the springback in the part. The simulation results are used to adapt the tool geometry to compensate for the springback. In this way, the costs associated with tool reworks can be reduced significantly. The accuracy of such simulations is highly dependent on the constitutive models employed in the simulations that can describe the material behavior during the forming process [59].

Figure 1.1 The experimental unloading/reloading curve and the prediction of the linear elastic assumption.

In recent years, with the emergence of advanced high strength steels (AHSS), springback has gained the attention of many researchers. In the case of AHSS, the springback is significantly larger than in conventional mild steels. The reason for this is twofold: 1. AHSS exhibits a larger strength to modulus ratio and 2. the use of AHSS allows for thinner sheets. Both factors result in a larger stress gradient in the sheet thickness after forming and ultimately giving rise to a larger springback. Therefore, an accurate prediction and compensation of the springback is important for the widespread utilization of AHSS.

In the past years, most of the research was focused on the development of novel plasticity models to provide an accurate stress prediction in the material. Yet, the material behavior upon unloading has been largely overlooked. It is generally accepted that the springback of the deformed material is solely driven by the recovery of elastic strain upon unloading. However, experimental evidence has shown that this is an invalid assumption. It has widely been observed that a plastically deformed material shows a nonlinear unloading/reloading behavior. In Figure 1.1 the experimental stress–strain curve of a DP800 undergoing an unloading/reloading cycle is shown. It can be seen that the total recovered strain is considerably larger than what is predicted by the linear elastic assumption commonly used in FE simulations.

The nonelastic portion of the total recovered strain accounts for up to 20%

of the total recovered strain when unloaded to zero stress. Considering that the springback is governed by the total recovered strain upon unloading of the deformed part, modeling the unloading behavior is essential for an accurate springback prediction.

1.2 Objective of this thesis

The aim of this research is to unravel the main mechanisms that are responsible for the observed nonlinear unloading/reloading behavior. This is carried out by performing a combination of theoretical, experimental and numerical studies on DP600 and DP800 from the family of AHSS. Based on the underlying physics, constitutive models are developed to improve the springback prediction capability of FE simulations. Finally, in this research a test rig is developed and built to evaluate the performance of the proposed models in predicting the springback angle of DP800 in a realistic forming process.

1.3 Outline

Chapter 2 gives an overview of the available literature on the topic at hand. In this chapter, the main mechanisms behind the nonlinear unloading behavior and the modeling approaches within the literature are reviewed.

In Chapter 3, the details of the experimental procedures used in this study and the obtained results are given. This chapter consists of microstructure characterization, uniaxial experiments and draw-bend experiments.

In Chapter 4, a constitutive model based on dislocation driven anelasticity is proposed. The proposed model is fitted to the experimental data to predict the stress-strain response of a DP800 steel subjected to unloading/reloading cycles. The results are compared with the prediction by the commonly used initial E-modulus and the E-modulus degradation models.

In Chapter 5, the inhomogeneous deformation at the microstructure is studied as an explanation for the observed nonlinear unloading behavior and Bauschinger effect. To this end, the stress and strain partitioning in a dual phase microstructure is analyzed using the crystal plasticity finite element modeling (CPFEM) approach. It is shown that, during unloading, some fractions of the material undergo compression and re-yield in compression. A computationally efficient model based on a numerical homogenization scheme

is presented. In this model, the stress inhomogeneity in the material is modeled by considering a distribution in yield stress of material fractions. The model performance is evaluated with respect to predicting the stress–strain response of a DP600 steel regarding the nonlinear unloading behavior and Bauschinger effect.

In Chapter 6, the performance of the models developed in Chapter 4 and Chapter 5 for improving the springback prediction are evaluated. The results are also compared with the classical E-modulus degradation model and the case where the E-modulus is taken as a constant. The draw-bend experiments described in Chapter 3 are used as a benchmark for evaluating the performance of each model in predicting the springback angle.

Finally, in Chapter 7, the conclusions drawn from this study are summarized and recommendations for future research are presented.

Nonlinear unloading behavior

2.1 Introduction

This chapter gives an overview of the main aspects of nonlinear unloading behavior with respect to the characterization methods, responsible mechanisms, modeling and its significance to the accuracy of springback prediction.

A survey on the studies performed on the characterization of the complex unloading/reloading behavior is presented in Section 2.3. Next, the theories that are proposed in literature on the mechanisms responsible for the observed behavior are summarized in Section 2.4. The numerical models developed to capture the unloading behavior are reviewed in Section 2.5. Finally, the relevance of modeling the unloading behavior for improvement of the springback prediction of the FE models are discussed in Section 2.6.

2.2 Terminology

As mentioned earlier, the unloading response of material cannot simply be described by a linear relation between the elastic strain and stress. Therefore, to describe the nonlinear hysteresis appearance of the unloading/reloading stress-strain curves, more than a name and value is needed. Unfortunately, there is no generally accepted and unified terminology within the field. Those researchers who consider this phenomenon as a pure decrease of the linear elastic response in the material often refer to it as the E-modulus degradation as a function of plastic strain.

Other researchers oppose this term and argue that the E-modulus is a fundamental material property and should be considered independent of plastic strain.

These two schools of thought have been a source of arguments amongst researchers on the mechanism, characterization techniques and numerical implementation of such a phenomenon.

Researchers who are aware of the nonlinear, hysteresis nature of the unloading/reloading stress-strain curves use different names and notations to describe the unloading response. As an example, the slope of the straight line connecting the stress point at the start of unloading to the point at zero stress is variously called “chord modulus”, “secant modulus”, “effective unloading modulus”, “springback modulus”, “average unloading modulus”, “apparent unloading modulus”. Therefore, consistent terminology is required to avoid confusion.

Figure 2.1 illustrates schematically the nonlinear unloading/reloading behavior in an exaggerated way. The names and notations designated in Figure 2.1 will be used throughout this thesis as well as when referring to external works. Elastic modulus refers to the material stiffness due purely to interatomic forces and is assumed to be constant. The instantaneous modulus, also known as the tangent modulus, is the slope of the stress-strain curve at a specific point. It is defined as the derivative of the uniaxial stress with respect to the uniaxial strain (i.e. $\partial\sigma_{11}/\partial\varepsilon_{11}$).

The reduction in the chord modulus is often called “E-modulus degradation” in the literature. Although it is an inaccurate denomination, for simplicity’s sake the term “E-modulus degradation” will be used in this work to refer to this concept.

During unloading, the stress-strain curve deviates from the linear elastic line showing an extra strain recovery on top of the elastic strain. This additional strain is called anelastic strain. The term anelasticity corresponds to different material behavior depending on the context. In this work the word anelasticity refers to the original definition by Zener [104]:

“anelasticity is the term chosen by the author to describe the behavior of metals in region of small strain, where, however, strain is not linear single-value function of stress alone, but yet where no permanent plastic deformation takes place”

The above definition is valid for describing the macroscopic behavior of the material regardless of the mechanism resulting in the nonlinear unloading be-

Figure 2.1 Schematic illustration of the hysteresis behavior during unloading-reloading.

havior. The permanent plastic deformation corresponds to the non-recovered strain when the material is macroscopically unloaded to zero stress. As a complement to Zener's definition, the anelastic behavior is a dissipative process as a result of which the unloading-reloading cycles appear as hysteresis loops.

2.3 Complex unloading/reloading behavior

The degradation of the elastic stiffness in metals that were subjected to plastic straining, has been observed for a long time and studied regardless of its importance in springback simulations. Initially, in 1932, Taylor and Quinney [84] reported the change in E-modulus during loading/unloading of annealed tubes of aluminum. However, no explanation on the origin of such observation was given. In 1941 Lawson [52] reported that the internal friction of polycrystalline specimens of oxygen-free copper increases and then decreases when the stress is continuously increased. He found a similar behavior with the elastic modulus in such way that the E-modulus decreased about 6% for a stress of 90 kg/cm² while a stress increase up to 160 kg/cm² resulted in a decrease of 4% in the E-modulus of the copper specimen. Ledbetter and Sun [54] used an ultrasonic method to study elastic constants of deformed polycrystalline copper. They observed that the deformed specimens usually show a lower

elastic stiffness in comparison with the annealed texture-free copper specimen. The authors argued that the “internal-structure changes” such as texture, non-pinned dislocation density and anisotropic dislocation array are responsible for this reduction.

Similar research has been conducted by various researchers on the influence of plastic deformation on “degradation of the elastic stiffness” of pure copper [18, 53, 58]. All authors agree that the dislocation micromechanisms are responsible for such phenomenon. Additionally, they all have reported a recovery of the elastic stiffness during annealing at moderate temperatures (100-200 °C). However, the measured values they have obtained are not always in agreement due to the different characterization methods and materials they have used.

Although this phenomenon was known by the material science community, it has been of minor importance in structural analysis and was often neglected by the mechanical engineering community. Yet, in the past two decades, due to the significant influence of the unloading behavior on springback, special attention was paid to automotive steel grades. Pérez et al. [75] investigated the inelastic response of two TRIP grade steels (TRIP700 and 800) at micro and macroscale. The authors used SEM and TEM electron microscopy methods to analyze the microstructure, dislocation arrangements and density. Additionally, they conducted tensile experiments to investigate the mechanical response of the material during loading and unloading. The analysis of the elastic response of the material showed a decrease in instantaneous tangent modulus during both loading and unloading where a higher decrease was observed in the case of TRIP700. They attributed this decrease to microplastic deformation during unloading. Based on microscopic analysis, they stated that the density of the mobile dislocations is higher in TRIP700 than TRIP800 which leads to larger microplastic strain and consequently a larger decrease in the instantaneous modulus.

Yang et al. [100] investigated the microscopic variation of the E-modulus by means of nano-indentation. The authors determined the E-modulus on a scanning line passing through a grain boundary by making use of load-displacement curves of the indentation. They found the elastic modulus on and around the grain boundary to be lower than the values obtained from the core of the grain. They related this observation to the reversible motion of the dislocations piled up at the grain boundaries. Additionally, the authors reported the typical degradation of the effective unloading modulus as a function of plastic strain. Yang et al. [100] used a polynomial function to describe this variation in the effective E-modulus.

Benito et al. [9] investigated the variations in the E-modulus of pure iron with plastic deformation. They reported that the mean value of the E-modulus drops to 196 GPa at $\varepsilon = 0.06$ from its original elastic stiffness (210 GPa). After that, further deformation resulted in a slight increase and stabilization of the modulus to 198 GPa. In order to investigate the root cause of such behavior, the authors performed a comprehensive analysis on the residual stresses, the texture and the dislocation structure of the specimens. They claimed that the residual stresses and texture are irrelevant to the variations in the E-modulus. The authors also found a correlation between the dislocation density and the variation in the E-modulus.

However, all the authors mentioned above have formulated the phenomenon as a variation (or degradation) in E-modulus and have not realized the nonlinear recovery behavior. The nonlinear unloading behavior has been widely realized by researchers who were performing uniaxial tensile tests with loading–unloading–reloading cycles after increments of plastic deformation [15, 20, 65, 68]. Considering that the deviation from linear behavior is very small, this was only possible thanks to the improvements made on strain measurement devices in the last decades.

The elastic behavior of materials is a result of interatomic forces and often considered linear and modeled by Hooke’s law. However, the elastic theory includes higher order terms which can lead to a slight deviation from purely linear elastic behavior [98]. However, it was discussed by Sun and Wagoner [81] that the second order elastic effect is too small to explain the nonlinear recovery that is observed in the uniaxial tensile experiments.

Cleveland and Ghosh [15] made a comprehensive study on nonlinear unloading and reloading of 6022-T4 aluminum and GP50XK60 high strength steel. The authors developed a uniaxial model to describe the compliance of the material during unloading. In their model, the total compliance is composed of elastic, microplastic and plastic compliances. The stress–strain behavior of the material can be described by these three compliances which are determined experimentally.

In a similar study by Luo and Ghosh [65], the elastic and inelastic behavior of DQSK steel sheets was investigated. They found the average unloading modulus to be decreasing with plastic pre-strain while the measurements of the Poisson’s ratio showed an increase in average with plastic pre-straining. To measure the initial elasticity modulus at zero stress, the authors made use of a dynamic resonance method. Based on a physical constitutive model previ-

ously proposed by Ghosh [31], they developed a set of equations to describe the inelastic and elastic strains as a function of plastic pre-strain. Last but not least, they proposed a decaying exponential relation between the average modulus and the plastic pre-strain.

Similar behavior has been reported by other researchers who were intending to model the reduction of the effective E-modulus as a function of plastic strain to use in springback simulations. These models are further discussed in Section 2.5.1.

2.4 Responsible mechanisms

As summarized by Chen et al. [13], various mechanisms for the nonlinear unloading behavior following plastic deformation have been discussed over the years: inhomogeneous deformation at the microscale [1, 32, 42, 86, 94], damage evolution [34, 35, 91, 92], twinning/detwinning and kink bands in HCP alloys [12, 37, 105, 106] and dislocation driven anelasticity [2, 3, 9, 15, 20, 28, 31, 33, 65, 69, 85, 87, 88].

Twinning/detwinning and kink bands are specific to the HCP and are rarely observed in metals with BCC and FCC crystal structures. Yet the phenomenon is observed in FCC and BCC metals as well. Therefore, this cannot be held as the main mechanism in case of steels with BCC and FCC crystal structure.

The theory of chord modulus degradation due to damage proposes that upon plastic deformation, voids and microcracks initiate and grow in the material which results in a reduction of the load bearing capacity and a decrease of the stiffness. As will be shown later on in Chapter 3, the chord modulus increases over time at room temperature and is restored to its initial value after a short baking at a relatively low temperature. These experimental observations are not compatible with the damage theory as it is not expected that the material would heal in such conditions. Similar reports on the sensitivity of the chord modulus to the waiting time [20, 68, 77] and annealing [99] are reported in the literature.

Amongst all the theories, “inhomogeneous deformation at the microscale” and “the dislocation driven anelasticity” are the two most plausible that can result in the observed nonlinear unloading behavior in AHSS.

The former theory states that, due to the inhomogeneous nature of the polycrystalline metals, the stress and strain are partitioned among the grains with

different strength. As a result, the weaker grains already go into compression during unloading and likely yield in reverse. The plastic deformation of small fractions of the material during unloading results in the observed nonlinearity. The stress and strain partitioning in the case of dual phase steels is more severe than that of the single phase steels due to the large difference in the strength of the ferrite and martensite phases.

For dislocation driven anelasticity, two main mechanisms are discussed in the literature to explain the anelastic deformation: 1. short-range dislocation motion [15, 20, 31, 33, 65] and 2. bowing of the dislocation line between pinning points [2, 3, 9, 28, 69, 85, 87, 88]. The short-range dislocation motion is explained by the mechanism in which the dislocations pile up against different kinds of obstacles such as grain boundaries or solute atoms during loading. As the stress is further increased, new dislocations are generated and the density of piled up dislocations increases. These piled up dislocations are repulsive to each other and only the applied stress keeps them together. Upon unloading, when the stress acting on the dislocations drops, the piled up dislocations move backwards to their equilibrium distance that is associated with an additional strain recovery. When the material is loaded again, the density of mobile dislocations increases and new dislocation pile-ups are formed.

The other mechanism known for anelastic strain is bowing of the dislocation segment between two anchoring points. According to this description, a dislocation segment pinned at two nodal points cannot glide but only bow out under influence of the applied stress. The dislocation line tension is in equilibrium with the applied stress acting on the dislocation segment. While the applied stress forces the dislocation segment to bow out, the dislocation line tension pulls the dislocation to its initial state. By decreasing the applied stress, the dislocation segments unbow and result in an additional strain recovery.

2.5 Constitutive modeling of unloading/reloading behavior

Generally the models to describe the unloading/reloading behavior of the material are divided into two groups. The first group are the models which treat the unloading behavior linearly by the chord modulus. These models usually are known as the E-modulus degradation models. The second group are the models that capture the experimentally observed nonlinear unloading/reloading behavior. In the following sections both modeling approaches are reviewed.

2.5.1 E-modulus degradation models

Practically, the simplest way to include the E-modulus degradation in a simulation is to make the modulus of elasticity a function of equivalent plastic strain and update the modulus of elasticity in the simulation at every iteration. This approach has been widely adopted by authors who are aware of nonlinear unloading behavior as well as those who consider the E-modulus as a single value which decreases as a function of plastic strain. The former group of authors consider the effective E-modulus (or the chord modulus) as the input of the simulations determined from experimental data. Morestin and Boivin [68] were among the first researchers to emphasize the importance of taking into account the variation in the elastic stiffness for springback analysis. They used a piecewise linear function to model the reduction of the effective elastic modulus as a function of plastic strain. The authors implemented the model in a semi-analytical software (PLIAGE) and performed a U-draw simulation [68]. They reported a significant improvement in springback prediction by considering the modulus variations in the model.

Some other authors have adopted the same approach as Morestin and Boivin and have varied the effective E-modulus as a function of plastic strain using power-law [60], polynomial [100, 102], piecewise linear [64, 103], linear [24] and exponential [101] functions. The exponential model proposed by Yoshida [101] has been widely adopted for springback simulations.

2.5.2 Nonlinear unloading models

The deficiency of the E-modulus degradation model in accurate springback prediction has been realized and pointed out by various authors. In that respect, attempts have been made in order to capture and implement the nonlinear elastic unloading into the simulations. Eggertsen et al. [20] and Sun and Wagoner [81] have taken similar approaches based on the two-yield-surface plasticity theory and proposed two-surface constitutive models in which the inner surface defines the transition between the linear and nonlinear elasticity and the outer surface gives the yield criteria (see Figure 2.2). In the model proposed by Eggertsen et al. [20], the inner surface evolves kinematically while having the same shape as the outer one. The yield criterion and the hardening law for the inner surface are given by

$$F = \bar{\sigma} (\sigma - \gamma) - \sigma_s = 0 \quad (2.1)$$

Figure 2.2 Illustration of the inner and the outer surface of the models by Eggertsen et al. [20] (left) and Sun and Wagoner [81] (right).

and

$$\dot{\gamma} = \frac{H'_r(p) \dot{p}}{\bar{\sigma}_r} (\sigma - \gamma) \quad (2.2)$$

where γ is a back-stress tensor which describes the center of the inner surface, σ_s is the size of the inner surface, i.e. the size of the linear elastic region, and H'_r is the slope of the hardening curve. When the stress states are out of the inner surface ($F > 0$), a transition from the linear elastic behavior to nonlinear elastic behavior takes place. The nonlinear material behavior inside the outer surface (yield surface) is given by

$$H'_r(p) = \sigma_s + ap^{1/n} \quad (2.3)$$

where a and p are parameters which describe the hardening curve and are determined from experimental data. According to the authors, this model is independent from the hardening law and the yield criteria.

Sun and Wagoner [81] have also proposed a two-surface constitutive model called QPE (Quasi-Plastic-Elastic) model to describe the nonlinear elastic behavior of the material. The QPE model describes three different deformation modes: elastic, QPE (nonlinear elastic) and plastic. In the same fashion as the model suggested by Eggertsen et al. [20], the transition between the linear elasticity and QPE mode is realized by the inner surface f_1 and the transition between the QPE and the plastic mode is given by an outer surface f_2 which evolves according to the nonlinear hardening rule of Chaboche given as

$$f_1 = \bar{\sigma} (\sigma - \alpha) - R_1 = 0 \quad (2.4)$$

Figure 2.3 The experimental unloading–reloading cycle and the predictions of the models proposed by Eggertsen et al. [20] (left) and the QPE model [81] (right).

and

$$f_2 = \bar{\sigma} (\sigma^* - \alpha^*) - R_2 = 0 \quad (2.5)$$

where R_1 and R_2 are the sizes of the inner and the outer surfaces respectively, and α and α^* are the centers of the inner and the outer surfaces respectively. In the transition between the QPE and plastic mode, f_1 follows an evolution rule in such way that σ and σ^* would coincide and share the same normal at the transition point.

The experimental data of an unloading–reloading cycle and the predictions of the model proposed by Eggertsen et al. and the QPE model are shown in Figure 2.3.

Both phenomenological models by Eggertsen et al. and Sun and Wagoner (QPE) defined a linear elastic region within the inner yield surface. The basis on defining an elastic domain in which the material behaves purely elastically is not clear. Moreover, as discussed in [14] there is no actual linear elastic region during unloading. This results in an ambiguity in identification of the model parameters from the experiments.

2.6 Springback prediction

As mentioned earlier, the accuracy of springback prediction is highly dependent on a correct estimation of the stresses in the structure prior to unloading. This

Figure 2.4 Schematic representation of stress–strain response of the material subjected to tension–compression.

is directly linked to the yield criteria and the plastic constitutive model that are used in a simulation.

During a typical deep-drawing process, sheet material undergoes loading, unloading and reverse loading. A material that is subjected to a loading–unloading–reverse loading cycle typically exhibits nonlinear unloading behavior, the Bauschinger effect, transient behavior and permanent softening (see Figure 2.4). Therefore, for an accurate prediction of stress and strain in the material it is vital to incorporate constitutive relations that can capture the above-mentioned material behavior in the simulation.

The behavior of the material subjected to compression after tension is usually characterized by three features in the stress–strain response: 1. the Bauschinger effect, characterized by early re-yielding in compression, 2. the transient behavior that is recognized by the smooth elastic to plastic transition and 3. the permanent softening which is realized as the stress offset between the tension–compression curve and the monotonic curve. Various authors have emphasized the importance of taking into account the above-mentioned effects for an accurate springback simulation. In that respect, hardening laws have been developed to encompass such features.

The simplest hardening rule is the isotropic hardening model, in which the

Figure 2.5 The springback profile of the U-draw test predicted using different hardening models [57].

yield surface expands isotropically while the center of the yield surface does not shift. As can be seen in Figure 2.4, the isotropic hardening law cannot describe the Bauschinger effect, transient behavior nor the permanent softening in the material. Therefore, hardening models incorporating the kinematic hardening rule and mixed hardening rule were devised to describe such behaviors.

According to Wagoner et al. [93], there are three main hardening models that are widely used for accurate springback prediction: 1. Armstrong–Frederick type hardening models, 2. multi-surface type hardening models and 3. a novel hardening model without simple kinematic hardening. The latter, known as HAH model, is proposed by Barlat et al. [7].

Lee et al. [57] compared the performance of the hardening behaviors of the two-surface model, the isotropic hardening model, the kinematic hardening model and the modified Chaboche model in the springback prediction of the U-draw test (NUMISHEET 1993 benchmark).

The modified Chaboche model (an Armstrong–Frederick type model with multiple back-stress terms) presented in [56], accounts for the Bauschinger and transient behavior but not the permanent softening behavior. According to Figure 2.5, the springback simulation using the two-surface model, which considers the Bauschinger effect and the permanent softening behavior, is in best agreement with the experimental result. However, the modified Chaboche model and the isotropic models overpredict the springback angle [57].

Eggertsen and Mattiasson [19] have investigated the accuracy of the isotropic, mixed, Armstrong–Frederick, Geng–Wagoner, and two-surface Yoshida–Uemori models in springback prediction of the U-draw test. They evaluated each model with and without considering the E-modulus degradation model of Yoshida et al. [101]. The results, as demonstrated in Figure 2.6, show that the Geng–Wagoner model as well as the Yoshida–Uemori model give the most accurate results when the E-modulus degradation is considered. Both Yoshida–Uemori and Geng–Wagoner models are capable of describing the transient Bauschinger effect and permanent softening; the Yoshida–Uemori model describes the work-hardening stagnation as well. From their results it can be conceived that the influence of taking into account the E-modulus degradation is bigger than the difference between the hardening models.

In a similar study, Ghaei et al. [30] compared three different models to analyze the accuracy of springback prediction for a U-shape channel draw process (presented as NUMISHEET 2005 benchmark 3) of a DP600 steel. The models they considered in their work were: isotropic, combined (isotropic plus nonlinear kinematic) and Yoshida–Uemori two surface hardening models. The evaluation was performed using the initial E-modulus as well as in combination with the E-modulus degradation according to the empirical model of Yoshida et al. [101]. According to their simulation results, the Yoshida–Uemori hardening model in combination with the E-modulus degradation model gives a better springback prediction. However, the improvement gained by using the complex model of Yoshida–Uemori instead of the combined hardening model is very small in comparison with the enhancement earned from using the E-modulus degradation model of Yoshida et al. [101].

In the works mentioned above, the authors have modeled unloading behavior linearly and the focus was on evaluating the effect of the different hardening models and the E-modulus degradation model on the springback prediction performance. Recently, a few studies have been performed on incorporating nonlinear unloading models in the springback simulations.

Lee et al. [55] extended and combined the QPE model with the homogeneous anisotropic hardening model (HAH) to improve the prediction of springback in AHSS. The HAH model, proposed by Barlat et al. [7], has the advantage of capturing the Bauschinger effect, transient behavior, work-hardening stagnation and permanent softening [55]. In order to check the effect of considering the nonlinear unloading behavior by using the QPE model, the authors have compared the springback simulation results obtained by the conventional HAH-Chord model (i.e. E-modulus degradation) with that of the HAH-QPE

Figure 2.6 Predicted tip deflection in the U-draw test after springback without E-modulus degradation (Elastic modulus) and with E-modulus degradation (Unloading modulus) for different hardening models and comparison with the experimental result [19].

model (see Figure 2.7). The authors have also included the simulation result of the isotropic hardening with the Chord model (IH-Chord) in Figure 2.7 for comparison.

Ghaei et al. [29] incorporated a modified version of the QPE model with a mixed hardening (isotropic plus nonlinear kinematic) model to simulate the springback of the NUMISHEET 2005 U-draw process made of TRIP780 steel. They compared the springback prediction capability of their model with the constant E-modulus and the E-modulus degradation model. The result of their analysis is shown in Figure 2.8.

Both Lee et al. and Ghaei et al. have reported a considerable improvement in the springback prediction by incorporating the nonlinear unloading behavior in comparison with the E-modulus degradation model.

Figure 2.7 The springback profile of the U-draw test predicted using the HAH model in combination with the QPE model and the E-modulus degradation model (HAH-Chord) in comparison with the experimental result [55].

Figure 2.8 The predicted springback profile of the U-draw test predicted using a combined hardening model with constant E-modulus (CEM), E-modulus degradation model (Chord) and the modified QPE model (Current model) in comparison with the experimental result [29].

2.7 Conclusions

As discussed earlier, there is no general agreement on the nature and mechanism of the nonlinear unloading/reloading behavior and the reduction of the chord modulus. As a matter of fact, the hysteresis behavior of the metals subjected to unloading–reloading cycles has not been realized by all the researchers. Therefore, the phenomenon has been widely attributed as the E-modulus degradation in the literature.

Nevertheless, all the authors agree on the importance of considering nonlinear unloading behavior or E-modulus degradation for an accurate springback prediction. Up to now, most researchers have tried to take into account the effective E-modulus degradation by simply making the E-modulus a function of equivalent plastic strain in the springback simulations. However, there have been few attempts to consider the real nonlinear unloading behavior in the simulations.

Eggertsen et al. [20] and Sun and Wagoner [81] have taken similar approaches based on the two-yield-surface plasticity theory [57] and proposed two-surface constitutive models in which the inner surface defines the transition between the linear and nonlinear elasticity and the outer surface is the yield criterion.

Some investigations have shown that considering the nonlinear unloading/reloading behavior in simulations will improve the springback prediction accuracy, as compared with the classical E-modulus degradation model. However, these models are built based on computational convenience and lack a physical basis.

Therefore, a model based on the underlying mechanism is to be developed that can give an accurate prediction of the nonlinear unloading/reloading behavior with a minimum number of fitting parameters. The model parameters in such a model are physically meaningful and can be obtained from the mechanical tests on the material.

Among the mechanisms responsible for the nonlinear unloading/reloading behavior that are discussed in literature, the “dislocation driven anelasticity” and “inhomogeneous deformation at the microscale” appear to be the most conceivable mechanisms in case of advanced high strength steels. These mechanisms will be used as the basis of models for predicting the nonlinear unloading/reloading behavior.

Experimental methods

In this chapter the experimental setups and procedures used in this study and the obtained results are presented. The uniaxial experiments were performed to characterize the mechanical behavior of the material subjected to a monotonic or cyclic load. The data obtained from the uniaxial experiments were used to calibrate the models used for springback simulations. On top of that, an extensive experimental work was dedicated to obtain a better understanding of the mechanism behind the nonlinear unloading/reloading behavior. The experiments were designed to tackle the mechanisms that are possibly responsible for the observed nonlinear unloading/reloading behavior (see Section 2.4). More specifically, the compatibility of the obtained results with theories of *dislocation driven anelasticity*, *inhomogeneous deformation at the microscale* and *damage* are discussed.

The draw-bend experiments were performed using an in-house draw-bend machine to investigate the springback behavior of the DP800 steel grade. The draw-bend experiment serves as a benchmark problem to validate the newly developed models and to compare them with the other models. In this chapter, the details on the draw-bend machine are given and the obtained results are shown.

Finally, the EBSD experiment was performed to characterize the microstructure of the DP600 steel in terms of grain size distribution and grain orientations. The data obtained from the EBSD experiment was used to reconstruct the microstructure of the DP600 steel grade for CPFEM simulations in Chapter 5.

Table 3.1 Chemical composition of DP600 and DP800.

Material	Chemical composition in wt 10 ⁻³ %								
	C	Mn	Si	P	S	Al	Ti	Nb	V
DP600	94	1874	50	13	1	300	3	-	4
DP800	149	2054	408	12	0	614	8	20	4

3.1 Materials

This study is focused on two dual phase (DP) steel grades: DP600 and DP800, both from the family of advanced high strength steels (AHSS). The microstructure of DP steels consists mainly of the martensite and the ferrite phases. The hard martensitic phase is embedded in the soft ferritic matrix. The volume content of the martensite in DP steels usually ranges between 5 and 30% [47]. DP steels usually exhibit a high ultimate tensile strength, due to the martensitic phase, and a relatively low yield stress enabled by the ferritic phase [82].

The DP600 and DP800 specimens were cut from sheets with thickness of 1.2 and 1.0 mm respectively. The main alloying elements for both grades are summarized in Table 3.1 according to the manufacturer.

The electron backscatter diffraction technique was used in order to characterize the microstructure of the material. The EBSD maps were obtained using a JEOL JSM-7200F scanning electron microscope (SEM) equipped with an EBSD detector.

The experiment was performed on the DP600 grade in as-received condition. The measurements were performed on a rectangular area of $60 \times 45 \mu\text{m}^2$ with a step size of 70 nm in the rolling (RD) and normal (ND) plane. The MATLAB toolbox MTEX [6] was used to process the EBSD data. An angle misorientation threshold of 10° was chosen to detect the grain boundaries. From the EBSD data an average grain size of $5.8 \mu\text{m}$ was determined. The grain map of DP600 constructed from the EBSD data is shown in Figure 3.1.

Figure 3.1 The grain map of DP600 constructed from the EBSD data.

3.2 Uniaxial experiments

All the uniaxial tests presented in this work were conducted at room temperature using a Zwick/Roell 100 kN electro-mechanical testing machine. The uniaxial experiments are classified into uniaxial tensile and uniaxial creep experiments.

3.2.1 Uniaxial tensile tests

The uniaxial tensile experiments were performed in rolling (RD), diagonal (DD) and transverse (TD) directions to characterize the mechanical response of DP600 and DP800 steels. The dog-bone specimens were cut in accordance with the ASTM E-8 standard. The geometry of the tensile test specimen is depicted in Figure 3.2.

The strain was measured using a custom-made double-sided clip-on extensometer over a gauge length of 25 mm. The double-sided extensometer measures the strain on both sides of the specimen and outputs the average. At low

Figure 3.2 The geometry of the tensile test specimen (dimensions in mm).

Figure 3.3 The true stress–strain curves of DP600 and DP800 in different directions.

strain levels, misalignment and bending of the sample can have a significant contribution on scatter and uncertainty in the strain measurement which is minimized by averaging the strain measured on both sides of the sample.

The experiments were carried out at a constant crosshead speed of 5 mm/min resulting in a strain rate of 0.0005 s^{-1} . The obtained true stress–strain curves are plotted in Figure 3.3.

The stress–strain curves obtained from the experiments in the rolling direction were used to obtain the isotropic hardening parameters of the materials. The mechanical properties of the two steel grades are summarized in Table 3.2. The yield stress was defined using the 0.2% offset criterion. The elongation values correspond to the true strain at the limit of the uniform elongation and the UTS values are obtained from the engineering stress–strain curves.

3.2.2 Uniaxial creep

Creep usually denotes a slow viscous flow of a solid under macroscopically non-zero stress, via atomic diffusion (through lattice or along grain boundary) and dislocation motion (glide or climb). The mechanism of creep depends on composition, microstructure features, temperature and stress level. Dislocation creep is believed to be the dominant creep mechanism at room temperature [72].

Table 3.2 Mechanical properties of DP600 and DP800.

Material	Test direction	σ_y (MPa)	UTS (MPa)	Elongation (%)
DP600	RD	412	652	25
	DD	420	665	23
	TD	419	666	16
DP800	RD	461	815	21
	DD	512	841	12
	TD	455	802	18

Figure 3.4 Room temperature creep tests for DP800-RD at different stresses.

Creep experiments were performed on DP800 steel in the rolling direction at room temperature. The creep strain was measured at different loads. For that, the specimens were loaded to 5700, 7000, 8200, 9500 and 10000 N corresponding to 450, 560, 660, 790 and 860 MPa for a period of 10000 s. The creep strain measurement was started at the moment that the creep load was reached. During the creep experiment the load was controlled and maintained at the prescribed level. The creep strains as a function of time for different loads are plotted in Figure 3.4.

3.3 Uniaxial loading–unloading–reloading

For the cyclic loading–unloading–reloading experiments (LUR), the controller was programmed to load the specimen to a certain force and then unload it to zero force. Subsequently, the material was reloaded to a higher load level and unloaded again. The procedure was repeated with an increase in the pre-load at every cycle. The cyclic tests were carried out at a constant crosshead speed of 2, 5 and 10 mm/min which results in strain rates of 0.0002, 0.0005 and 0.001 s⁻¹ respectively. For every strain rate the experiment was performed on three specimens. As these are the main type of tests where nonlinear unloading manifests itself, different aspects will be separately analyzed and discussed below.

3.3.1 Nonlinear unloading/reloading behavior

A typical repeated LUR stress–strain response is shown in Figure 3.5 for DP600 and DP800 grades tested in the rolling direction. Each LUR cycle is repeated after some plastic deformation. The LUR presents two main features: 1. the unloading–reloading cycle forms a hysteresis loop which is symmetric along the chord modulus and 2. the width of the hysteresis loops increases with plastic deformation.

A magnified view of a LUR cycle is shown in Figure 3.6 which corresponds to the 7th cycle of the LUR experiment belonging to DP800 in the rolling direction. The graphical representation of the chord modulus as well as the decomposition of the total recoverable strain into the elastic and anelastic parts are shown in Figure 3.6.

To describe the nonlinear anelastic behavior firstly a relation for the total recoverable anelastic strain should be established. Considering that the total recoverable strain (ε_t^{rv}) is partially elastic (ε_t^e) and partially anelastic (ε_t^{an}), the contribution of the anelastic strain can be determined by subtracting the elastic strain from the entire recovered strain according to

$$\varepsilon_t^{an} = \varepsilon_t^{rv} - \varepsilon_t^e = \varepsilon_t^{rv} - \frac{\sigma_f}{E} \quad (3.1)$$

where E is the elastic modulus of the material (i.e. 204 GPa) and σ_f is the flow stress of the deformed material. Subscript t refers to the total strain recovered when the material is unloaded to zero stress.

Figure 3.5 LUR experiment results for DP600-RD (top) and DP800-RD (bottom) tested in the rolling direction.

Figure 3.6 Magnified view of the 7th unloading–reloading cycle of DP800-RD.

The average recovered anelastic strain is plotted versus the pre-strain in Figure 3.7 for DP600 and DP800 steels. The error bars represent the standard deviation of the three measurements. Based on the results, no strong conclusion can be drawn on the strain rate dependency of the anelastic strain in the range it was evaluated.

The reduction of the chord modulus as a function of the pre-strain for different strain rates is plotted in Figure 3.8 for DP600 and DP800 steels. For the first unloading cycle at 0.5% pre-strain, the chord modulus is around 185 GPa which is significantly lower than the handbook value of steel E-modulus (i.e. 210 GPa). The chord modulus continues to decrease to approximately 140 GPa and 155 GPa for DP600 and DP800 after 8% plastic strain.

The effect of the sheet anisotropy on the anelastic behavior of the DP800 steel was investigated by repeating the LUR experiment on the specimens cut in diagonal (DD) and transverse (TD) directions. The averaged values of the three measurements per direction are plotted in Figure 3.9. No strong direction dependency between the specimens in the rolling and diagonal directions is found, while the magnitude of the recovered anelastic strain in the transverse direction is significantly lower.

In the LUR experiments shown so far, the material behavior was investigated when it was fully unloaded to zero force and reloaded to the flow stress of the material (a full unloading–reloading cycle). The stress–strain response of

Figure 3.7 Variation of the total recovered anelastic strain with pre-strain in DP600-RD (left) and DP800-RD (right).

Figure 3.8 Reduction of the chord modulus with pre-strain in DP600-RD (left) and DP800-RD (right).

Figure 3.9 Variation of the total recovered anelastic strain with pre-strain for different test directions in DP800.

DP800 when it is subjected to partial loading–unloading cycles is shown in Figure 3.10. In this experiment, the material was first loaded to 800 MPa and then unloaded to zero stress. Next, it was partially loaded and unloaded repeatedly. Partial loading–unloading cycles generate smaller inner loops within the fully unloaded–reloaded hysteresis loops.

The experimental results shown in this section were focused on quantifying the magnitude of the anelastic strain at different pre-strains and the test direction (i.e. RD, DD and TD). However, they do not provide any hint on the mechanism governing the nonlinear unloading/reloading behavior.

Experimental conditions that affect the dislocation dynamics or residual stresses (by a stress relaxation mechanism) in the material are expected to influence the nonlinear unloading/reloading behavior. Both dislocation motion and stress relaxation are thermally activated processes and are therefore sensitive to time and temperature.

In order to gain a better insight into the physics behind the observed nonlinearity, experiments that are involving waiting time and baking treatment are presented in the following sections.

Figure 3.10 The stress–strain response of DP800-RD subjected to partial loading–unloading cycles.

3.3.2 Effect of waiting time

To investigate the influence of waiting time on the anelastic behavior of DP800, repeated loading–unloading measurements with a waiting time in between each cycle were performed. The experiments were performed in the rolling direction of the sheet. For each experiment, the specimen was loaded up to a certain pre-load (i.e. 8, 9 and 9.5 kN) and was unloaded to zero load followed by a waiting stage before the next loading–unloading cycle. In each experiment the material was loaded up to a constant maximum load and the anelastic strain recovered upon the unloading stage was evaluated. It is worth mentioning that the waiting time between each loading–unloading cycle was increased at each step. The loading–unloading cycles were carried out at a constant crosshead speed of 5 mm/min. The schematic illustration of the experimental procedure and resulting strain–stress curve are shown in Figure 3.11 and 3.12 respectively.

The effect of waiting time on the magnitude of the recovered anelastic strain can be explained by both *dislocation driven anelasticity* and the theory of *inhomogeneous deformation at the microscale*.

During the waiting time the interstitial atoms (e.g. carbon and nitrogen) diffuse to the dislocations and confine them [16]. As a result, the dislocations that were once free to move and contribute to the nonlinear unloading/reloading behavior are locked by the interstitial atoms and therefore a smaller anelastic

Figure 3.11 The schematic representation of the experimental procedure in Section 3.3.2.

Figure 3.12 Variation of the total recovered anelastic strain as a function of waiting time in DP800-RD.

strain is recovered.

On the other hand, in support of the theory of inhomogeneous deformation at microscale, it can be argued that the reduction in the anelastic strain recovery is due to local stress relaxation in microstructure during the waiting time.

However, these results are in direct contradiction to the *damage* theory as the material is not expected to heal during the waiting time.

3.3.3 Effect of baking

The effect of baking on the nonlinear behavior was investigated by baking the DP800 test specimens at 220 °C for 20 min. The baking treatment was performed on both as-received and pre-deformed (14%) specimens. The stress–strain plots of the baked specimen before straining and the baked specimen after 15% pre-straining are plotted in Figure 3.13.

Comparing the tensile experiment results of the non-baked and the baked specimens it is clear that the upper yield point and the yield point elongation phenomena are restored in the baked specimen. In these stress–strain curves, a sharp yield-point and subsequently a drop in the stress followed by a plateau are observed. The occurrence of the upper yield point and the yield point elongation are not the focus of this research. Nevertheless, its link to the material loading behavior up to the upper yield point is of a particular interest. The occurrence of such phenomena is associated with the lack of mobile dislocations in the material [16]. Usually, in a pre-deformed material enough mobile dislocations exist to accommodate the prescribed deformation. This facilitates the gradual plastic transition observed in the monotonic loading of the as-received material. Remarkably, the stress–strain response of the baked material up to the upper yield point is almost linear. The averaged E-modulus (from three experiments) of the as-received baked specimens were found to be 204 ± 3 GPa and 202 ± 6 GPa for the pre-strained baked specimens. This shows no significant reduction in the E-modulus due to deformation.

From this observation, it can be concluded that the reduction of the average E-modulus is not caused by *damage*. The explanation that was given for the effect of waiting time in the previous section also holds to justify both *dislocation driven anelasticity* and the theory of *inhomogeneous deformation at the microscale*. The difference is that at higher temperatures the kinetics of interstitials diffusion and stress relaxation are much faster than at room temperature.

Figure 3.13 The stress–strain curve of DP800-RD after baking and baking after deformation.

3.3.4 Time-dependent behavior

The elastic strain, defined as the deformation of the atomic lattice, is by its nature instantaneous. Assuming that the nonlinear unloading behavior of the material is driven by a thermally activated process (e.g. dislocation based), it is expected to be time-dependent. The time dependency of the unloading behavior can appear as a phenomenon called elastic after-effect where the strain continues to recover over time after the material is unloaded. This latent strain recovery would not be noticed in unloading–reloading cycles at moderate and high strain rates. In order to investigate such behavior, the material was first loaded up to a certain force and was then unloaded almost instantaneously. In a fully unloaded state the strain measurement was continued for three hours. The experiments were performed in the rolling direction of DP800 specimens. The time-dependent strain recovery as a function of time is plotted in Figure 3.14.

The time-dependent anelastic strain shows a creep-like behavior. The rate of the strain is high at the beginning and drops quickly in the first 100 seconds and then slowly decreases to zero. The magnitude of the time-dependent strain is larger for the specimen that was loaded to a higher stress. Unlike creep, the time-dependent anelastic strain is active when the material is macroscopically unloaded.

3.3.5 Ratcheting behavior

Ratcheting or cyclic creep is one other interesting phenomenon that occurs when the material is subjected to an unbalanced cyclic load. Figure 3.15

Figure 3.14 The time-dependent anelastic strain recovery as a function of time in DP800-RD.

shows the stress–strain response of DP800 steel in the rolling direction when it is repeatedly unloaded and reloaded in a force controlled fashion. As it is shown in the magnified view in Figure 3.15 (right) the hysteresis loops never close and the material ratchets in the direction of the mean stress. The ratcheting strain increment is measured as the accumulated strain at the end of two consecutive cycles. The ratcheting strain increment and the total anelastic strain recovered in every cycle is plotted in Figure 3.16. The increment of the ratcheting strain is larger in initial cycles and saturates to a constant value. During the ratcheting cycles the total anelastic strain recovered in every cycle is also altered. It is interesting to note that the recovered anelastic strain decreases at the first few cycles and then starts to increase.

The ratcheting behavior shows that during the unloading and reloading cycles, some permanent plastic deformation occurs below the flow stress of the material. This observation supports the theory of inhomogeneous deformation at the microscale that claims the nonlinear unloading/reloading behavior to be a result of a microscopic plastic deformation during unloading–reloading cycles (see Section 2.4).

Figure 3.15 Ratcheting behavior of DP800-RD.

Figure 3.16 The ratcheting strain increment and the anelastic strain versus the cycle number in DP800-RD.

3.4 Uniaxial tension–compression

The uniaxial tension–compression experiments were conducted using an in-house anti-buckling fixture to prevent the specimen from failing due to buckling under compressive load. In this setup, the test specimen is supported by two sets of plates perpendicular to the specimen’s surface. Each set consists of five plates with a thickness of 3 mm. At each end, the plates are positioned within the slit of rigid blocks and fixed by two pins. The two sets of the plate are brought and pressed together by four screws that are passing through the rigid blocks and the pins. A spring is inserted at the end of each pin and is compressed against the blocks by bolts screwed into the pins. The assembly of the anti-buckling fixture is shown schematically in Figure 3.17.

The main advantage of this design is its ability to support the entire specimen gauge length during a test. This is due to the fact that the supports are able to change their length together with elongation and shortening of the specimen during the tension and compression stages. Using springs with bolts for clamping the fixture to the specimen ensures that a constant pressure is applied on the specimen throughout the whole experiment. The clamping force can be adjusted compressing the springs using bolts. For the tension compression experiments, the DP600 and DP800 specimens were cut in the rolling direction of the sheets according to the geometry shown in Figure 3.18.

The specimens were stretched to various levels of pre-strain and then compressed back to zero strain. The experiments were conducted at a constant cross head speed of 2 mm/min which resulted in a strain rate of 0.0004 s^{-1} in the measurement section. Due to the constraints imposed by the specimen dimensions and the anti-buckling fixture, the strain was measured on the thin side of the specimen using a Messphysik laser speckle extensometer over a gauge length of 20 mm.

The resulting stress–strain response of the DP600 and DP800 grades is shown in Figure 3.19. The behavior of the material subjected to compression after tension is usually characterized by three features in the stress–strain response: 1. the Bauschinger effect, characterized by early re-yielding in compression, 2. the transient behavior that is recognized by the smooth elastic to plastic transition and 3. the permanent softening which is realized as the stress offset between the tension–compression curve and the monotonic curve.

For a better illustration of the above-mentioned observations, the tension–

Figure 3.17 The assembly of the anti-buckling fixture.

Figure 3.18 The geometry of the tension/compression test specimen (dimensions in mm).

compression response of DP800 is plotted alongside with the monotonic tension curve in Figure 3.20. As can be seen in Figure 3.20 there is no clear yield point in compression. However, the yield point in compression is conventionally chosen using the 0.2% offset criterion.

Figure 3.19 The tension-compression response of DP600-RD (top) and DP800-RD (bottom).

Figure 3.20 Monotonic and tension–compression response of DP800 illustrating early re-yielding, transient behavior and permanent softening.

3.5 Draw-bend experiments

The deep-drawing process is widely used for producing complex parts from sheet metal. In this process the parts are formed by drawing the sheet metal over a radius. The sheet tension is governed by the blank holder force and the friction condition between the sheet and the tooling. After the forming process, springback in the part is driven by stresses in the material that has experienced stretching, bending and unbending.

In this study in order to validate the springback simulation results, a draw-bend experiment was utilized. This experiment closely mimics an industrial deep-drawing forming process in a lab setting. The advantage of using the draw-bend experiment is that the holding force is not provided by friction but is actively applied and controlled. In this way the material behavior can be tested in isolation from the disturbing factors present in the deep-drawing processes. The schematic representation of the deep-drawing process and the equivalent draw-bend experiment are shown in Figure 3.21. For this study, a draw-bend machine was designed and built at the University of Twente (see Figure 3.22).

The draw-bend machine consists of two electro-mechanical actuators on a triangular frame at a 90 degree angle with respect to each other. Both actuators are capable of delivering 100 kN force and are controlled independently using

Figure 3.21 Schematic illustration of a deep-drawing process (left) and a draw-bend process (right).

a PID controller. The controller is programmed such that one actuator maintains a prescribed constant force while the other actuator pulls the sheet at a constant speed over a certain displacement. During the test, the force of each actuator is measured by a load cell.

During the forming process, the strip is pulled over a roll. The roll can be altered with 5, 10, 15, 20 and 25 mm radius. The roll can be set to rotate freely (almost frictionless condition) or fully fixed. The roll condition can be used to study the effect of the friction condition on springback. However, in this study, in order to reduce the effect of friction, the experiments were performed using the free rotation setting.

Rectangular specimens (600 mm $\times$ 50 mm $\times$ 1 mm) were cut with their length along the rolling direction of DP800 sheet. With one end clamped, the sheet was bent manually over the tool and clamped in the other end. Next, the sheet was subjected to the prescribed holding force F_h . At the next step, the sheet was drawn over the tool radius for 170 mm at 10 mm/s while the holding force was kept constant. The deformation rate was chosen low enough to ensure a near isothermal condition.

After the forming process, the specimen was unloaded and removed from the clamps to springback freely. The specimens were digitized using an image scanner approximately 3 min after forming. Using an image processing technique, the springback angle was evaluated by fitting lines to the undeformed ends of the specimen and calculating the angle between the two lines.

The springback angle used in this context refers to the deviation of the bend angle from 90° after unloading. The draw-bend process is demonstrated schemat-

ically in Figure 3.23.

A roll radius of 15 mm and a holding force $F_h = 13.8$ kN were selected as the settings of the baseline experiment. A combination of different holding forces with the baseline roll radius was tested. Additionally, experiments with the baseline holding force and different roll radii were performed. Each experiment was repeated once.

The springback angle for different holding forces and rolling radii are shown in Figure 3.24 and 3.25 respectively. Expectedly, the springback angle decreases with increasing the holding force and the roll radius.

Figure 3.22 UTwente draw-bend machine.

Figure 3.23 Schematics of the draw-bend experiment [95].

Figure 3.24 Springback profile (left) and angle (right) for different holding forces and $R = 15 \text{ mm}$ (DP800).

Figure 3.25 Springback profile (left) and angle (right) for different roll radii and $F_h = 13.8$ kN (DP800).

3.6 Time-dependent springback

In order to evaluate the time-dependent springback [63] in DP800, the springback angle was measured at different time intervals after the draw-bend test for three specimens. The time-dependent springback angles ($\Delta\theta_t$) for the specimens with the holding force of 13.8, 18.4, 23 kN and a roll radius of 15 mm are presented in Figure 3.26.

The time-dependent springback shows a saturating behavior. The rate of the angle variations is high at the initial stage and then the angle saturates at a constant value after approximately 42 hours. The magnitude of the time-dependent springback is found to decrease as the holding force increases. The long-term springback angle deviation was evaluated by remeasuring the springback angle of all the specimens after a time span of five months. The springback angle for different holding forces and rolling radii right after deformation and after five months are shown in Figure 3.27.

In Figure 3.28 the springback angle variation after five months ($\Delta\theta_\infty$) is plotted as a function of the initial springback angle. The overall trend shows that the time-dependent springback increases with the initial springback angle. This can be attributed to the higher stress gradient through the thickness of the specimens with higher springback.

In Section 3.2 creep and time-dependent anelasticity were introduced as two time-dependent mechanisms that can be responsible for the observed time-

Figure 3.26 The time-dependent springback angle as a function of time.

Figure 3.27 Springback angle for different holding forces (left) and different roll radii (right), right after deformation (black) and after 5 months (red).

Figure 3.28 Time-dependent springback angle versus the initial springback angle

dependent springback. The time-dependent springback is driven by concurrent action of both mechanisms. Yet, creep is possibly the dominant mechanism as the creep strain is orders of magnitude larger than the time-dependent anelastic strain. In Chapter 6 the time-dependent springback is modeled using a power-law creep model fitted to the experimental data presented in Section 3.2.2.

3.7 Summary and conclusions

In this chapter the mechanical behavior of DP600 and DP800 steel grades were investigated using various experiments. With a focus on broadening our understanding regarding the nonlinear unloading behavior, various experiments were conducted. It was observed that the unloading/reloading behavior of the material is complex, showing direction dependency, time-dependent behavior and sensitivity to baking treatment, and is associated with other phenomena such as yield-point phenomena and ratcheting behavior. Based on the results shown in this chapter it can be concluded that the nonlinear unloading behavior cannot be explained by *damage*. However, *dislocation driven anelasticity* and *inhomogeneous deformation at the microscale* remain the two potential mechanisms behind the nonlinear unloading/reloading behavior.

The cyclic behavior of the dual phase steels was investigated using the tension–compression experiments. Following a load reversal, the material shows three characteristics as it compresses: 1. early re-yielding in compression

(Bauschinger effect), 2. a smooth transition from elastic to plastic and 3. a lower flow stress in compression than for monotonic deformation (permanent softening). The smooth transition from nonlinear unloading to Bauschinger effect in compression suggest that the two phenomena are based on similar physics related to dislocation mechanics.

The tension–compression stress–strain data will be used to calibrate the models used for springback simulations in Chapter 6.

A draw-bend test setup was employed to investigate the springback behavior of DP800 steel grade using different roll radii and holding forces. It was shown that the springback angle decreases as the holding force and the roll radius increase. The springback angle was found to increase over time, showing a time-dependent springback behavior. The magnitude of time-dependent springback angle varies from 1° to 6° for different test conditions. In general, the time-dependent springback magnitude increases with the initial springback angle. The draw-bend experiments will be used in Chapter 6 for validation purposes.

Dislocation driven anelasticity

4.1 Introduction

In this chapter, based on the theory of *dislocation driven anelasticity*, a mixed physical-phenomenological model is proposed to describe the observed non-linearity for different levels of pre-strain in a uniaxial setting. This one-dimensional uniaxial model is generalized to a 3D constitutive model incorporating elastic, anelastic and plastic strains.

The concept is demonstrated using an isotropic hardening law and von Mises yield function. The model is calibrated using the LUR experiment on DP800 presented in Chapter 3.

The performance of the model is evaluated by comparing the predicted cyclic unloading–reloading stress-strain curves with the experimental results. It is shown that by incorporating anelastic behavior in the model, the prediction of the cyclic behavior of the material is improved significantly.

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4.2 Theory of dislocation driven anelasticity

The dislocation based anelasticity has been proposed by some researchers as the root cause of the observed nonlinear unloading behavior [2, 3, 9, 13, 15, 49, 65, 85, 88]. The theory claims that during the unloading stage, in addition to the elastic strain, an extra strain is recovered due to the motion of the dislocations below the flow stress of the material. Yet, unlike the plastic deformation, this strain is recoverable. It was claimed by [2, 3, 9, 78, 85, 88] that the reversible motion of the dislocations that are bowed out on a slip plane between two pinning points is the main mechanism responsible for the observed behavior. A schematic representation of the dislocation bow-out is shown in Figure 4.1.

At the unloaded state, the dislocation segments tend to straighten between their pinning points to obtain the lowest defect energy by minimizing the dislocation length. When the material is loaded, the dislocation segments bow out under the influence of the applied stress. At any point, the dislocation line tension T is in balance with the shear stress τ acting on the dislocation. While the applied stress forces the dislocation segment to bow out, the dislocation line tension pulls the dislocation to its initial state. By decreasing the applied stress, the dislocation segments unbow and result in recovery of anelastic strain. The magnitude of the anelastic strain is proportional to the area A that is swept by the dislocation during the bowing and unbowing process and the number of the dislocation segments. The former is a function of the applied stress and the latter is dependent on the dislocation density in the material.

Schoeck [78] derived an expression for anelastic shear modulus by approximating the shear stress–strain relation for a pinned segment of a dislocation that is bowed out under the influence of a shear stress. Van Liempt and Sietsma [88] extended the model of Schoeck to account for stresses up to the yield stress and converted the shear stress–strain relation to the von Mises equivalent stress–strain observed in experiments. Their model could give an accurate description of the material behavior in the pre-yield region by fitting the dislocation density and the dislocation segment length to the experimental data.

Arechabaleta et al. [2] used the model proposed by van Liempt and Sietsma to quantify the dislocation structure in pure iron and a low-alloy steel. They obtained the dislocation density and the effective dislocation segment length by fitting the model to the experimental data at different pre-strain levels. They confirmed the dislocation densities obtained from the fit by XRD mea-

Figure 4.1 Schematic of the bow-out mechanism of a dislocation segment.

surements.

4.3 The anelastic strain model

A number of studies on the nonlinear unloading behavior used the dislocation density to quantitatively describe the phenomenon [28, 49, 69, 87, 88]. Hull and Bacon [45] state that the magnitude of the recoverable anelastic strain is proportional to the dislocation density according to

$$\varepsilon^{\text{an}} = \rho b \bar{x} \quad (4.1)$$

where ρ is the dislocation density, b is the Burgers vector and $\bar{x}$ is the average distance moved by dislocations. As the material plastically deforms, the dislocation density increases, which leads to an increase in the flow stress. The Taylor equation is one of the most established expressions relating the flow stress of a material to its dislocation density [83] according to

$$\sigma_f = \sigma_0 + \bar{M} \alpha G b \sqrt{\rho} \quad (4.2)$$

where $\bar{M}$ is the Taylor factor, α is the dislocation strengthening parameter and a material related constant, G is the shear modulus of the material, ρ is the dislocation density and σ_0 is the lattice friction stress of the material in the absence of dislocation interactions. In order to replace σ_0 with the flow stress of the as-received material σ_y , Equation (4.2) can be rewritten as [90]

$$\sigma_f = \sigma_y + \bar{M} \alpha G b (\sqrt{\rho} - \sqrt{\rho_0}) \quad (4.3)$$

where ρ_0 is the dislocation density of the material when it becomes plastically deformed at σ_y . Rewriting Equation (4.3) yields an expression for the dislocation density evolution as a function of hardening behavior of the material

according to

$$\rho = \left(\frac{\sigma_f - \sigma_y}{M\alpha Gb} + \sqrt{\rho_0} \right)^2 \quad (4.4)$$

Using Equation (4.4) in combination with Equation (4.1), the evolution of the anelastic strain is given as a function of the work hardening behavior of the material as

$$\varepsilon_t^{\text{an}} = \left(K (\sigma_f - \sigma_y) + \sqrt{\varepsilon_0^{\text{an}}} \right)^2 \quad (4.5)$$

In order to fit Equation (4.5) to the experimental data, first the yield stress σ_y has to be determined. The value of σ_y can be approximated from the Kocks–Mecking plot. In the Kocks–Mecking plot the first derivative of the stress–strain curve is plotted against the stress. At the yield stress the slope of the Kocks–Mecking plot changes abruptly. The details of the determination of the yield stress from Kocks–Mecking plots are discussed in [88].

The yield stress obtained from this method has a physical motivation and marks the stress at which Frank-Read sources become activated. This is unlike the $\sigma_{0.2}$ value that is determined using the arbitrary 0.2% offset criterion.

For quantitative estimation of σ_y , the Kocks–Mecking plot of the stress–strain data of the material is constructed in Figure 4.2. In Figure 4.2 two distinctive regions can be recognized. At the beginning of loading the instantaneous modulus decreases slowly from the theoretical value of the E-modulus, then drops rapidly in the first region, abruptly leveling off to a slower decline rate in the second region where plastic deformation takes place. The transition point between the first region and the second region is regarded as the yield point.

From the Kocks–Mecking plots, the yield stress of DP800 in the rolling direction was found to be approximately 460 MPa. Substituting $\sigma_y = 460$ MPa into Equation (4.5), the model can be fitted to the experimental data taking K and $\varepsilon_0^{\text{an}}$ as the fitting parameters (Figure 4.3).

As shown in Figure 4.3, Equation (4.5) is fitted to the LUR experimental data of DP800-RD presented in Chapter 3. In Figure 4.3, the total recovered anelastic strain is plotted against the flow stress (i.e. the stress from which the material was unloaded). It should be noted that the fitting was performed only on the data set obtained from experiments of 0.0005 s^{-1} strain rate. The model for anelastic strain fits very well for all flow stress values with only two fitting parameters.

The total recoverable anelastic strain upon unloading to zero stress is given by Equation (4.5). Nevertheless, another equation is needed to describe the

Figure 4.2 The Kocks–Mecking plot of DP800-RD showing a well defined yield point.

Figure 4.3 Variation of the total recovered anelastic strain with unloading stress in DP800-RD.

Figure 4.4 Normalized anelastic strain ($\varepsilon^{\text{an}}/\varepsilon_t^{\text{an}}$) as a function of unloading stress.

nonlinearity upon unloading. A phenomenological approach is taken to identify the function describing nonlinearity. From the experimental results it has been found that unloading behavior can be well described according to

$$\varepsilon^{\text{an}} = \left(K (\sigma_f - \sigma_y) + \sqrt{\varepsilon_0^{\text{an}}} \right)^2 \frac{\sinh(\kappa s)}{\sinh(\kappa)} \quad (4.6)$$

where κ is a fitting parameter and s is the dimensionless normalized stress defined as

$$s = \frac{|\Delta\sigma|}{\sigma_f} \quad (4.7)$$

where $\Delta\sigma$ is the stress decrement upon unloading from the flow stress.

The value of κ was obtained from fitting Equation (4.6) to the experimental data shown in Figure 4.4 using the least squares method using all the unloading cycles. Equation (4.6) represents the observed unloading behavior very well.

As a simplification to the model, the nonlinear reloading curve is assumed to be symmetric with the unloading curve along the chord modulus line. Hence, the reloading curve can be described using the same function taking $\Delta\sigma$ in Equation (4.7) as the increment of stress from zero stress. The anelastic model parameters are summarized in Table 4.1.

In order to describe the anelastic model in terms of the equivalent plastic strain ε^p , σ_f in Equation (4.6) can be substituted with a hardening law. For

Table 4.1 Anelastic model parameters for DP800.

K (MPa $^{-1}$)	σ_y (MPa)	$\varepsilon_0^{\text{an}}$	κ
$5.1 \cdot 10^{-5}$	460	$3 \cdot 10^{-4}$	0.68

Table 4.2 Swift hardening parameters for DP800.

C (MPa)	ε_0	n
1710	0.0072	0.28

simplicity, the Swift hardening law is used which is given as

$$\sigma_f = C (\varepsilon_0 + \varepsilon_{\text{eq}}^p)^n \quad (4.8)$$

where C is the strength coefficient and n is the hardening exponent. The monotonic tensile experiment was used to determine the parameters for the Swift hardening law. These parameters are given in Table 4.2.

Substituting Equation (4.8) in Equation (4.6) for σ_f gives an expression for the anelastic strain.

4.4 Constitutive modeling

In order to implement the model in a FEM code, the constitutive rate equations are required. Additive decomposition of the total strain rate into elastic $\dot{\varepsilon}^e$, anelastic $\dot{\varepsilon}^{\text{an}}$ and plastic $\dot{\varepsilon}^p$ strain rates yields

$$\dot{\varepsilon} = \begin{cases} \dot{\varepsilon}^e + \dot{\varepsilon}^{\text{an}}, & \text{if } \phi < 0 \\ \dot{\varepsilon}^e + \dot{\varepsilon}^{\text{an}} + \dot{\varepsilon}^p, & \text{if } \phi = 0 \end{cases} \quad (4.9)$$

where ϕ is the yield function. The rate of the elastic strain is given according to

$$\dot{\varepsilon}^e = \mathbb{E}^{-1} : \dot{\sigma} \quad (4.10)$$

where $\mathbb{E}$ is the isotropic elasticity tensor. For Drucker postulated associated flow, the rate of plastic strain is given as

$$\dot{\varepsilon}^p = \dot{\lambda} \frac{\partial \phi}{\partial \sigma} \quad (4.11)$$

Assuming the same flow potential for the anelastic strain, the rate of the anelastic strain can be written as

$$\dot{\epsilon}^{\text{an}} = \dot{\xi} \frac{\partial \phi}{\partial \sigma} \quad (4.12)$$

where the scalar multiplier $\dot{\xi}$ can be written as

$$\dot{\xi} = \frac{3}{2} \dot{\epsilon}^{\text{an}} \quad (4.13)$$

Here the scalar value $\dot{\epsilon}^{\text{an}}$ is the time derivative obtained from Equation (4.6), whereas $\dot{\epsilon}$ in Equation (4.12) is a symmetric second order tensor.

The assumption of the associated flow rule is justified supposing that the anelastic strain is resulted from the motion of dislocations distributed in a polycrystalline material under the action of stress. The dislocations contributing to the anelastic strain were generated and oriented on the slip planes which were activated during the plastic deformation. Therefore, it is assumed that the net anelastic strain rate tensor is collinear with the normal of the yield function.

In order to model the hysteresis behavior observed in unloading–reloading cycles, a load reversal criterion is introduced according to

$$\dot{\sigma}_n : \dot{\sigma}_{n+1} < 0 \quad (4.14)$$

Equation (4.14) implies that the load reversal condition is satisfied when the relative angle between the two successive stress increments exceeds $\frac{\pi}{2}$ [71]. The $\Delta\sigma$ in Equation (4.7) is evaluated from the point where the load reversal condition is satisfied.

Clearly, this formulation is independent of the flow function and the hardening law; however, for demonstration of the model, the von Mises flow function and the Swift hardening law are incorporated in the following section.

4.5 Incremental solution algorithm

For the implementation of the model an implicit numerical scheme is used for return mapping that is proposed by Wilkins [97]. More details on the formulations can also be found in [80].

The rate equations presented in the previous section are solved using the Backward Euler scheme. Therefore, the equations are evaluated with state variables

at the end of the time step. In the following equations, subscript n and $n + 1$ are used to denote a quantity at the beginning and the end of a time step respectively.

At the end of each time step the evolution of two independent variables is determined:

$$\Delta V = \left\{ \begin{array}{c} \Delta \boldsymbol{\sigma} \\ \Delta \lambda \end{array} \right\} \quad (4.15)$$

where $\Delta V = V_{n+1} - V_n$. Within every step the independent variables are updated at every iteration according to

$$\Delta V_n^{i+1} = \Delta V_n^i + \delta V_n^{i+1} \quad (4.16)$$

where the superscript i is denoted to the iteration number within the time step.

The set of residual functions for every increment at the i^{th} iteration is elaborated as

$$\mathbf{R}_\sigma^i = \Delta \boldsymbol{\varepsilon} - \mathbb{E}^{-1} : \Delta \boldsymbol{\sigma} - \Delta \boldsymbol{\varepsilon}^{\text{an}} \frac{\partial \phi}{\partial \boldsymbol{\sigma}} - \Delta \lambda \frac{\partial \phi}{\partial \boldsymbol{\sigma}} \quad (4.17)$$

and

$$R_\phi^i = -\phi^i \quad (4.18)$$

Equation (4.17) is derived from the sum of the elastic, anelastic and plastic strain increments and Equation (4.18) results from the consistency condition.

The linearization of the residual functions and casting them into a system of equations yields

$$\left[\begin{array}{c} \mathbf{M}^i \end{array} \right] \left\{ \begin{array}{c} \delta \boldsymbol{\sigma}^{i+1} \\ \delta \lambda^{i+1} \end{array} \right\} = \left\{ \begin{array}{c} \mathbf{R}_\sigma^i \\ R_\phi^i \end{array} \right\} \quad (4.19)$$

where $\mathbf{M}$ contains the derivatives of the residual functions with respect to the independent variables:

$$\left[\begin{array}{c} \mathbf{M}^i \end{array} \right] = \left[\begin{array}{cc} \frac{\partial \mathbf{R}_\sigma^i}{\partial \boldsymbol{\sigma}} & \frac{\partial \mathbf{R}_\sigma^i}{\partial \lambda} \\ \left(\frac{\partial R_\phi^i}{\partial \boldsymbol{\sigma}} \right)^T & \frac{\partial R_\phi^i}{\partial \lambda} \end{array} \right]^T \quad (4.20)$$

The components of the M matrix are elaborated according to

$$\frac{\partial \mathbf{R}_\sigma^i}{\partial \boldsymbol{\sigma}} = -\mathbb{E}^{-1} - \frac{\partial \varepsilon^{\text{an}}}{\partial \boldsymbol{\sigma}} \frac{\partial \phi}{\partial \boldsymbol{\sigma}} - \Delta \varepsilon^{\text{an}} \frac{\partial^2 \phi}{\partial \boldsymbol{\sigma}^2} - \Delta \lambda \frac{\partial^2 \phi}{\partial \boldsymbol{\sigma}^2} \quad (4.21)$$

$$\frac{\partial \mathbf{R}_\sigma^i}{\partial \lambda} = -\frac{\partial \phi}{\partial \lambda} - \frac{\partial \varepsilon^{\text{an}}}{\partial \lambda} \frac{\partial \phi}{\partial \boldsymbol{\sigma}} \quad (4.22)$$

$$\frac{\partial R_\phi^i}{\partial \boldsymbol{\sigma}} = -\frac{\partial \phi}{\partial \boldsymbol{\sigma}} \quad (4.23)$$

$$\frac{\partial R_\phi^i}{\partial \lambda} = -\frac{\partial \phi}{\partial \lambda} = -\frac{\partial \phi}{\partial \sigma_f} \frac{\partial \sigma_f}{\partial \varepsilon_{\text{eq}}^p} \frac{\partial \varepsilon_{\text{eq}}^p}{\partial \lambda} \quad (4.24)$$

The equations above are derived for plastic deformation. If there is no plasticity (i.e. $\phi < 0$), the stress increment is calculated from Equation (4.17) with $\Delta \lambda = 0$.

4.6 Tangent matrix

For the global FEM Newton–Raphson process, a tangent matrix at the integration point is required. The tangent matrix is determined as

$$\mathbf{K} = \frac{\delta \boldsymbol{\sigma}}{\delta \boldsymbol{\varepsilon}} \quad (4.25)$$

A perturbation method is used to determine the tangent matrix by means of the general expression in Equation (4.19). A relation between the incremental stress $\delta \boldsymbol{\sigma}$ and incremental strain $\delta \boldsymbol{\varepsilon}$ is established according to

$$\begin{bmatrix} \mathbf{M}_{11} & \mathbf{M}_{12} \\ \mathbf{M}_{21} & \mathbf{M}_{22} \end{bmatrix} \begin{Bmatrix} \delta \boldsymbol{\sigma} \\ \delta \lambda \end{Bmatrix} = \begin{Bmatrix} \delta \boldsymbol{\varepsilon} \\ 0 \end{Bmatrix} \quad (4.26)$$

Expanding Equation (4.26) yields

$$\mathbf{M}_{11} \delta \boldsymbol{\sigma} + \mathbf{M}_{12} \delta \lambda = \delta \boldsymbol{\varepsilon} \quad (4.27)$$

and

$$\mathbf{M}_{21} \delta \boldsymbol{\sigma} + \mathbf{M}_{22} \delta \lambda = 0 \quad (4.28)$$

After eliminating $\delta\lambda$ from Equation (4.27) using Equation (4.28) and rewriting the equation, the tangent matrix is obtained as

$$\mathbf{K} = \frac{\delta\boldsymbol{\sigma}}{\delta\boldsymbol{\varepsilon}} = (\mathbf{M}_{11} - \mathbf{M}_{12}\mathbf{M}_{22}^{-1}\mathbf{M}_{21})^{-1} \quad (4.29)$$

4.7 Model validation

In order to evaluate the performance of the proposed model in predicting the nonlinear unloading–reloading cycles, the stress–strain curves obtained from the model and the LUR experiment are compared. The stress–strain curves obtained from the model prediction and the experiment and a magnified view of the 5th, 6th and 7th cycles are plotted in Figure 4.5. For comparison, the unloading paths predicted by the chord modulus and E-modulus are plotted alongside the anelastic model in the bottom right of Figure 4.5.

As can be seen in Figure 4.5, the predicted unloading curve using the anelastic model is very well in agreement with the experimental data; whereas, the unloading paths obtained by purely elastic unloading and unloading using the chord modulus tend respectively to underpredict and overpredict the strain during unloading. The chord modulus gives an accurate prediction of the strain only when the material is fully unloaded to zero stress.

The difference between the strain from the model prediction and from the experimental data is used as a measure of error. The root mean square (RMSE) of the strain deviation is given as

$$\text{RMSE} = \sqrt{\frac{\sum_{i=1}^n e_i^2}{n}} \quad (4.30)$$

where e_i is the difference between the strain predicted by the model and the experimental data at the i^{th} data point during unloading.

The strain deviation between the experimental data and the prediction of each model for the 7th unloading cycle is plotted in Figure 4.6 (left). Clearly the highest deviation from the experimental data is seen when the pure elastic model is used to predict the unloading behavior. The chord modulus approximation yields an exact prediction only at zero stress while for the points that were unloaded halfway the chord modulus is not accurate. This is a source of error in springback prediction of an industrial forming process where some residual stresses are present in the part after the springback. It can clearly be

Figure 4.5 Stress–strain response of DP800-RD with LUR cycles (top left); magnified view of the 5th cycle (top right), 6th cycle (bottom left) and 7th cycle (bottom right).

seen that this error has been significantly reduced using the nonlinear anelastic model.

The RMSE obtained from the unloading curves at different plastic pre-strains are plotted in Figure 4.6 (right). The error in strain prediction during unloading using the linear pure elastic unloading approach increases as the plastic strain increases. This error is significantly lower when the chord modulus approach is adopted. Lastly, the unloading curve predicted by the anelastic model shows a very small deviation from the experimental data compared to the other two approaches. The RMSE stays below 0.00005 for plastic strains up to 5.5%. A comparison between the sum of the RMSEs of all the cycles for each approach suggests that using the proposed model, an improvement by a factor of 8 and 14 is obtained in comparison with the chord modulus and

Figure 4.6 The strain deviation between the experiment and the prediction of anelastic model, chord modulus and E-modulus (left); RMSE of strain prediction during unloading for the anelastic model, chord modulus and E-modulus (right).

E-modulus approaches respectively.

4.8 Summary and conclusions

In this chapter, based on the theory of dislocation driven anelasticity a model was developed to predict the observed nonlinearity for different levels of pre-strain. As the material deforms plastically, the magnitude of the anelastic strain increases. This results in a reduction of the chord modulus. The evolution of anelastic strain corresponds to the dislocation density in the material and can be related to the hardening behavior of the material through the Taylor equation. According to the proposed model, a 2nd order polynomial relation between the anelastic strain and the flow stress of the material was established. The model was calibrated to the experimental data obtained from LUR experiments presented earlier in Chapter 3.

Assuming the associated flow rule for the anelastic strain, the model was generalized to a 3D constitutive model incorporating elastic, anelastic and plastic strains. For the implementation of the model an implicit, backward Euler solution algorithm was used.

Comparing the model prediction with the experimental data showed a significant improvement over strain prediction incorporating the anelastic behavior compared to the E-modulus and chord modulus. The model is capable of predicting the nonlinear unloading/reloading behavior in sequential LUR cycles with great accuracy. Hence, using the proposed model in FE simulations of forming processes will result in better prediction of the residual stresses and thus springback behavior of complex parts. The model presented in this chapter will be used in Chapter 6 for FEM simulation of the draw-bend process.

The model presented here is advantageous from the point of view that it can be directly calibrated with the LUR experimental data with a minimum number of fitting parameters. On top of that, it is efficient to be used for an industrial forming simulation and is compatible with the classical elasto-plastic formulation.

On the other hand, the anelastic model relies on a load reversal criterion. This criterion is established to predict the experimentally observed hysteresis behavior upon load reversal. However, this behavior is not explained by the dislocation bow-out mechanism.

Stress inhomogeneity in polycrystalline metals

The observed macroscale behavior is directly linked to the material mechanics at the microscale that is governed by complex interaction with and in between the grains. Therefore, in order to study and describe macroscopically observed behavior, it is important to understand the material physics at microscale.

In recent years, material modeling at microstructural level has become popular. Various researchers have used the finite element method in conjunction with crystal plasticity and continuum plasticity models. Often, the analysis is performed using a representative volume element (RVE). A RVE is a small volume of microstructure that represents the general characteristics of the whole material. Various levels of details are added to an RVE by taking into account for instance the crystallographic orientation of the grains, grain size and morphology and spatial distribution of the second phase in the matrix.

Most of the microstructure models aim to investigate the influence of the microstructure on flow behavior, anisotropy and formability. Only few studies have been dedicated specifically to investigate the effect of the inhomogeneous microstructure on the unloading behavior [10, 32]. Govik et al. [32] used a 3D RVE model with continuum plasticity to investigate the unloading behavior of a dual phase steel. They reported that local plastic deformation in ferrite due to interaction between the ferrite and martensite is partially responsible for the nonlinear unloading behavior. In a recent work, Bong et al. [10] made use of a 3D RVE consisting of 27 grains to model the microstructure of a DP980 dual phase steel. Contrary to Govik, they claimed that the stress inhomogeneity

between the ferrite and martensite has no effect or a minor effect on the nonlinear unloading behavior. However, their model could capture the nonlinear unloading behavior only after including the elastic dislocation interactions in their crystal plasticity based model. The authors concluded that the nonlinear unloading/reloading and the Bauschinger effect are resulted from “the elastic interactions of large populations of discrete dislocations”.

In the present chapter, the inhomogeneous deformation at the microstructure is discussed as a potential explanation for the nonlinear unloading behavior and subsequent Bauschinger effect. The stress and strain partitioning in a dual phase steel is assessed using the crystal plasticity finite element modeling (CPFEM) approach on a realistic RVE consisting of ferritic and martensitic grains.

A computationally efficient model based on a numerical homogenization scheme is presented and the model performance in predicting the stress–strain response with respect to the nonlinear unloading behavior and Bauschinger effect is evaluated.

5.1 Sources of inhomogeneity

In macroscale material mechanics, the material is usually assumed to be homogeneous. However, at the microscale the material behavior in general is inhomogeneous. This is a consequence of the inhomogeneous nature of the polycrystalline materials being composed of grains with different orientations, sizes, phases, etc. The material behavior at the grain scale is highly anisotropic and size dependent. Hence, the mechanical response of a grain is closely linked to its orientation and morphology. Besides, alloyed materials often crystallize in more than one phase each with different mechanical properties. All these effects result in partitioning of stress and strain within the microstructure. In this section the main sources resulting in inhomogeneity of stress and strain distribution in polycrystals are discussed.

5.1.1 Grain orientation distribution

Crystalline materials deform plastically by slipping on defined crystallographic planes in certain directions, known as slip systems. In order to initiate slip, the shear stress resolved in the direction of the slip has to reach a critical value. Hence, for a certain applied load, the resolved shear stress depends on

Figure 5.1 The grain map of DP600 constructed from the EBSD data.

the orientation of the grain. The grains which are oriented such that they have a higher resolved shear stress are more likely to deform, giving rise to crystallographically soft and hard grains. Figure 5.1 shows the orientation map collected using the electron backscattered diffraction (EBSD) orientation mapping method to reproduce the grain structure of DP600.

As well as the orientation dependent plastic behavior, grains exhibit elastically anisotropic behavior. This leads to inhomogeneous elastic behavior between grains oriented differently. As a result, an additional stress partitioning between grains develops during elastic deformation. The general elasticity tensor of cubic crystals in Voigt notation is given as

$$\mathbf{E} = \begin{pmatrix} c_{11} & c_{12} & c_{12} & 0 & 0 & 0 \\ \bullet & c_{11} & c_{12} & 0 & 0 & 0 \\ \bullet & \bullet & c_{11} & 0 & 0 & 0 \\ \bullet & \bullet & \bullet & c_{44} & 0 & 0 \\ \bullet & \bullet & \bullet & \bullet & c_{44} & 0 \\ \bullet & \bullet & \bullet & \bullet & \bullet & c_{44} \end{pmatrix} \quad (5.1)$$

where c_{11} , c_{12} and c_{44} are independent elastic coefficients. The elastic constants for a BCC iron single crystal are given as: $c_{11} = 242$ GPa, $c_{12} = 150$ GPa and $c_{44} = 112$ GPa [44]. The anisotropic elastic modulus of iron in 3D space is presented in Figure 5.2.

5.1.2 Variation in grain size

For more than half a century the Hall–Petch [36, 76] relation has been known to link the yield strength and the mean grain size of a polycrystalline metal.

Figure 5.2 3D representation of the anisotropic elastic modulus of iron with body centered cubic (BCC) structure.

On the basis of the dislocation pile-up model developed by Eshelby et al. [22], the Hall–Petch relation was mathematically expressed as

$$\sigma_f = \sigma_0 + kd^{-0.5} \quad (5.2)$$

where k is the material constant, σ_0 is the lattice friction stress and d is the average grain diameter. Later on, the Hall–Petch effect was confirmed by several experimental studies [4, 5, 39, 62]. Yet, it has been pointed out that for very large grain sizes a d^{-1} dependence was found [43]. Furthermore, experiments on many nanocrystalline materials showed that for grain sizes smaller than 10 nm, the yield strength would either remain constant or decrease with decreasing grain size [11].

However, in Hall–Petch formulation, only the mean grain size has been taken into account and the fact that the grains form a population of a stochastic nature with various sizes is not considered. It can be deduced that in the polycrystalline metals the larger grains are more prone to yield than the smaller ones. This leads to plastic inhomogeneity due to the grain size dispersion. The histogram of the grain size distribution obtained from EBSD mapping of a DP600 steel is plotted in Figure 5.3 (left). The grain size distribution is assessed by the equivalent circular diameter where the grain area is measured and the average grain diameter is calculated from a circle with an equivalent area. From the grain size distribution, the resulting distribution of the Hall–Petch strengthening ($\sigma_{hp} = kd^{-0.5}$) using a Hall–Petch coefficient of

Figure 5.3 Distribution of the grain size in DP600 obtained from EBSD measurement (left) and distribution of the Hall–Petch strengthening resulting from the grain size distribution (right).

$k = 300 \text{ MPa} \mu\text{m}^{0.5}$ [46] is plotted in Figure 5.3 (right). It can be seen from Figure 5.3 that the variations in grain size lead to a large dispersion of the local yield stress under the Hall–Petch assumption.

5.1.3 Coexistence of different phases

It is generally accepted that the mechanical properties of the phases are different, especially in the case of dual phase steels where a large contrast in mechanical properties of ferrite and martensite exists. In dual phase steels, strain partitioning between the ferrite and martensite phases has been observed experimentally by performing in situ mechanical testing in combination with microstructural imaging during deformation. The digital image correlation (DIC) technique has been used to obtain the strain field during deformation from the captured images. It has been indicated that the strain accommodated by the ferrite is significantly larger than what is experienced by the martensite. Such studies are useful in understanding not only the deformation behavior of each phase but also the interaction between the two phases. As an example, it was reported that strain tends to localize in ferrite grains surrounded by martensite islands while the strain was found to be more homogeneous in regions occupied largely by ferrite grains [38]. This suggests that the inhomogeneity in strain distribution is affected not only by the martensite content but also by the size and distribution of the martensite islands in the ferritic matrix.

5.1.4 Residual stresses

Residual stresses are usually categorized into types I, II and III across the scales. Type I residual stresses are macroscopic. They appear over the structure involving a large number of grains. Type II residual stresses occur across a number of grains (intergranular) associated with microscopic inhomogeneities. This type of residual stress is very common in polycrystals due to the mismatch between the mechanical properties of the neighboring grains. Type III residual stresses refer to the stresses within a grain (intragranular) at a sub-micron scale. The type III residual stresses are a result of interaction between the dislocations stress fields and other crystalline defects. The stresses that arise inside the grain due to dislocation pile-ups are an example of type III residual stress.

The development of residual stresses in the material is intrinsic to the material production and subsequent forming process. Sheet materials undergo several thermo-mechanical processing stages during the production phase. Substantial residual stresses develop in the material through the ensuing process steps. Particularly in the case of dual phase steel, stresses are induced during the austenite to martensite phase transformation. Austenite has a face centered cubic (FCC) crystal structure while martensite has a body centered tetragonal (BCT) crystal structure which is less dense. Therefore, the austenite to martensite phase transformation is accompanied by a volume expansion. The volume expansion results in residual stresses in the martensite grain and the surrounding ferritic matrix. The absence of yield point phenomena in the dual phase steels is generally accredited to formation of dislocations in ferrite during the phase transformation [25, 61]. On top of that, due to the microstructural inhomogeneities discussed earlier, the plastic deformation induced during the final rolling operations (i.e. skin pass and temper rolling) results in further residual stresses in the sheet.

5.2 Modeling framework

In this section, inhomogeneous deformation in the material is studied using two modeling approaches: one based on a full-field crystal plasticity finite element method (CPFEM) and the other based on the mean-field elasto-plastic self-consistent (EPSC) approach.

The CPFEM model combines the crystal plasticity theory of the crystal lat-

tice deformation with the finite element discretization method to simulate the stresses and strains within the microstructure. This model is used as a tool to study the stress and strain partitioning that arises from the grain misorientations and the mechanical contrast between the ferritic and martensitic grains. The model used in this analysis is built based on the microstructure of the DP600 dual phase steel presented in Figure 5.1.

In the EPSC approach, the grains are idealized as spherical inclusions inside a homogenized matrix. The mechanical behavior of the inclusions is modeled using the elasto-plastic constitutive law. In this model, the inhomogeneous deformation in the material is taken into account by introducing a distributed yield stress among the constituents. The EPSC model is used to model the cyclic mechanical response of a DP600 dual phase steel.

5.2.1 Crystal plasticity finite element modeling (CPFEM)

In order to study the role of the grain orientations and the hard phase on the stress-strain partitioning in the polycrystalline material, a CPFEM model is used. In CPFEM the finite element mesh is used to represent the microstructure. The mechanical response at each integration point is described by the single crystal constitutive model. Using this approach the localized deformation in the individual grains can be modeled.

The analysis was performed on a realistic microstructure constructed based on the crystal orientations obtained from the EBSD measurements as shown in Figure 5.1. In this way the microstructure is generated using the Voronoi tessellation algorithm [50]. An additional microstructural quality is introduced to the generated microstructure by considering the aspect ratio of the grains. The crystallographic orientations are assigned to grains such that they represent the material's texture.

One feature typically observed in dual phase steels is martensite banding. During the processing stage, the martensite grains tend to form a banded structure due to segregation of the alloying elements [89]. The martensite banding was introduced to the model by concentrating the martensite grains in rows. The resulting microstructure is presented in Figure 5.4. The martensite grains are colored in black. The model consists of 84 grains of which 34 are martensite accounting for 14% of the total volume fraction.

The crystal plasticity code employed in this study is a unique code based on rate-independent elasto-plastic formulation developed at University of Twente

Figure 5.4 The grain mapping of the CPFEM RVE.

by E.E. Aşık and E.S. Perdahcioğlu [74]. More details on the rate-independent elasto-plastic crystal plasticity formulation can be found in [67]. Simulations were carried out in 2D using the commercial finite element code ABAQUS in which the elasto-plastic crystal plasticity model is incorporated with a user defined material subroutine UMAT. 9102 quadratic triangular plane strain (CPE6) elements were used to discretize the microstructure. Periodic boundary conditions were applied to the edges of the mesh and the part was stretched uniaxially by 10% in the rolling direction.

Figure 5.5 (top left) shows the spatial distribution of the martensitic grains (black) in the ferritic matrix (gray) in the deformed microstructure. The local plastic strain distribution in the RVE is plotted in Figure 5.5 (top right). It is observed that the strain tends to localize in the ferritic phase, forming shear bands under a 45 degree angle with the tensile direction. On top of that, a strong interaction between ferrite and martensite is observed. Some locations with high strain concentration are seen in the ferrite where it is wedged by surrounding martensite islands, while some other larger ferritic grains in the vicinity of martensitic grains deform less.

The distribution of the first stress component σ_{11} after 10% stretch before and after unloading is plotted in the bottom left and right of Figure 5.5 respectively. Naturally, when the material is subsequently compressed, the grains which are already in compression will yield earlier, giving rise to appearance of the Bauschinger effect.

Figure 5.5 Distribution of martensitic grains in ferritic matrix (top left). Distribution of the sum of shear strains (top right). Distribution of σ_{11} before unloading (bottom left). Distribution of σ_{11} after unloading (bottom right).

Comparing the stress distribution between the two phases reveals that the stress concentrates in the martensitic grains. Large stress concentration is also present in the ferritic grains situated between martensitic grains. This results in formation of stress concentrated regions along the martensite bands. Right after unloading, as shown in Figure 5.5 (bottom right), the martensitic grains are still largely in tension while the surrounding ferritic grains are in compression.

The statistical distribution of the first stress component σ_{11} after 10% stretch before and after unloading is plotted in Figure 5.6. In the unloaded state, the peak of the distribution shifts to zero stress. The distribution shows that a significant portion of the material is in compression when the macroscopic stress is zero.

The evolution of the σ_{11} distribution with respect to the RVE's strain is shown in Figure 5.7. As the material deforms further, the distribution shifts to higher stresses and becomes wider. This indicates that the stress distribution is becoming more inhomogeneous with plastic deformation. The mean of the distribution coincides with the uniaxial stress applied to the RVE showing that the scale transition in the RVE is consistent. This is shown in Figure 5.8 where

Figure 5.6 The distribution of the σ_{11} stress in the RVE before (red) and after unloading(gray).

the stress–strain response of the RVE is plotted along with the mean value of σ_{11} distribution. The bar height represents the standard deviation of the σ_{11} distribution in the RVE. The standard deviation gives a quantitative measure of the stress inhomogeneity in the material.

The mechanical response of the RVE subjected to a cyclic load is plotted in Figure 5.9. The model is capable of capturing the main features of the cyclic behavior of the material, such as the Bauschinger effect, observed as the smooth elastic to plastic transition, and the nonlinear unloading/reloading behavior. The material parameters in the CPFEM model are not tuned to accurately predicting the mechanical response of the material at hand. Therefore, it is futile to quantitatively compare the simulated curves with the experimental data. Generally speaking, in comparison with the experimental observations, the CPFEM model overpredicts the macroscopic stress in compression and underpredicts the magnitude of the nonlinear strain recovery.

The nonlinear unloading behavior appears as a result of an increasing plastic deformation in the microstructure during unloading. This is shown in Figure 5.10 where the portion of the material that is becoming plastic during unloading is plotted as a function of the macroscopic stress at different levels of pre-strain. From Figure 5.10 it can be seen that during unloading some elements of the CPFEM model deform plastically. The plastic portion of the material rapidly increases as the macroscopic stress approaches zero. More-

Figure 5.7 The evolution of the σ_{11} stress distribution with the RVE's deformation.

Figure 5.8 The mean and the standard deviation of the σ_{11} stress.

Figure 5.9 The simulated stress–strain response of the RVE subjected to a tension–compression load (left) and subjected to a unloading–reloading cycle (right).

over, the plastically deforming portions of the material increase with increasing the pre-strain before unloading. This observation is in agreement with the increasing nonlinear unloading behavior and Bauschinger effect as the material deforms further.

5.2.2 Mean-field homogenization

Homogenization refers to the procedure of replacing a inhomogeneous material with an equivalent homogeneous material (see Figure 5.11). In this procedure the fields within a domain are solved and the averaged results are projected on the boundary of the domain [73]. This domain is normally selected to be the RVE of the material of interest. The RVE analysis can be conducted using full-field or mean-field approaches. In the full-field approach, the field calculations are conducted on the RVE using finite element method (FEM) or fast Fourier transform (FFT) discretization. As was shown in Section 5.2.1, the full-field approach yields a detailed description of stress and strain distribution in the microstructure. However, the computational cost is overwhelming, making its application for structural analysis limited.

An alternative is the mean-field approach which is based on analytical solutions and gives an approximation of the mean value of the stress and strain fields at the inhomogeneities as well as the overall response. The two most basic schemes are Taylor (iso-strain) and Reuss (iso-stress). Taylor assumes a uniform strain field in the RVE. Hence, all the inhomogeneities undergo the same amount

Figure 5.10 The proportion of the material that is plastically deforming during unloading as a function of the stress normalized by unloading stress (σ_f).

of deformation. On the other hand, Reuss assumes a uniform stress field. Therefore, the stress is distributed equally between the inhomogeneities. A survey of the theory of mean-field homogenization is given in [73]. For the sake of completeness a brief overview is given here.

One of the main concepts in the homogenization approach is the scale transition. The fields calculated in the RVE need to be averaged and transferred to the macroscale. This is performed by using an averaging operator defined as

$$\langle g(\mathbf{x}) \rangle_\omega = \frac{1}{V} \int_\omega g(\mathbf{x}) \, dV = \bar{g} \quad (5.3)$$

where g is a field defined over the domain ω , $\bar{g}$ is the macroscopic value and $\langle g(\mathbf{x}) \rangle$ is the averaged value of g over the coordinates $\mathbf{x}$ of the volume ω . From Equation (5.3) it can be deduced that the volume can be split into sub-domains:

$$\langle g(\mathbf{x}) \rangle_\omega = \frac{1}{V} \sum_k \int_{\omega_k} g(\mathbf{x}) \, dV_k, \quad \omega = \bigcup_k \omega_k \quad (5.4)$$

Rewriting Equation (5.4) in terms of volume fractions of the sub-domains yields

$$\langle g(\mathbf{x}) \rangle_\omega = \sum_k f_k \langle g(\mathbf{x}) \rangle_{\omega_k}, \quad f_k = \frac{V_k}{V} \quad (5.5)$$

According to the Hill-Mandel condition (Hill [40] and Mandel [66]), the mechanical work computed in both scales should be consistent. The Hill-Mandel condition in mathematical terms is expressed as

$$\bar{\sigma} : \bar{\varepsilon} = \langle \sigma \rangle_{\omega} : \langle \varepsilon \rangle_{\omega} = \langle \sigma : \varepsilon \rangle_{\omega} \quad (5.6)$$

As an outcome of the Hill-Mandel condition, the relations for scale transition of stress and strain can be written as

$$\bar{\sigma} = \sum_k f_k \langle \sigma \rangle_{\omega_k} \quad (5.7)$$

and

$$\bar{\varepsilon} = \sum_k f_k \langle \varepsilon \rangle_{\omega_k} \quad (5.8)$$

The average strain and stress fields on the sub-domains can be linked to the average strain and stress on the RVE via fourth order concentration tensors:

$$\langle \varepsilon \rangle_{\omega_k} = \mathbb{A}_k : \langle \varepsilon \rangle_{\omega} \quad (5.9)$$

and

$$\langle \sigma \rangle_{\omega_k} = \mathbb{B}_k : \langle \sigma \rangle_{\omega} \quad (5.10)$$

where $\mathbb{A}_k$ and $\mathbb{B}_k$ are the fourth order strain and stress concentration tensors, respectively. The selection of the concentration tensors depends on the homogenization scheme. In the case of Taylor and Reuss schemes, respectively the strain and stress concentration tensors are equal to unity. In the following section the self-consistent homogenization scheme is described.

5.2.3 Self-consistent homogenization

The self-consistent homogenization scheme was first proposed by Kröner [51] to predict the mechanical response of polycrystals. Generally, in self-consistent schemes every grain is idealized as a spherical inclusion embedded in a homogeneous infinite matrix. The matrix has the effective properties of the polycrystalline material that is felt by each grain. The interaction between the spherical inclusions and the homogeneous matrix was given by Eshelby [21]. The schematic representation of the self-consistent RVE is shown in Figure 5.12.

The local mechanical behavior of each grain is related to the macroscopic behavior via the strain concentration tensor according to

$$\mathbb{A}_i = [\mathbb{S} : (\mathbb{C}^{-1} : \mathbb{C}_i - \mathbb{I}) + \mathbb{I}]^{-1} \quad (5.11)$$

Figure 5.11 Schematic illustration of homogenization procedure.

where $\mathbb{S}$ is the Eshelby tensor and $\mathbb{C}$ and $\mathbb{C}_i$ are the modulus of the matrix and the i^{th} inclusion respectively. The Eshelby tensor depends on the shape of the inclusion and the modulus of the matrix. For spherical inclusions and elastic matrix, the Eshelby tensor is calculated as

$$\mathbb{S} = \frac{1 + \nu}{9(1 - \nu)} \mathbf{I} \otimes \mathbf{I} + \frac{2(4 - 5\nu)}{15(1 - \nu)} \left(\mathbb{I} - \frac{1}{3} \mathbf{I} \otimes \mathbf{I} \right) \quad (5.12)$$

where ν is the Poisson's ratio.

The equations presented so far are based on the assumption that the constituents are elastic and isotropic. Hill [41] introduced an extension to the self-consistent homogenization model to account for the elasto-plastic material behavior. As the material becomes plastic, its modulus becomes nonuniform (i.e. dependent on the state of the strain) and anisotropic. Therefore, in order to evaluate the Eshelby tensor, a linearized reference modulus should be defined [17]. There is no unique definition for the reference modulus. Commonly, the secant or tangent modulus (and their variants) are used as the reference modulus to calculate the Eshelby solution and the strain concentration tensor. A comparison between using different reference moduli is provided in [17]. In this study, isotropic projection of the consistent elasto-plastic tangent modulus is used. Correspondingly the Eshelby tensor is given as

$$\mathbb{S} = \frac{3\kappa}{3\kappa + 4\mu} \mathbf{I} \otimes \mathbf{I} + \frac{6}{5} \frac{\kappa + 2\mu}{3\kappa + 4\mu} \left(\mathbb{I} - \frac{1}{3} \mathbf{I} \otimes \mathbf{I} \right) \quad (5.13)$$

Figure 5.12 The schematic representation of the self-consistent RVE.

where κ and μ are the volumetric and deviatoric coefficients of the isotropic projection of the consistent elasto-plastic tangent modulus. Ultimately, the tangent modulus of the RVE is calculated as

$$\mathbb{C} = \sum_i f_i \mathbb{C}_i : \mathbb{A}_i \quad (5.14)$$

From Equation (5.11) and (5.14) it is evident that obtaining the tangent modulus requires an iterative solution algorithm. The details on the solution algorithm are given in [73].

5.2.4 Yield stress distribution

Most elasto-plastic constitutive models are based on the definition of a yield point which is represented by a yield surface in 3D stress space. The yield point of the material is usually determined experimentally using the 0.2% offset criterion. However, the offset method is based only on a convention and lacks a physical basis [88]. Yet in reality, metals often exhibit a smooth elastic to plastic transition rather than a clear yield point.

As was pointed out in Section 5.1, the plastic deformation in the material is inhomogeneous and partitioned among the grains. The onset of the plastic deformation in each grain is dependent on its orientation with respect to the applied load and its strength. Hence, a yield probability distribution function is introduced in order to consider the relative contribution of yielding at the grain scale to the macroscopic stress response. In this manner the net effect of the microstructural inhomogeneities is condensed into a yield stress distribution. This can be expressed mathematically as

$$\int_0^\infty f(\sigma_y) d\sigma_y = 1 \quad (5.15)$$

Table 5.1 Material parameters used for martensite.

E (GPa)	σ_y (MPa)	C (MPa)	n
210	2592	3918	0.045

where $f(\sigma_y)$ is a probability distribution function which gives the fraction of the material that yields within infinitesimal interval of σ_y and $\sigma_y + d\sigma_y$. The mode of such distribution is the stress at which most of the grains become plastic.

However, for numerical implementation, the model should be restricted to a limited number of fractions leading to a discrete distribution function. Accordingly, the material can be considered as a collection of fractions that yield, each at a certain stress. The discrete yield stress distribution of the fractions can be described as

$$\sum_{i=1}^n F(\sigma_{yi}) \Delta\sigma_{yi} = 1 \quad (5.16)$$

In this work the contribution of microscopic yielding on the macroscopic response of DP600 was modeled using 20 fractions. Every fraction was modeled as an elasto-plastic inclusion using the von Mises yield function in combination with the Ludwik hardening law according to

$$\sigma_f = \sigma_y + C (\varepsilon_{eq}^p)^n \quad (5.17)$$

where σ_y is the yield stress, C is the strength coefficient and n is the hardening exponent.

The mechanical properties of martensite were assigned to one fraction with a volume fraction of 0.14. The material properties of martensite used in this study are listed in Table 5.1.

A truncated normal distribution was assigned to yield stress of the other 19 fractions, resembling the ferritic phase. The lower and the upper bounds of the truncated distribution were chosen as 0 and 1200 MPa respectively. At this point the distribution and hardening parameter of ferrite are unknown. In the next section a procedure for determining the model parameters is presented.

Table 5.2 The distribution and hardening parameters of ferrite.

E (GPa)	σ_{mean} (MPa)	σ_{SD} (MPa)	C (MPa)	n
210	350	400	987	0.91

5.2.5 Parameter identification

An optimization method was used to determine the distribution and Ludwik hardening parameters of the ferritic phase. In such procedure an objective function is defined as the root mean square error (RMSE) between the simulated and the measured stress–strain data. For this purpose the FMINCON function from the optimization toolbox in MATLAB was used.

The procedure typically starts with an initial guess of the values and continues by sequentially optimizing the parameters to minimize the objective function. The Ludwik hardening parameters (C and n), the mean (σ_{mean}) and the standard deviation (σ_{SD}) of the distribution were identified using the tension-compression curve of the material that was pre-deformed at 7.7%.

In order to minimize the number of fitting parameters, the hardening behavior of every fraction is assumed to be the same. However, the stress partitioning due to the microstructural inhomogeneities and the residual stresses are captured via the yield stress distribution parameters. The parameters obtained from the fitting procedure for DP600 are summarized in Table 5.2. The histogram plot of the yield stress distribution of the ferritic phase obtained from the fitting procedure is shown in Figure 5.13.

5.2.6 Prediction of mechanical response

The simulated stress–strain curves using the EPSC material model and the experimental results are shown in Figure 5.14. Overall, there is fairly good agreement between the experimental stress–strain curves and the ones predicted numerically. Noticeably, the Bauschinger effect and permanent softening are captured by the model for various levels of pre-straining.

The model performance in capturing the nonlinear unloading/reloading behavior is shown in Figure 5.15. Evident in Figure 5.15 the model demonstrates the experimentally observed hysteresis loop observed during the unloading–reloading cycles. In Figure 5.16 the predicted stress–strain response of the

Figure 5.13 Histogram plot of the yield stress distribution of ferritic phase.

Figure 5.14 The simulated versus the experimental tension-compression stress-strain curves for DP600-RD.

Figure 5.15 The simulated versus the experimental loading-unloading-reloading curves at different pre-strains for DP600-RD.

material with partial unloading and reloading cycles is demonstrated. Partial unloading–reloading cycles generate smaller inner loops within the fully unloaded–reloaded hysteresis loops. This behavior is in agreement with the experimental observation reported in Chapter 3.

A key achievement of this approach is that the model is shown to be capable of predicting various features of the cyclic stress–strain responses using a simple assumption and few fitting parameters, while the model is efficient enough to be used for modeling an industrial forming process.

It is observed that both CPFEM and EPSC models are under-predicting the magnitude of the nonlinear unloading strain as well as the transient behavior. This can be explained by the fact that neither of the models includes the intragranular stresses developed within the grains. Various mechanisms have been proposed to explain the origin of the intragranular stresses. Long-range internal stresses due to formation of dislocation cells [48, 70], dislocation pile-ups [79, 96] and strain gradient effects [8, 23, 26] are known as the main mechanisms responsible for the intragranular stresses. In general terms, the net effect of the intragranular stresses is referred to as back-stress. The presence of the intragranular back-stresses facilitates the local plastic deformation during unloading and in compression which enhances the nonlinear unloading behavior and Bauschinger effect.

Figure 5.16 The simulated partial unloading/reloading cycles (left) and partial loading/unloading cycles (right).

5.3 Conclusions

Deformation in the microscale is inhomogeneous and causes stress and strain partitioning between grains with different orientations and strengths. This is shown using the full-field CPFEM simulations performed on a dual phase RVE where a large stress and strain partitioning in the grains is observed. The martensitic grains are to a large extent responsible for the stress partitioning. Due to the large contrast in strength of ferrite and martensite, the stress tends to concentrate in the martensitic grains. This explains the distinct Bauschinger effect and the enhanced nonlinear unloading behavior observed in dual phase steels. The statistical analysis of σ_{11} stress distribution showed that a broad stress distribution is present in the material. The stress inhomogeneity increases with increasing deformation.

The CPFEM model used in this study was able to qualitatively capture the nonlinear unloading behavior as well as the Bauschinger effect upon stress reversal. Contrary to popular belief, it is shown that the inhomogeneous stress distribution at the microstructure is, to some extent, responsible for the observed nonlinear unloading and the Bauschinger effect. This is mainly governed by the early re-yielding of the weaker grains in compression during the unloading stage.

A computationally efficient model based on the mean-field elasto-plastic self-consistent (EPSC) approach was developed to predict the cyclic mechanical response of a DP600 dual phase steel. Based on the knowledge obtained from the full-field model, a yield stress distribution was incorporated in the EPSC

model to mimic the stress partitioning in the microstructure. The model successfully captured the nonlinear unloading and Bauschinger effect simultaneously by fitting only one set of material parameters to the experimental data.

Validation

6.1 Introduction

In this chapter the performance of the models proposed in Chapter 4 and Chapter 5 for improving the springback prediction are evaluated. The results are also compared with the classical E-modulus degradation model and the case where the E-modulus is taken as a constant. To this end, the draw-bend experiments described in Chapter 3 were used as a benchmark for evaluating the performance of each model in predicting the springback angle.

A schematic representation of the draw-bending process is illustrated in Figure 6.1. A uniform uniaxial stress develops in the sheet as a result of the applied holding force (F_h). As the sheet is drawn over the roll under the action of the traction force (F_t), a bending stress develops in the sheet causing a non-uniform stress distribution through the thickness of the sheet with the neutral axis shifted towards the center of the roll. At this point the outer fibers (above the neutral axis) are in tension while the inner fibers (below the neutral axis) are compressed. When the sheet exits the roll it unbends, losing its curvature. As a result, compressive stresses develop in the outer fibers and tensile stresses appear in the inner fibers. At the interface between the roll and the sheet a friction force (F_μ) in the opposite direction of the sheet motion arises.

The draw-bend process is a very well suited benchmark to assess the ability of the models to predict springback since the material experiences cyclic loading as it bends and unbends; besides, it covers a wide range of strains using different

Figure 6.1 Schematic representation of the draw-bending process and the stress distribution for different modes of deformation.

roll sizes and holding forces. The draw-bend process is well controlled and therefore is not affected by the scatter of the process parameters that are often seen in industrial forming processes. Additionally, the draw-bend process is not sensitive to the in-plane sheet anisotropy as the stresses in the width direction are small. Therefore, incorporating an anisotropic yield function for simulating the draw-bend process is not critical.

In this chapter, first a description of the finite element (FE) model is given. The incorporated material models and their abilities in capturing the nonlinear unloading/reloading behavior as well as predicting the stresses as the material is subjected to tension–compression cycles are discussed. Then, the effectiveness of each model in predicting the springback profile and angle is investigated. Finally, the time-dependent springback was modeled by incorporating a creep model in the FE model.

6.2 FE model

Considering the width to thickness ratio of the sheets ($w/t = 50$), the draw-bend process is close to the plane-strain condition. However, performing the simulations in 2D using the plane-strain elements results in a significant error in predicting the springback angle, as it cannot capture the anticlastic curvature [95]. Moreover, the stresses developed in the thickness direction (σ_{33}) at the contact region make the shell assumption invalid. Therefore, the draw-bend

process was modeled using 3D solid elements, regardless of the high computation cost associated with it, to minimize any error resulting from the element formulation.

The 3D model was constructed in ABAQUS 6.12 commercial FEM software. Only half of the physical strip was modeled due to the symmetry along the center of the strip. In order to reduce the computation time, a dense mesh was only used on the middle part of the sheet which travels over the roll using quadratic brick elements with reduced integration (C3D20R). The other parts were meshed coarsely using quadratic wedge elements (C3D15). The densely meshed part of the sheet was discretized using 10 elements in width, 250 in length and 5 (2 for the EPSC model) in thickness direction. The roll was modeled as an analytical rigid surface. The draw-bend model assembly and the attached mesh is shown in Figure 6.2.

The contact between the sheet and the roll was modeled using the penalty contact algorithm with the surface to surface discretization method. A constant Coulomb friction coefficient of 0.07 was used for all the simulations. The procedure for determining the friction coefficient and the sensitivity of the springback angle on the friction coefficient is discussed in Section 6.4.

In large deformation problems any tensorial valued state variables must be rotated to account for rigid body motion of the material. The tensorial state variables are rotated incrementally to the current configuration at the beginning of each increment by the DROT matrix given by ABAQUS in UMAT. This matrix represents the increment of rigid body rotation of the basis system.

The process is simulated in four consecutive steps: bending, loading, drawing and springback, in the same way as in the experiments. The four steps of the draw-bend process are shown in Figure 6.3. All the steps were performed in an implicit analysis.

In the first step the sheet is bent over the roll by applying a displacement boundary condition to the left end of the sheet while the right end is fixed (pinned).

Next, the sheet was loaded until the desired holding force (F_h) in the loading step. This was performed by applying a load boundary condition to the left end of the sheet while keeping the right end fixed.

At the drawing step, the sheet was drawn over the roll for 170 mm by applying a displacement boundary condition to the right end of the sheet while keeping the holding force active on the other side of the sheet.

In the last step, the sheet was allowed to spring back freely by deactivating the applied force and disabling the sheet-tool interaction.

Figure 6.2 Draw-bend model assembly.

Figure 6.3 Draw-bend simulation steps.

6.3 Material models

The material model is the key ingredient in the FE model for an accurate springback prediction. In this work four different material models were employed to simulate the draw-bend process: 1. Mixed hardening model; 2. Mixed hardening model with E-modulus degradation; 3. Mixed hardening model in conjunction with the anelastic model and 4. Elasto-plastic self-consistent model (EPSC). In this section the emphasis is placed on two aspects of the material model: first the capability of the model to predict the material behavior as the material deforms and second the ability of the model to predict the unloading behavior upon springback.

The material models are implemented in the FEM code by employing a UMAT (user defined material subroutine) in ABAQUS. In this way, the stress and state variables are updated at integration points by calling the UMAT subroutine. The first three material models rely on a mixed hardening model. In the mixed hardening model the expansion and translation of the yield function are governed by the isotropic and the kinematic hardening laws (see Figure 6.4). The isotropic expansion of the yield function is described using the Ludwik hardening law according to

$$\sigma_f = \sigma_y + C (\varepsilon_{eq}^p)^n \quad (6.1)$$

where σ_y is the yield stress, C is the strength coefficient and n is the hardening exponent.

The translation of the yield function in the stress space is described by a back-stress tensor which evolves according to a kinematic hardening law. In this formulation, the back-stress evolution is given by Armstrong and Frederick [27] according to

$$\dot{\alpha} = \dot{\lambda} \left(h_k \frac{\partial \phi}{\partial \sigma} - h_l \alpha \right) \quad (6.2)$$

where h_k and h_l are the kinematic hardening constants. The mixed hardening model is capable of describing the Bauschinger effect and the transient behavior. For the mixed hardening model, the isotropic von Mises yield function was considered. To determine the parameters of the mixed hardening law, a similar optimization procedure such as the one described in Section 5.2.5 was used. The model was fitted to the tension-compression curve with 7.7% pre-strain. The experimental data and the mixed hardening model prediction are shown in Figure 6.5 (top).

Figure 6.4 Schematic representation of the mixed hardening model prescribing the expansion and transition of the yield surface in the stress space.

The basic mixed hardening model assumes a constant E-modulus to describe the loading/unloading behavior of the material. In this context, the sole usage of the term “mixed hardening” is associated with assuming a constant linear E-modulus.

The mixed hardening model was also incorporated with the classical E-modulus degradation model. In this approach, the nonlinear unloading behavior of the material is approximated by the chord modulus as a function of plastic strain. The reduction of the chord modulus with plastic strain follows an exponential function proposed by Yoshida [101] according to

$$E_{\text{ch}} = E_0 - (E_0 - E_\infty) (1 - \exp(-\gamma \varepsilon_{\text{eq}}^{\text{p}})) \quad (6.3)$$

where E_0 is the modulus of the as-received material, E_∞ is the saturated modulus at large plastic strain and γ is a material constant and determined from the fit. The experimental chord modulus of DP800 (from Chapter 3) along with the fit to Equation (6.3) are plotted in Figure 6.6.

In order to account for the nonlinear unloading/reloading behavior, the anelastic model presented in Chapter 4 was incorporated into the mixed hardening model. The anelastic model is capable of capturing the nonlinear unloading/reloading behavior at various pre-strain levels. The performance of the

Figure 6.5 Experimental and simulated tension–compression curves by mixed hardening model (top) and EPSC model (bottom) for DP800.

Figure 6.6 The experimental chord modulus of DP800 as a function of pre-strain and the fit result of Equation (6.6).

anelastic model in capturing the nonlinear unloading/reloading behavior is presented in Figure 6.7 (top) and compared with the prediction of the E-modulus degradation model and the mixed hardening model with a constant E-modulus.

Finally, the EPSC model was implemented for simulating the draw-bend process. The details on the EPSC model are provided in Chapter 5. In this chapter, the number of material fractions is reduced to ten in order to increase the computational efficiency of the model. The predicted tension–compression stress-strain curves by the EPSC model and the experimental curves are shown in Figure 6.5 (bottom). The same fitting procedure as described in Section 5.2.5 was used to obtain the model parameters for DP800. As is shown in Figure 6.5, the EPSC model with ten material fractions is capable of capturing the main characteristics of the tension–compression experiment and shows a fairly good fit.

As was discussed in Chapter 5, the EPSC model can inherently predict some nonlinear unloading/reloading behavior. However, the model does not give an exact prediction of the unloading behavior since it is not fitted to the unloading experimental data as the anelastic model. The predicted nonlinear unloading/reloading behavior of the EPSC model is compared with the experimental data in Figure 6.7 (bottom).

The model parameters for each model used in this study are summarized in

Figure 6.7 Experimental and simulated unloading/reloading curves by mixed hardening model (top) and EPSC model (bottom) for DP800.

Table 6.1 Summary of model parameters for DP800.

Model		Parameter			
Mixed hardening*	σ_y (MPa)	C (MPa)	n	h_k	h_l
	410	1068.7	0.885	13316.3	55
Modulus degradation	E_0 (GPa)	E_∞ (GPa)	γ		
	187	154	30.5		
Anelastic model	E (GPa)	K (MPa $^{-1}$)	$\varepsilon_0^{\text{an}}$	κ	
	204	5.01×10^{-5}	2.27×10^{-4}	0.68	
EPSC	E (GPa)	σ_{mean} (MPa)	σ_{SD} (MPa)	C (MPa)	n
	210	400	600	1034.4	0.5

* $E = 187$ GPa for the mixed hardening model with constant E-modulus.

Table 6.1.

6.4 Coefficient of friction

An inverse method was used in order to determine the coefficient of friction (COF); it was adjusted such that the resultant reaction force at the pulling side of the sheet matches the one measured in the experiment. To that end, the draw-bend simulation was performed using COFs of 0.05, 0.06, 0.07, 0.08 and 0.09 for a holding force of 13.8 kN and radius of 15 mm. In Figure 6.8, the experimentally measured force and the model prediction for different COF values are plotted leading to an optimal friction coefficient of 0.07.

The normalized springback angles for different COFs are plotted in Figure 6.9 for the roll radius of 15 mm and different holding forces. The springback angles are normalized by the springback angle with COF of 0.07. The influence of the COF on the springback angle was found to be minor for low holding forces. Whereas, it becomes more sensitive to the COF as the holding force increases. By increasing the holding force the contact forces become larger making the contribution of the friction force on the traction force significant. This leads to a larger reduction of the springback angle.

Figure 6.8 A comparison between the experimentally measured traction force for a holding force of 13.8 kN and the ones obtained from the draw-bend simulation using different COFs.

Figure 6.9 The sensitivity of the springback angle on the COF for different holding forces.

6.5 Simulation results

In order to analyze the stresses and strains through the thickness of the sheet, data sampling points on the symmetry plane of the sheet were selected. Figure 6.10 shows the location of the data sampling points on the integration points through the thickness of the sheet. The selected sampling zone enters and then exits the roll during the drawing step. Therefore, it serves as a suitable representative of the deformation in the sheet. The simulations were performed using the “mixed hardening” model with constant E-modulus.

The stress history of the through thickness integration points (IP) during the drawing step is shown in Figure 6.11 ($R = 15$ mm and $F_h = 13.8$ kN). At the beginning of the step the sheet is already pre-tensioned to the applied holding force. As the material enters and exits the roll it experiences tension–compression or compression–tension stress history depending on its location in the thickness of the sheet. Figure 6.11 signifies the importance of considering the Bauschinger effect for the draw-bend simulations.

The distribution of the equivalent plastic strain in the thickness of the sheet is plotted in Figure 6.12. The vertical axis represents the thickness of the sheet; here 0.5 corresponds to the upper surface of the sheet and -0.5 is the bottom surface of the sheet in contact with the roll. For low holding forces, the deformation is concentrated at the outer fibers of the sheet. As the holding force increases, the distribution of the plastic strain throughout the sheet thickness becomes more homogeneous.

As is shown in the right hand side of Figure 6.12, the distribution of the plastic strain is highly dependent on the roll size. For the roll radii of 15 mm and larger the plastic strains mostly remain below 5%. Whereas, with smaller roll radii the magnitude of plastic strains rapidly increases. From Figure 6.12 it is evident that the draw-bending settings used in this study cover a large range of plastic strains in the sheet. The magnitude of the plastic strain in the sheet is important from the modeling point of view as both nonlinear unloading behavior and evolution of the back-stress are functions of plastic deformation.

The distribution of the first stress component σ_{11} through the thickness of the sheet before the springback is displayed in the left hand side of Figure 6.13. As the holding force increases, the neutral axis shifts away from the roll center. This results in a reduction of the bending moment, thereby reducing the springback upon release.

As shown in the right hand side of Figure 6.13 the neutral axis does not shift

Figure 6.10 The location of the data sampling points on the integration points through the thickness of the sheet.

Figure 6.11 The σ_{11} history through the thickness during the drawing step ($R = 15$ mm and $F_h = 13.8$ kN).

Figure 6.12 Through thickness equivalent plastic strain for a fixed roll size of 15 mm and various holding forces (left) and for a fixed holding force of 13.8 kN and various roll sizes (right) at the end of the drawing step.

as the roll size changes; yet the magnitude of the stresses increases as the roll radius decreases, leading to a larger springback.

The profiles of the sheet after springback using different material models are plotted together with the experimental profile in Figure 6.14. The experimental profile was selected randomly from the available results without giving a preference. The predicted springback angle using each model and the one measured experimentally are plotted in Figure 6.15.

From the springback prediction results, it is evident that all the models capture the overall trend of the springback angle with variation of the holding force and roll size. Comparing the absolute angle prediction of each model with the experimental results shows that the mixed hardening model with the constant E-modulus generally underpredicts the springback angle, while incorporating the E-modulus degradation model with the mixed hardening model leads overall to an overprediction of the springback angle. From Figure 6.7 it is evident that the unloading strain is underpredicted by the initial E-modulus; whereas, the E-modulus degradation model overpredicts the unloading strain prior to the fully unloaded state. The experimental unloading strain lies between the prediction of the initial E-modulus and the E-modulus degradation model. Hence, as shown in Figure 6.15, the springback angles predicted by the anelastic model also fall between the prediction of the other two models. The difference between the springback angles predicted by each model shows the significance of the unloading modulus on the springback behavior.

Figure 6.15 demonstrates a comparison between the predicted springback an-

Figure 6.13 Through thickness σ_{11} for a fixed roll size of 15 mm and various holding forces (left) and for a fixed holding force of 13.8 kN and various roll sizes (right) at the end of the drawing step.

gles in absolute terms. However, in order to assess the prediction capability of each model, the relative error should be considered. The relative error is defined as the difference between the predicted value (θ_{model}) and experimental value ($\theta_{\text{experiment}}$) normalized by the experimental value. The relative error percentage (ϵ) is expressed as

$$\epsilon(\%) = \frac{\theta_{\text{model}} - \theta_{\text{experiment}}}{\theta_{\text{experiment}}} \times 100. \quad (6.4)$$

A positive error value means an overprediction in springback angle and vice versa. Figure 6.16 shows the relative error percentage of the angle prediction for each model. For various holding forces, with an exception for 23 kN, the relative error for the anelastic model is in the range of 1-3%. However, in the case of the mixed hardening model with a constant initial E-modulus, this error grows larger as the holding force increases. The angle prediction error using mixed hardening model with E-modulus reduction remains roughly constant at about 7%. It is interesting to note that the highly regarded E-modulus degradation model does not always improve the springback prediction accuracy. This is evident for the holding force of 4.6 and 9.2 kN where the error made by using the E-modulus degradation model is larger than the prediction of the basic mixed hardening model with a constant initial E-modulus.

Lastly, the prediction error by the EPSC model remains under 15% even for the holding force of 23 kN. The mixed hardening based models fail to give an accurate prediction of the springback angle for the holding force of 23 kN.

Figure 6.14 The springback profile from different material models and draw-bend settings and comparison with the experimental results.

Figure 6.15 The predicted and experimental springback angles for a fixed roll size of 15 mm and various holding forces (top) and for a fixed holding force of 13.8 kN and various roll sizes (bottom).

Figure 6.16 The relative error percentage in prediction of the springback angles for a fixed roll size of 15 mm and various holding forces (top) and for a fixed holding force of 13.8 kN and various roll sizes (bottom).

For different roll sizes up to 15 mm, the error in springback prediction by the anelastic model remains well below 5% and increases significantly for the larger roll sizes. It is also seen that even the E-modulus degradation model (which generally overpredicts the springback angle) underpredicts the springback angle for the roll sizes of 20 and 25 mm. As can be noted in Figure 6.5, the mixed hardening model underestimates the compressive stresses at lower pre-strain levels. This leads to an underprediction of the springback angle. Hence the error in stress prediction by the mixed hardening model becomes the limiting factor for the accurate springback angle prediction for larger roll sizes. On the other hand, the EPSC model gives a fairly accurate prediction for the roll sizes of 15 mm and larger.

6.6 Time-dependent springback

The mechanism behind the time-dependent springback is attributed to room temperature creep driven by the residual stresses. As time passes, the internal residual stresses diminish, resulting in a new equilibrium geometry and springback angle. In order to find whether creep is the main mechanism for motivating the time dependent springback, creep experiments were conducted at constant loads on DP800 steel at room temperature (see Section 3.2.2). The experimental data were fitted to a power-law creep according to

$$\dot{\epsilon}_{\text{cr}} = A\sigma^n t^m \quad (6.5)$$

where A , n and m are fitting parameters, t is the total time and σ is the equivalent stress. The experimental creep data and the fit are shown in Figure 6.17. The parameters obtained from the fit are $A = 4.193 \text{ s}^{-1}$, $n = 4.83$ and $m = -0.895$. The creep experimental data obtained at 660, 790 and 860 MPa were used to obtain the fitting parameters.

In order to model the time-dependent springback, after the springback a visco analysis was performed in ABAQUS using the power-law creep and parameters described above. The time-dependent springback analysis was performed for a draw-bend simulation with a holding force of 13.8 kN and a roll size of 15 mm. The mixed hardening model with a constant E-modulus was employed for this analysis. The measured and simulated change of springback angle as a function of time is plotted in Figure 6.18. The simulation result shows a slight overprediction of the time-dependent springback angle. However, the

Figure 6.17 The experimental creep strain rates and the fit results according to Equation (6.5) for DP800.

Figure 6.18 The measured and simulated change of springback angle as a function of time ($R = 15$ mm and $F_h = 13.8$ kN).

trend and the total time-dependent springback angle is in agreement with the experimental data.

Time-dependent springback simulations using a creep model have been performed before [63]. In this work this type of analysis has been repeated to have a comparison between the improvement gained by using a complex modulus and long-term change of angle.

6.7 Conclusions

In this chapter the developed constitutive models were implemented in the implicit ABAQUS commercial FEM software to simulate the springback in the draw-bend process.

It was shown that the mixed hardening model with a constant initial E-modulus underpredicts the springback angle; whereas, incorporating the E-modulus degradation model generally results in an overprediction of the springback angle. Despite what is widely reported in the literature, using the E-modulus degradation model does not always improve the springback prediction. In various cases shown in this work, the error made by overpredicting the springback angle using the E-modulus degradation model was larger than the error caused by using the constant initial E-modulus, although this could be due to other errors in the model.

Capturing the nonlinear unloading/reloading behavior using the anelastic model resulted in an improved springback prediction in most of the cases. Nevertheless, capturing the nonlinear unloading behavior does not guarantee an improvement in springback prediction. For an accurate springback prediction, an accurate hardening model in combination with a description of the nonlinear unloading/reloading behavior is required.

An example of the case where the springback prediction is limited by the hardening model was shown for the large roll sizes (20 and 25 mm). In this case the hardening model intrinsically underestimates the stresses in the material and therefore underpredicts the springback. Here, a better result was obtained using the E-modulus degradation model which usually overestimates the springback angle. Modeling the nonlinear unloading/reloading behavior in this case resulted in a reduction of the predicted springback angle and increased the error in springback prediction.

The EPSC model used in this study did not prove to be the most accurate model, yet it proved to be the most consistent model. In most cases the error made by the EPSC model stayed below 10% and did not show an extreme deviation from the experimental results in any of the simulations. It should be noted that the EPSC model is a predictive type of model with only one additional fitting parameter to the isotropic Ludwik hardening model and has one less fitting parameter than the mixed hardening model. Therefore, it is not directly fitted to the experiments to give an exact description of the material behavior upon unloading/reloading. It can, however, be used as a tool to analyze the material behavior in complicated load cases where it can not be studied experimentally.

Last but not least, the role of time-dependent springback in the accuracy of the springback prediction should not be underestimated. As was shown in Chapter 3, after a period of five months, the time-dependent springback was responsible for approximately 5% change in the springback angle. This is in the same order as the improvement in the springback angle prediction made by considering the nonlinear unloading behavior. The time-dependent springback simulations showed that the room temperature creep can be held responsible for this angle variation.

Conclusions and Recommendations

The aim of this research was to investigate the main mechanisms behind the observed nonlinear unloading behavior and develop constitutive models for improving the springback prediction of the sheet metal forming process.

The dislocation driven anelasticity in Chapter 4 and the inhomogeneous deformation at the microstructure in Chapter 5 were investigated as the potential mechanisms responsible for the observed behavior.

As a result of this study, a better understanding has been gained of the mechanisms that are responsible for the nonlinear unloading behavior. Accordingly, constitutive models were developed to capture the nonlinear unloading/reloading behavior. The performance of the developed models in springback prediction was evaluated using the draw-bend test rig designed and built for this study.

The main conclusions drawn from this work are summarized in the following sections and recommendations for future work are presented.

Experiments

The loading–unloading–reloading experiments outlined in Chapter 3 showed that the magnitude of the anelastic strain in AHSS is sensitive to waiting time and temperature. From these results it was conceived that a dislocation based

mechanism of some sort is responsible for such material behavior. This finding ruled out damage and textural effects as potential mechanisms to the nonlinear unloading behavior and chord modulus reduction.

The experimental results described in Chapter 3 provided useful insights on the mechanism behind the nonlinear unloading behavior. However, the macroscopic experiments do not provide enough evidence to pinpoint the main mechanism. As the two mechanisms operate concurrently, it is difficult to conclude which mechanism is dominating. Hence, each model has to be evaluated and validated at its corresponding length scale. In the next sections recommendations on the experimental evaluation of each model are given.

The draw-bend setup provides an excellent test rig for evaluation and validation of various constitutive models in a well controlled framework. In this work, the draw-bend experiments were performed at a low deformation rate and rotating roll setting. This was done to isolate the strain rate effects and the friction interaction from the results. Nevertheless, the draw-bend setup can be used specifically to study the tribological properties between the sheet and the roll at different contact pressures and temperatures.

At higher holding forces, the effect of the contact condition on the springback angle becomes more critical. The draw-bend setup allows experiments to be performed in a controlled manner to evaluate the effect of new friction constitutive models on the springback prediction.

Dislocation driven anelasticity

According to the theory of dislocation driven anelasticity, the evolution of anelastic strain corresponds to the dislocation density in the material. Therefore, the magnitude of the anelastic strain can be related to the hardening behavior of the material through the Taylor equation. According to the proposed model, a 2nd order polynomial relation between the anelastic strain and the flow stress of the material was established. The model was shown to be capable of predicting the nonlinear unloading/reloading behavior at different levels of pre-strain.

However, from a physical point of view, the dislocation bow-out theory faces some deficiencies in explaining certain experimental observations. For example, the model by itself cannot explain the hysteresis loops in the stress-strain curves upon unloading-reloading cycles. Besides, the relation between the

anelastic behavior and the material behavior in compression up to the yield point in compression (i.e. Bauschinger effect) is not clear. Hence, incorporating a one dimensional uniaxial model in a complex three dimensional forming simulation where the material undergoes a complicated stress path requires additional assumptions.

In order to investigate the dynamics of the dislocation driven anelasticity, discrete dislocation dynamics (DDD) simulations are recommended. In this way the dislocation bow-out/bow-in as a result of an external load can be modeled explicitly. At a higher scale, the DDD model can be incorporated into a crystal plasticity model with multiple grains with different orientations. The results obtained from such simulations can provide a platform for developing a continuum model that can be used for large-scale forming simulations of complex materials.

As for the experimental validation of the dislocation model, the mechanical tests should be performed at the grain scale where the influence of the other factors such as grain to grain interactions are isolated.

Stress inhomogeneity in polycrystalline metals

The alternative theory was established on the inhomogeneous deformation in polycrystalline metals. The grains, depending on their size, orientation, composition and presence of residual stresses, can appear softer or harder than the neighboring grains resulting in stress and strain partitioning at the microscale. The microscale simulations using a crystal plasticity finite element model (CPFEM) in Chapter 5 showed that the stress and the strain are inhomogeneously distributed across the microstructure after some plastic deformation. It was shown that the inhomogeneity, seen as a widening in the distribution profile of the σ_{11} stress component, increases with plastic deformation. Upon unloading, fractions of material already go into compression while the macroscopic stress is still positive.

When the material is further unloaded, a fraction of the material will yield under compression; in the fully unloaded state, approximately 25% of the material has plastically deformed. This microscopic plastic deformation during unloading gives rise to the observed nonlinearity upon unloading. The material fraction that yields during unloading grows larger with increasing pre-deformation before unloading.

Based on the insight obtained from CPFEM, a model based on the elasto-plastic self-consistent (EPSC) homogenization scheme was proposed. In this model, the material inhomogeneity in the material is modeled by considering a distribution in yield stress of material fractions. The EPSC model was shown to capture a number of phenomena simultaneously without requiring additional assumptions while maintaining computational efficiency for FE simulations. However, the EPSC model is a predictive type of model that relies on a small number of fitting parameters which makes the model quantitatively inaccurate in capturing the nonlinear unloading/reloading behavior.

The capability of the CPFEM in predicting nonlinear unloading and the material behavior in compression can be improved by accurate identification of the material parameters. Obtaining material parameters at single crystal level is a challenging task. To this end, inverse methods that combine numerical and experimental procedures are to be developed.

Moreover, strain gradient effects were ignored in the presented CPFEM model. An extra hardening is caused by the geometrically necessary dislocations (GNDs) where a large strain gradient exists. Hence, the stress partitioning between the grains with strain mismatch will be larger. Including the strain gradient effects in the model would result in predicting a larger nonlinear unloading behavior and a more significant Bauschinger effect.

The stress inhomogeneities in the mean-field EPSC model were idealized as a distribution in the yield stress of the material fractions. The model does not incorporate the residual stresses. A methodology for scale transition from the CPFEM model to the EPSC model is required. In this way, the residual stresses inherited from the deformation history and the processing steps can be translated to the EPSC model.

Accurate in-situ digital image correlation (DIC) measurements under a microscope are recommended for measuring the local strains in the microstructure during the loading-unloading cycles. On that basis, the results obtained from the CPFEM model on local plastic deformation during unloading can be verified.

Springback prediction

The proposed models were implemented as user material subroutine (UMAT) for ABAQUS finite element package. The draw-bend experiments were used as a

benchmark to evaluate the performance of the developed models in predicting the springback angle. It was shown in Chapter 6 that using the initial E-modulus generally results in an underprediction of the springback angle, while the E-modulus degradation model usually overpredicts the springback angle. Capturing the experimentally observed nonlinear unloading behavior using the anelastic model proposed in Chapter 4 improves the springback angle prediction considerably. The EPSC model was utilized successfully for springback prediction of the draw-bend process. The EPSC model was shown to predict the springback angle with reasonable accuracy while it uses a smaller number of fitting parameters than the other models used in the study.

The measurements of the draw-bend springback angle, reported in Chapter 3, exhibited a significant variation of the springback angle over time. In fact, the time-dependent springback angle variation is in the same order as the improvement in the springback angle prediction made by considering the nonlinear unloading behavior. Therefore, it is important to take it into account. In Chapter 6, it was demonstrated that by incorporating a power-law creep model, the time-dependent springback can be captured.

The literature on the time-dependent springback is very limited and no study on its significance in the industrial applications is made. More investigations on finding the industrial relevance of the time-dependent springback are recommended. For an improved prediction of the time-dependent springback behavior, a creep model suitable for forming processes, where the history of deformation is taken into account, should be developed.

Nomenclature

Abbreviations

AHSS	advanced high strength steel
BCC	body centered cubic
BCT	body centered tetragonal
CPFEM	crystal plasticity finite element model
DIC	digital image correlation
EBSD	electron backscatter diffraction
EPSC	elasto-plastic self-consistent
FCC	face centered cubic
FFT	fast Fourier transform
IP	integration point
LUR	loading/unloading/reloading
RD	rolling direction
TD	transverse direction
ND	normal direction

Roman symbol

$\mathbb{A}$	strain concentration tensor
$\mathbb{B}$	stress concentration tensor
b	magnitude of the Burgers vector
d	grain size
E	E-modulus
$\mathbf{E}$	elasticity tensor in Voigt notation
$\mathbb{E}$	fourth order elasticity tensor
G	shear modulus
$\mathbf{I}$	second-order unit tensor
$\mathbb{I}$	forth-order unit tensor
$\mathbf{K}$	tangent matrix
k	Hall–Petch coefficient
$\bar{M}$	Taylor factor
$\mathbb{S}$	Eshelby’s tensor

Greek symbol

α	back-stress tensor
α	dislocation strengthening parameter
ρ	dislocation density
ϵ	relative error percentage
ε	small strain tensor

ε_{eq}	equivalent strain
ε_{cr}	creep strain
θ	springback angle
γ	plastic multiplier
μ	coefficient of friction
ν	Poisson's ratio
$\boldsymbol{\sigma}$	Cauchy stress tensor
σ_0	lattice friction stress
σ_{f}	flow stress
σ_{y}	yield stress
τ	resolved shear stress
ϕ	yield function

Operators

$[\cdot]$	components of a tensor in matrix form
$\langle \cdot \rangle_{\omega}$	average over the volume ω
$(\cdot)^{\text{T}}$	transpose
$\mathbf{a} \otimes \mathbf{b}$	dyadic product: $a_i b_j$
$\mathbf{A} : \mathbf{B}$	double tensor contraction: $A_{kl} B_{kl}$
$\dot{x}$	time derivative of x

General subscripts and superscripts

$(.)^e$	elastic part
$(.)^{an}$	anelastic part
$(.)^p$	plastic part
$(.)^{rv}$	recoverable part

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