The Causes of “Shear Fracture” of Dual-Phase Steels

Dissertation

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By

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ABSTRACT

Dual Phase (DP) steels are a class of advanced high strength steel (AHSS) in increasing use for sheet formed automotive parts. In spite of attractive combinations of high strength, high ductility, and low cost, the widespread adoption of DP steels has been limited because practical die tryouts exhibit forming failures far earlier than predicted by standard industrial methods. These failures, often referred to as “shear fractures,” occur in regions of high curvature and with little apparent necking, in contrast to “normal” or tensile fractures. Conventional wisdom attributes shear fractures to a postulated damage mechanism related to the special microstructure of DP steels.

In order to reproduce, characterize, and analyze such fractures in a laboratory setting and to understand their origin of the inability to predict them, a novel draw-bend formability (DBF) test was devised based on displacement control. DP steels from several suppliers with tensile strengths ranging from 590 to 980 MPa were tested over a range of rates and bend ratios (R/t, inner bend radius / sheet thickness). The new DBF test reliably reproduced three kinds of fractures identified as Type I, II, and III, corresponding to tensile fracture, transitional fracture, and shear fracture, respectively. These tests revealed a surprising result: the occurrence of shear fractures increased at higher deformation rates. This degradation of formability was shown to be principally a result of deformation-induced heating, which is greatly accentuated for AHSS because of their
high plastic energy absorption and commensurate high temperature increases, up to 100 degrees C.

In order to understand and quantify the role of deformation-induced heating on plastic localization, temperatures were measured and simulated using a novel new empirical plasticity constitutive form describing the flow stress as a function of strain, strain-rate, and temperature. Designated the “H/V model”, the new constitutive model consists of three multiplicative functions describing (a) strain hardening and its temperature sensitivity, (b) strain-rate sensitivity, and (c) temperature sensitivity. This form allows a natural transition from unbounded strain hardening at low temperatures toward saturation behavior at higher temperatures, consistent with many observations. Thermo-mechanical finite-element simulations using the H/V model confirmed its accuracy and the magnitude of the role on shear fracture. Failure types were predicted, as well as quantitative.

For most of the DP steels tested, heating induced by deformation was identified as the dominant effect in producing unpredicted fractures. This is a result of standard industrial techniques that do not take non-isothermal effect into account, in particular constructing forming limit diagrams from low-speed / isothermal testing, and use of isothermal finite element modeling to analyze industrial sheet forming operations. Microstructural damage can also contribute to shear fracture, but it was a secondary factor for all but one of the alloys tested, in one test direction.
DEDICATION

To my beloved wife, Sang Hee,

and my lovely daughter, Aria (Yubin).
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I would like to express my sincere appreciation to my academic advisor, Dr. Robert H. Wagoner, for his ongoing mentoring, continuous support and encouragement. His teaching and advice assisted and guided me not only throughout the completion of my dissertation but also during the entirety of my studies at The Ohio State University.

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FIELDS OF STUDY

Major Field: Mechanical Engineering

Studies in: Material Characterization, Metal Forming Technology, and Design
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CHAPTER 1: INTRODUCTION

In the field of sheet metal forming, one of the main driving forces is the trends and innovations made in transportation and material industries. Advanced High Strength Steels (AHSS) has been increasingly used for the automotive industry in recent years due to demands for safety and weight reduction as an alternative of conventional steels. However, manufacturing automotive parts with AHSS have imposed many challenges: springback, edge cracking, early fracture, high residual stress, die wear, and high load capacity for press (Horvath and Fekete, 2004; Demeri, 2006). Among them the early fractures have not been any big problems for conventional low carbon steels, but they have been observed often during forming of automotive parts with AHSS. Those fractures and the inability to predict them have been critical barriers for the wide spread of AHSS (Wagoner, 2006; Wu et al., 2006). Therefore, the characterization of these early fractures and understanding the key factor of the inability to predict them are in very high demand by the industry.

1.1. Note about Dissertation Organization

This dissertation is organized around two peer-reviewed papers for which the author had the main responsibility: “A Plastic Constitutive Equation Incorporating Strain, Strain-Rate, and Temperature (Sung et al., 2010b)” (Chapter 2), and “Draw-Bend Fracture of Dual-Phase Steels (Sung et al., 2010a)” (Chapter 3). The first paper was
accepted for publication to the International Journal of Plasticity on Feb. 28, 2010. The second paper is in preparation, to be submitted to the ASME Journal of Engineering Materials and Technology. In order to understand these two contributions of the candidate and this dissertation, it was necessary to summarize two works to which the candidate contributed and collaborated, but for which he did not have primary responsibility. The first of these is summarized in Section 4.1, which is based upon a peer-reviewed paper in preparation: "Finite Element Simulation of Shear Failure of Advanced High Strength Steels (Kim et al., 2010b)"; and the second in Section 4.2, which is based upon a conference paper: "Failure Analysis of Advanced High Strength Steels (AHSS) During Draw Bending (Kim et al., 2009a)". Section 4.1 provides linkages between the constitutive equation introduced and developed in this dissertation (Chapter 2) and the draw-bend tests that comprise the second contribution of this dissertation (Chapter 3). Section 4.2 focuses on the question of microstructural-based damage in AHSS as an alternative mechanism to the effects of deformation-induced heating.

Whatever is not shown in the two papers that comprise Chapters 2 and 3, such as the detail of experimental procedures and the resulting data, is covered in the Appendices. Besides the chapters explained above, Chapter 1 covers the background and the objectives of this study, and some literature reviews. Other literature reviews are contained within the specific chapter related to a particular subject. The conclusions of Chapters 2 and 3 are reiterated in Chapter 5 for the purpose of summary.

1.2. Background
<table>
<thead>
<tr>
<th>Mat.</th>
<th>Cost</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mild Steel</td>
<td>$0.5</td>
</tr>
<tr>
<td>Al</td>
<td>$1.0</td>
</tr>
<tr>
<td>Mg</td>
<td>$1.5</td>
</tr>
<tr>
<td>PMC</td>
<td>$1.2-6.5</td>
</tr>
<tr>
<td>AHSS</td>
<td>$0.55</td>
</tr>
</tbody>
</table>

Table 1.1: Comparison of alternate materials for conventional mild steels (Hall, 2008). 

In today’s increasing interests on the safety and energy, engineers need to react to a wide array of safety regulations by National Highway Traffic Safety Administration (NHTSA), the New Car Assessment Program (NCAP) and Insurance Institute for Highway Safety (IIHS), lowered vehicle emission requirements, and stringent Corporate Average Fuel Economy (CAFE) standards (Horvath and Fekete, 2004). These strengthened regulations and standards have led to an increase in the use of lighter materials in the automobile industry. It is believed that the dramatic increase in fuel efficiency will be achieved by the following proportions: 40% weight reduction, 40% improved power train efficiency and 20% improved aerodynamics (Brown et al., 1995). It is estimated that 1% reduction in car weight saves fuel consumption with 0.6-1% (Hayashi, 1996). Aluminium and magnesium alloys, though attractive from a weight savings perspective, present challenges in terms of high cost, low formability, low weldability and high primary production emission, Table 1.1 (Hall, 2008).
Steel companies have responded to these new requirements by advanced high strength steels (AHSS) which offer impressive combinations of strength and ductility that can reduce the mass and improve the crash worthiness of sheet-formed automotive parts and vehicles (Demeri, 2006; Opbroek, 2009). Ducker Worldwide (Schultz and Abraham, 2009) projected that the use of flat rolled AHSS will increase threefold by 2020 as the result of their replacement of conventional and high strength steels in the North American production of light curb-weight vehicles, while the use of aluminum will stay where it is in 2009, Figure 1.1.

![Trend line only between 2009 and 2020](image)

Figure 1.1: Trend line of material usage (Schultz and Abraham, 2009).
Figure 1.2: Shear failure in AHSS during forming process near tool radius: (a) Shear fracture observed in B-pillar part of DP980, (b) Shear failure observed in Rail part of DP600. (Courtesy of EWI)

However, manufacturing automotive structural components with AHSS poses new forming challenges. One of these challenges is the early fracture (Wagoner, 2006). Such fractures, often dubbed “shear fracture” by industrial practitioners (Walp et al., 2006; Sklad, 2008) and occurring with little or no obvious through-thickness necking, had been observed so often at the radii of forming tools such as draw bead, draw die, and punch nose where the sheet was experiencing bending and unbending under tension during forming with AHSS (Sriram and Urban, 2003; Haung et al., 2008; Kim et al., 2009a) contrary to most traditional low carbon steel alloys that typically failed by stretching only.
(ASM, 1989; Damborg, 1998). Figure 1.2 is examples showing the shear fracture of AHSS near tool radius. The increase of blank holder force and decrease of R/t ratio (tooling radius to sheet thickness ratio) of punch or die in order to reduce springback, which is another problem of fabrication with AHSS, increased the shear fracture in the stamped automotive part with AHSS (Bai and Wierzbicki, 2008).

![Figure 1.2: Shear fracture of AHSS near tool radius.](image)

Figure 1.2: Shear fracture of AHSS near tool radius.

The forming-limit diagram (FLD), generated with a large hemisphere punch and various widths of sheets, is based on the localized necking approach (Keeler and Backofen, 1964; Marciniak and Kuczynski, 1967a; Goodwin, 1968; Keeler, 1969) and has been successfully used to characterize the formability of a material for many years (Burford and Wagoner, 1989; Graf and Hosford, 1990; Bleck et al., 1998; Rees, 2001). However, forming failures in AHSS are not always predictable by the conventional forming tests and application of FLD depending on the application and the grade.

![Figure 1.3: Shear fracture of a front rail with DP780(D).](image)

Figure 1.3: Shear fracture of a front rail with DP780(D) (Stoughton et al., 2006).
(Embury and Duncan, 1981; Sriram and Urban, 2003; Wagoner, 2006; Wu et al., 2006). As shown in Figure 1.3, failures occurred at the die corner radii of the front rail with the DP780(D) sheet, in which FLD with FE simulation did not predict failure properly (Stoughton et al., 2006).

These shear fractures and inability to predict those cause design limitation and manufacturing uncertainty (Walp et al., 2006). Therefore, to capture the key factor of the discrepancy between the predicted and observed failures and to find the fracture criteria are under high demand by forming industry.

1.3. Literature Reviews

Since the objectives of this research are related to the forming of advanced high strength steels and to predict the formability of those, extensive literature reviews on materials, formability tests and prediction of formability were conducted.

![Figure 1.4: Strength versus elongation relationship (Courtesy Auto/Steel Partnership)](Image)
1.3.1. Advanced High Strength Steels (AHSS)

Steel makers and metallurgists have continued to improve the mechanical properties of steels while maintaining the inverse relationship between ductility and strength, Figure 1.4 (Wagoner, 2006).

Traditional steel grades used in the automotive industry focus on ferrite and carbide-based metallurgy such as mild steel and HSLA. The 1\textsuperscript{st} generation AHSS, composed of ferrite matrix and secondary structures of martensite and retained austenite, has been replacing the conventional steels due to their excellent combination of high strength and good ductility. The fundamental metallurgy for 1\textsuperscript{st} generation AHSS is well established, but the several issues relating with application of AHSS to automotive parts still remain as barriers to the widespread adoption of these materials (Horvath and Fekete, 2004). The 2\textsuperscript{nd} generation AHSS are being developed based on an austenite matrix, but the high cost of alloying elements such as manganese (Mn) and nickel (Ni) becomes an obstacle for widespread adoption of these grades (Wagoner, 2006). The 2\textsuperscript{nd} generation AHSS includes Twinning Induced Plasticity (TWIP) steel and L-IP, Figure 1.4. Recently, metallurgists have begun to develop the 3\textsuperscript{rd} generation AHSS aiming to achieve material property combinations intermediate between 1\textsuperscript{st} and 2\textsuperscript{nd} generation AHSS, Figure 1.4, with reasonable cost using the knowledge acquired from development of 1\textsuperscript{st} and 2\textsuperscript{nd} generation AHSS. The development of 3\textsuperscript{rd} generation AHSS was triggered by “Advanced High Strength Steel Workshop” supported by Auto/Steel Partnership (A/SP), National Science Foundation (NSF), and Department of Energy (DOE) in October 2006 (Wagoner, 2006).
The 1st generation AHSS, which are the main materials for this study, can be defined as various ways depending on the organization, and generally defined as the HSS with some unique characteristics such as good formability or high strength above other AHSS (Opbroek, 2009). Thus, AHSS can be categorized in three distinct families (Mould, 2002; Opbroek, 2009): 1) DP and TRIP steels characterized by good formability; 2) CP (Complex Phase) and MS(martensite) steels characterized by very high strengths, and 3) PFSS (Pre-Form Strengthened Steel), HF (Hot formed boron steels), and PFHT (Post-Forming Heat Treatable) steels which are strengthened by rapid cooling after a forming operation. Among these 1st generation AHSS, DP and TRIP steels have been being intensively studied by the automotive industry and its suppliers. The attractive combination of formability and strength results from the control of microstructure, and these complex microstructures require the application of a more complex cooling cycle, in addition to more alloying elements. For DP and TRIP steels, a three-step cooling
pattern is normally utilized in general instead of a simple one-step cooling pattern used for manufacturing conventional steels, Figure 1.5 (Kang and Kwon, 2002; Kim et al., 2004).

**Dual-Phase (DP) steels**

DP steels consist of a ferrite matrix and islands of martensite, Figure 1.5. The main alloying elements for DP steels are carbon(C) and manganese(Mn), and additionally chromium(Cr), molybdenum(Mo), vanadium(V), and nickel(Ni) are added for the formation of martensite at practical cooling rates; i.e. those alloying elements help to increase the hardenability of the steel, which helps to avoid transformation of austenite to pearlite during cooling process (Kang and Kwon, 2002; Kim et al., 2004).

Since Rashid (Rashid, 1976) reported first on the advantages of DP steels, a lot of researches have been carried out in order to investigate various characteristics of DP steels. Now it is well known that the microstructure of DP steels experiences three stages of deformation during cold working (Byun and Kim, 1993; Bag et al., 1999): 1) both ferrite matrix and martensite particles deform elastically; 2) ductile ferrite phase deforms plastically first while the martensite phase continues to deform elastically; and 3) both ferrite and martensite phases deform plastically. Although this deformation happens microstructurally, but also affects to the macroscopic behavior. In stage 2, when the ferrite phase deforms, there is stress concentration in the boundary areas of ferrite contacting the hard phase, martensite, which stimulates more plastic deformation of ferrite in this region, resulting in high strain hardening exponent (n-value) at low strain; and then the n-value decreases as both ferrite and martensite phases deforms plastically in stage 3 as shown in
Figure 1.6. These imply that the macroscopic behavior already reflects the microscopic behaviors.

The inhomogeneous strain distributions between ferrite and martensite grains have been observed using a scanning electron microscope (SEM) during a tensile test by Shen et al. (Shen et al., 1986). This inhomogeneity can lead to the fracture of DP steels (Sun et al., 2009c). The grade of DP steel can be controlled by the volume fraction of martensite (Jiang et al., 1993). Bag et al. (Bag et al., 1999) showed that increase in strength with volume fraction only extends up to 55%, after which a reduction in strength is observed. Byun and Kim (Byun and Kim, 1993) stated the same conclusion, but with a different value of volume fraction.

Figure 1.6: Material characterization of HSLA, DP, and TRIP: (a) variations of $n$ value with strain, (b) flow stress for various AHSS (Konieczny, 2003)
Transformation-Induced Plasticity (TRIP) steels

TRIP steels consist of a primary ferrite matrix, embedded retained austenite, and hard particles such as bainite and martensite, Figure 1.5. TRIP steels usually use higher quantities of carbon (C) than DP steels in order to stabilize the retained austenite phase to below ambient temperature. Mn is added to promote hardenability and partitions between austenite and ferrite during interstitial annealing. Also silicon (Si) and aluminium (Al) are used to accelerate the ferrite-bainite formation and to suppress the carbide precipitation during bainite transformation (Kang and Kwon, 2002; Kim et al., 2004; Hofmann et al., 2006; Wang et al., 2006).

![Figure 1.7: Transformation Induced Plasticity effect: (a) TRIP effect with deformation (Cooman, 2004), (b) volume fraction of retained austenite with strain (Lee et al., 2004a).](image)

In TRIP steels, the plastic deformation induces transformation of retained austenite into martensite as shown in Figure 1.7. Figure 1.7(b) shows how the volume fraction of retained austenite changes with strain increase. This transformation and the homogeneous
distribution of the different phases lead to high strain hardening and therefore to excellent
elongation and strength (Bleck, 2002). Yokoi et al. (Yokoi et al., 1996) explained the
phase transformation in three steps and this was also reflected to macroscopic behavior:
1) at a critical tensile stress, retained austenite transforms irreversibly to martensite in
strain-concentration areas; 2) strain-induced martensitic transformation is accompanied
by a volume expansion of the transforming region, which leads to additional plastic
accommodation and work hardening of the surrounding microstructure, and 3) this phase
transformation would result in a delay of macroscopic necking and ultimately leads to
higher uniform and total elongations as shown in Figure 1.6 (a).

The stability of retained austenite against strain-induced martensite transformation
and the volume fraction of austenite are major factors governing the forming behavior of
TRIP steels (Lee et al., 2004a; Wang et al., 2006). The transformation of austenite to
martensite begins when the stress exceeds the transformation barrier which varies due to
the stability and volume fraction of austenite. Kulp et al. (Kulp et al., 2002) reported this
barrier was around 200MPa. The transformation is also influenced by temperature and
strain rate (Iwamoto and Tomita, 2001; Wei and Fu, 2002). Savrai and Pychmintsev
(Savrai and Pychmintsev, 2002) showed that the retained austenite was less stable under
tensile loading than under compressive loading. Therefore, the stabilization of austenite
has to be well balanced in order to prevent the formation of martensite during cooling and
allow the continuous martensite formation over a large strain range (Zaefferer et al.,
2004).

Other AHSS
CP steel has many of the same alloy elements found in DP and TRIP steels, but also has the micro-alloying elements that form fine strengthening precipitates such as Niobium (Nb), Titanium (Ti), or Vanadium (V), resulting in very fine ferrite matrix containing higher volume fraction of hard phases. CP steel has been used for the parts requiring the high energy absorption and high residual deformation capacity such as bumpers and B-pillar reinforcements (Werle and Dahlke, 2004; Opbroek, 2009).

MS steel is characterized with smaller amount of ferrite in a large matrix of martensite. MS steel has maximum ultimate tensile strength to 1700 MPa due to the large volume fraction of martensite, but low formability. The MS steel, therefore, can be used for the parts that require high strength and good fatigue resistance, with relatively simple cross sections like door intrusion beams, bumper reinforcement beams, and side reinforcements. Based on the targeted strength level, several micro-alloying elements like manganese (Mn), silicon (Si), Chromium (Cr), boron (B) and Molybdenum (Mo) are added to increase hardenability in various combinations. Also carbon(C) is added to increase the hardenability and strengthening the martensite (Mould, 2002; Demeri, 2006).

TWIP steel (Twinning Induced Plasticity) is a new type of steel consisting of fully austenite (FCC) phase in room temperature by adding 15–25 weight% Manganese(Mn) and Nickel(Ni), silicon(Di) and aluminum(Al). This steel shows extensive twining under mechanical load in grain boundary as can be seen in Figure 1.8. The twinning causes a high value of the work hardening behavior because the resultant twin boundaries act like grain boundaries and strengthen the steel. Therefore, TWIP steel exhibits high strength, excellent plastic elongation, and an ideal uniform work hardening behavior. These
characteristics are ideal for automotive steel in terms of high-energy absorption and weight reduction, but this steel has been applied rarely to real parts because it is too expensive due to the more alloying elements (Cugy et al., 2006b; Opbroek, 2009).

![Optical Microscopic image of TWIP steel before deformation](image1)

![Optical microscopic image of TWIP steel after deformation, e= 34%](image2)

Figure 1.8: The microstructure of Twinning Induced Plasticity (TWIP) Steel (Allain et al., 2004).

Steels used in hot forming process to produce parts of high strength in the range of martensite steel with no springback are called Hot Forming (HF) steels. High strength steels with higher amount of manganese (Mn) and boron (B) called boron steels are commonly used for hot stamping process. The hot forming process consists of the following steps (Altan, 2007): 1) austenizing by preheating to 900–950°C which provides improved formability of the sheet material; 2) forming to desired complex part at high velocity to avoid temperature loss; 3) cooling in the die at a rate of 50 to 100°C/s (water cooling) to form the martensitic structure from austenite. Figure 1.9 shows the microstructure, formability and strength of the steel at various stages during the hot stamping process. Hot forming Boron steel is most commonly used for crash-resistant
parts such as bumpers and pillars with high strength, UTS to 1500 MPa (Hein et al., 2006).

Figure 1.9: The schematic of Hot Forming Process (Altan, 2007).

1.3.2. Formability of AHSS

Figure 1.10: Combination of types of forming: B-bending, BS-biaxial stretching, D-drawing, P-Plane-strain stretching, U-unbebing (Keeler, 1970)

Formability of materials is hard to be characterized by one simulative test because stamping process consists of various forming modes which react to a different set of mechanical properties (Sriram and Urban, 2003). These complex forming operation
consists of combinations of three major deformation components: drawing, stretching and bending (Demeri, 1981; Miles, 2006). The relative amount of each deformation component varies from part to part and from location to location in the same part as shown in Figure 1.10. Therefore the formability needs to be explained based on different basic forming abilities such as drawability, stretchability, bendability and combinations of these.

**Drawability**

![Figure 1.11: Drawability of various steels (Cugy et al., 2006a)](image)

Drawability is generally characterized by maximum drawable height or Limiting Draw Ratio (LDR), the ratio of the largest height that can be drawn to a circular cup with certain punch diameter. Cugy et al. (Cugy et al., 2006a) used a special shape punch and compared the drawable heights of conventional, DP, TRIP, and X-IP steels, Figure 1.11. DP 590 attained a higher height than HSLA320, and TRIP 800 showed similar
drawability with DP 590 even though the strength was much higher than that of DP 590. X-IP, one of 2\textsuperscript{nd} generation AHSS, showed the best drawability among tested materials. Konieczny (2003) also presented the same results with cup drawing tests. Dykeman(2009) compared three DP780 steels showing different mechanical properties, and reported DP780 with larger total elongation had better drawability.

\textbf{Stretchability}

Stretchability is the ability of the material to be stretched without failure. Hydraulic bulge test and Limiting Dome Height (LDH) test are commonly used tests to evaluate the stretchability of the material (Miles, 2006). In hydraulic bulge testing, the height of the bulge/dome before burst and the corresponding strain indicates the stretchability of the material. The LDH test is similar to hydraulic bulge test where a metallic hemispherical punch replaces the hydraulic medium. It should be noted that friction between the sheet and the punch has an effect on the test measurements.

Figure 1.12 (a) compared formable heights of similar strength grades of TRIP and HSLA steels depending on width of specimen. The TRIP had better stretchability than HSLA due to higher and more uniform strain hardening behaviour in the dome area (Takahashi, 2003). A Forming Limit Diagram (FLD) was constructed by LDH tests with various widths of specimens, which specify the limits in terms of strain over a wide range of strain components and strain paths. It could be observed that the formability limits for TRIP steel was very similar to DDS steel due to high strain hardening and maintaining higher strain hardening over large strains, Figure 1.12(b). Because DP steels was characterized as decrease of strain hardening with increase of strain, DP steels resulted in lower stretching limit over the entire strain range compared to the TRIP steels and mild
steels, Figure 1.12(b) (Konieczny, 2003). Sriram et al. (2003) showed the FLD based on the Keeler-Breizer (K-B) equation matched well with experimental measurements but was more conservative in predicting the limits for both DP steels and TRIP steels.

Figure 1.12: Stretchability: (a) LDH test (Takahashi, 2003), (b) FLD for various steels (Konieczny, 2003).

**Bendability**

Bendability is the minimum bending radius attainable by a given material. During bending, the outer fiber of the material is subjected to tensile stress while the inner fiber is subjected to compressive stress. Therefore bending causes little thinning unlike drawing and stretching (Demeri, 1981). The sheet material during bending begins to fracture when the maximum tensile stress at the outer fiber exceeds a critical value, which is generally much higher than total elongation in uniaxial tensile test. The maximum strain in the fiber depends on the bend radius, bending angle and the sheet thickness.
Advanced high strength steel application guideline (Opbroek, 2009) reported that bendability was inversely proportional to strength among DP steels using three point bending test; Konieczny (2003) showed DP steel had better bendability than conventional high strength steel with similar strength, Figure 1.13.

![Figure 1.13: Bendability of DP600 and HSLA550 (Konieczny, 2003).](image)

**Bending under Tension**

It should be noted that forming limit diagram (FLD) was widely used for evaluating the formability of conventional steels because failure often occurs in non-contacting regions where bending is minimal and stretching dominates, but advanced high strength steels often fail by shear fracture where bending dominates as stated in the section 1.1. Recently Edison Welding Institute (EWI) analyzed fractured parts of AHSS provided by A/SP team, and found that fractures occurred as bending under tension mode for 12 parts out of 15 fractured parts (Wagoner et al., 2009e). Figure 1.3 shows that FLD cannot predict the fracture at the corner of A pillar of DP780, where the part encountered both bending and stretching. Therefore it is very important to evaluate the formability of AHSS under bending and stretching.
(a) MDS test (Walp et al., 2006)

(b) HSBT (ASBT) test

(c) SFS test (Shih et al., 2009)

(d) BUT test

Figure 1.14: Tests under bending and stretching
Many test methods have been developed to evaluate the formability of bending under stretching: a modified Duncan-Shabel (MDS) apparatus, stretch-bend test (HSB test: hemispherical stretch-test, ASB test- angular stretch-bend test), stretch forming simulator (SFS), and bending under tension test (BUT), Figure 1.14 (Demeri, 1981; Damborg et al., 1997; Sriram and Urban, 2003; Walp et al., 2006; Hudgins et al., 2007; Shih and Shi, 2008; Shih et al., 2009). Usually among formability tests, some tests have been more widely accepted than others based on the requirement that needed to be satisfied, existing testing capability, and tooling cost, but these were no ASTM or SAE standards for this test. For testing formability of bending under tension, the ASB test (or HSB test) has been used widely because it can be installed to the general press by manufacturing the tooling set, while other tests need the own system which yields a lot of capital investment.

The measure of bending under tension limit in the ASB test is represented by the punch stroke (height) at which failure occurs in the test for different radius/thickness (R/t) ratios. Sriram and Urban (2003) compared the formability of several steel alloys using ASB test. Among the tested AHSS, TRIP 600 steel showed great formability for all the R/t ratios, while DP steels ranked poor formability, Figure 1.15(a). The authors also showed the critical R/t is increasing with the increase of ultimate tensile strength of the material as shown in Figure 1.15(b). However, the ASB test has little draw component as found at the die radius in many industrial forming operations.
Walp et al. (2006) reported that the critical R/t values of several AHSS based on the wall stress change with MDS test: DP800(8) > CP800(5) > DP600(4) > TRIP800(3). Damborg et al. (1997) developed the BUT test, which was originally used for measuring friction, which mimicked the mechanics of deformation of sheet metal as it was drawn, stretched, bent and straightened over a die radius entering a typical die cavity. Thus the
test represents an exact sheet forming operation over a die with the capability of careful control and measurement of sheet tension force, which the other in-plane tests and stretch-dominated test cannot reproduce. Damborg et al. (1997) could detect both normal plastic localization/necking and shear fracture and Hudgins et al. (2009) reported critical R/t of several AHSS using the BUT test. Shih (2008, 2009) reproduced shear fractures using both SFS tests and BUT tests, and concluded that fracture limit curves generated using both tests were correlated very well (Shih and Shi, 2008; Shih et al., 2009).

Edge Split

Figure 1.16: Comparison of various forming characteristics for three DP steels and one TRIP steel. (Dykeman, 2009)

Another important failure mode for AHSS is edge-defect-initiated failure which had a minor effect on conventional steels. The study of this mode is important because the excess metal is usually trimmed off and then post forming operations are performed to form the sheet metal to the final part. Sriram et al. (2003) reported that edge stretch limits
of AHSS, such as TRIP, DP, RA, MS steels, were about half of those of mild steel and HSS using hole flanging experiments. TRIP steel showed similar or less hole expansion ratio than DP steels. Dykeman (2009) summarized the bendability, stretchability and edge stretch limits of DP and TRIP steels in one graph, Figure 1.16, and showed that bendability and hole stretch limits was less related to total elongation.

![Graph showing hole expansion test results](image)

Figure 1.17: Effect of piercing punch conditions on the hole expansion test results (Courtesy SSAB).

It should also be pointed out that the edge stretch limit depends also on the trimming method and tooling quality. Figure 1.17 shows how the status of the punch used for piercing influences the hole expansion result.
1.3.3. Failure Prediction of AHSS

In addition to the wide ranges of experiment data on AHSS formability, reliable failure predictions are also under high demand from the forming community. On the analytical perspective, Bai and Wierzbicki (Bai and Wierzbicki, 2008) derived a close-form solution for both global force responses and local strain and stress state using a stretch-bending test under assumption of plane strain. This was extended by Issa (Issa, 2009) by incorporating power hardening law and considering loading history. Recently, it was proved by analytically that the maximum tensile force that a sheet can take during plane-strain bending-under-tension could be significantly reduced at small R/t (Hudgins et al., 2010; Kim et al., 2010a).

On the numerical perspective, the micromechanical finite element models have been widely used in order to understand the local mechanics and mechanisms governing the macroscopic behaviour of heterogeneous materials including DP steels (Al-Abbasi and Nemes, 2003a, b, 2007, 2008; Choi et al., 2009b, a; Sun et al., 2009c; Uthaisangsuk et al., 2009) because the attractive macroscopic behaviours of AHSS are attributable to their microstructures.

One of the challenges to micromechanical modeling is realistic description of complex geometry. Idealized microstructure-based models have been favored in analytical approaches because of its simplicity instead of introducing realistic microstructural features of complex multi-phase materials, volume fractions and average orientations of the microstructural features are used to construct the geometry of models usually with spheres or ellipsoids embedded in a matrix (McClintock, 1968; Rice and Tracey, 1969; Al-Abbasi and Nemes, 2003a, 2007).
However, it is not easy to ensure that the simplified models represent the corresponding real materials since more detailed information such as particle morphology and clustering is missing. This limits the applicability of the idealized microstructure-based micromechanical models. Recently, some of research groups use the realistic microstructure-based models based on microscopic images of materials, Figure 1.18, or statistically generated microstructures (Prahl et al., 2007; Zeghadi et al., 2007; St-Pierre et al., 2008; Uthaisangsuk et al., 2009).

The above micromechanical modeling can predict the deformation behavior of AHSS, but not fracture. Because it has been reported that the fracture of AHSS (especially dual phase steels) is in a ductile manner; i.e. void nucleation, growth, and coalescence (McClintock, 1968) as shown in Figure 1.19, many micromechanical models consider conventional ductile fracture as the mechanism for fracture in AHSS (Horstemeyer et al., 2000; Lee et al., 2004c; McVeigh et al., 2007; McVeigh and Liu, 2008a; Sun et al., 2009b). While some of ductile fractures considered the decohesion of an embedded particle from the surrounding matrix or particle fracture as a damage nucleation mechanism (Dierickx et al., 1996), many ductile fracture assumed that void nucleation process was already completed and fracture was predicted as the growth and coalescence of these micro-voids (Schacht et al., 2003). For example of the latter, Gurson like model (Gurson, 1975; Tvergaard, 1982; Tvergaard and Needleman, 1984) introduced the initial micro-scale imperfection and define the empirical criterion of void coalescence. In this regard, a large volume of the ductile fracture studies focus on the relationship between plastic distortion and micro void evolution.
Choi et al. (Choi et al., 2009a, b) and Sun et al. (Sun et al., 2009c) reported that the microstructure level inhomogeneous strain distribution during deformation may be the key factor influencing ductility of AHSS, not void growth and coalescence. They investigated on both DP steels and TRIP steels based on the real microstructures with
various material properties of martensite, which is influencing strongly to the mechanical properties of AHSS (Kim, 1988).

![Image of fracture during tensile testing and detail of mechanisms at the phenomenological level.]

Figure 1.19: a) schematic of fracture during tensile testing due to micro-void nucleation, growth, and coalescence, b) detail of mechanisms at the phenomenological level (C.Ruggieri, 2004).

From an industry point of view, however, the increasing demand for numerical simulations of sheet metal forming processes and vehicle collisions calls for an accurate fracture model which can be easily calibrated from physical tests and efficiently implemented into FE codes. Apparently, most of the above mentioned micromechanical models and fracture models do not meet this requirement. Kim et al.(2009) also presented their results of FE simulation using the real microstructure of DP590, and concluded that those approaches were not computational cost effective.

1.4. Objectives

The discrepancy between predicted and observed fractures during forming of AHSS sheet has been the major factor limiting the wide use of AHSS in automotive industry (Wagoner, 2006; Walp et al., 2006; Wu et al., 2006). This inability to predict accurately the shear fracture can be an important cost factor and increases the lead times relating
with the die design and draw bead designs. Many research groups are investing a lot of efforts to find the alternative approach to predict the shear fractures experimentally and numerically. As shown in the literature reviews, simple stretching, drawing, and bending tests cannot reproduce such fractures, and microstructure based FE model cannot solve this problem in industry.

Therefore the overall objective of this research is to explore key parameters which cause the inaccuracy of finite element prediction using a laboratory test which can reproduce shear fracture under bending and tension condition and to develop fracture criteria which can improve the simulation accuracy. To this end, the specific objectives of the proposed research are to:

- Develop a reliable laboratory test using draw-bend system which can reproduce two fracture types consistently: 1) plastic localized failure and 2) shear failure.
- Characterize the shear fracture using the developed test in order to find the origin affecting the low predictability of shear fracture of DP steels.
- Develop a fracture map of selected DP steels as a function of R/t, restraining force or bending strain, which can be used as a guideline for evaluating the relative formability between competing materials and for detecting processing problems that lead to unsuitable microstructures.
- Develop fracture criteria which can represent the characteristics of DP steels.

For this study, several grades of DP steels, with tensile strength between 590 and 980 MPa, will be mainly used and one grade of TRIP steel will be used for the comparison.
CHAPTER 2:

A PLASTIC COBSTITUTIVE EQUATION INCORPORATING STRAIN, STRAIN-RATE, AND TEMPERATURE FOR DP STEELS

NOTE: The remainder of Chapter 2 (following this note) is a manuscript accepted for the publication to International Journal of Plasticity on Feb. 28, 2010. It represents one of two major contributions described in this dissertation, namely development of a new constitutive equation for AHSS. The only differences between the Feb. 12, 2010 manuscript and this chapter are organizational ones. That is, the figure numbers and section numbers and so on have been changed to fit into the dissertation format. Additionally, the designations of materials are a bit different with those in other chapters, Table 2.1.

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<tr>
<th>Chapter 2</th>
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<tr>
<td>DP590</td>
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<td>DP980</td>
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Table 2.1: Designation difference between Chapter 2 and other Chapters.

Additional detail deemed appropriate for a dissertation, but not a journal paper are added in Appendix C to I:

Appendix C (temperature increase of DP steels during deformation): For the comparison of temperature increase between one of conventional steel and several grades
of DP and TRIP steels, temperature increase was calculated in tensile tests under assumption of adiabatic condition and simulated using a coupled thermo-mechanical FE model of DBF test.

Appendix D (microstructure of selected AHSS): The microstructures of three grades of DP steels and a grade of TRIP steel were shown. Additionally, the microstructures of DP980(A) and DP980(D) were compared.

Appendix E (tensile test at elevated temperature): The details of the experiments at elevated temperature and correction procedure were introduced.

Appendix F (analysis of hydraulic biaxial bulge test): The details of the hydraulic biaxial bulge test and correction procedure for anisotropy were explained.

Appendix G (mesh of FE model of tensile test): The effects of mesh size on FE results of a tensile test were shown.

Appendix H (determination of thermal coefficients): Several thermal coefficients were found with JMatPro using the composition of DP590(B) steel sheet.

Appendix I (comparison of experiment and FE simulation in isothermal tensile tests): Comparisons of isothermal tensile tests between experiment and FE simulations were shown for three DP steels: DP590(B), DP780(D), and DP980(D).

2.1. Abstract

An empirical plasticity constitutive form describing the flow stress as a function of strain, strain-rate, and temperature has been developed, fit to data for three dual-phase (DP) steels, and compared with independent experiments outside of the fit domain. Dubbed the “H/V model” (for “Hollomon / Voce”), the function consists of three
multiplicative functions describing (a) strain hardening and temperature sensitivity, (b) strain-rate sensitivity, and (c) temperature sensitivity. Neither the multiplicative structure nor the choice of functions (b) or (c) is novel. The strain hardening function, (a), has two novel features: 1) it incorporates a linear combination coefficient, α, that allows representation of Hollomon (power law) behavior (α =1), Voce (saturation) behavior (α =0) or any intermediate case (0<α<1), and 2) it allows incorporation of the temperature sensitivity of strain hardening rate in a natural way by allowing α to vary with temperature (in the simplest case, linearly). This form therefore allows a natural transition from unbounded strain hardening at low temperatures toward saturation behavior at higher temperatures, consistent with many observations. Hollomon, Voce, H/V models and others selected as representative from the literature were fit for DP590(B), DP780(D), and DP980(D) steels by least-squares using a series of tensile tests up to the uniform strain conducted over a range of temperatures. Jump-rate tests were used to probe strain rate sensitivity. The selected laws were then used with coupled thermo-mechanical finite element (FE) modeling to predict behavior for tests outside the fit range: non-isothermal tensile tests beyond the uniform strain at room temperatures, isothermal tensile tests beyond the uniform strain at several temperatures and hydraulic bulge tests at room temperature. The agreement was best for the H/V model, which captured strain hardening at high strain accurately as well as the variation of strain hardening with temperature. The agreement of FE predictions up to the tensile failure strain illustrates the critical role of deformation-induced heating in high-strength / high ductility alloys, the importance of having a constitutive model that is accurate at large strains, and the implication that damage and void growth are unlikely to be determinant factors in the tensile failure of
these alloys. The new constitutive model may have application for a wide range of alloys beyond DP steels, and it may be extended to larger strain rate and temperature ranges using alternate forms of strain rate sensitivity and thermal softening appearing in the literature.

2.2. Background

Advanced high strength steels (AHSS) provide remarkable combinations of strength and ductility by careful control of microstructure of ferrite, martensite, and retained austenite components. Dual-phase (DP) steel microstructures consist of large islands (of approximately the grain size) of hard martensite embedded in a softer ferrite matrix, thus mimicking a typical discontinuous composite structure. DP steels are well-established but their widespread adoption has been limited because die tryouts show that forming failures can occur much earlier than predicted using standard forming limit diagrams (FLD) and commercial finite element (FE) programs. Unlike failures observed with traditional steels, unexpected DP steel failures occur in long channel regions of low R/t (bending radius / sheet thickness) as the sheet is drawn over the die radius while being stretched, bent and straightened. Such failures are often dubbed “shear fracture” by industrial practitioners (Huang et al., 2008; Sklad, 2008; Chen et al., 2009). Conventional wisdom has attributed this phenomenon to a special damage / void growth mechanism, possibly related to the large, hard martensite islands (Takuda et al., 1999; Horstemeyer et al., 2000; Lee et al., 2004b; Sarwar et al., 2006; Wagoner, 2006; Vernerey et al., 2007; McVeigh and Liu, 2008b; Xue, 2008; Sun et al., 2009a).
As explained in the section 1, widespread adoption of AHSS has been limited because die tryouts show that forming failures can occur much earlier than predicted using standard forming limit diagrams (FLD) and commercial finite element (FE) programs. Conventional wisdom has attributed this phenomenon to a special damage / void growth mechanism (i.e. ductile fracture), possibly related to the large, hard martensite islands (Takuda et al., 1999; Horstemeyer et al., 2000; Lee et al., 2004b; Sarwar et al., 2006; Wagoner, 2006; Vernerey et al., 2007; McVeigh and Liu, 2008b; Xue, 2008; Sun et al., 2009a).

Recent work was conducted by the authors to determine whether such unpredicted forming failures are a result of typical plastic localization or a special kind of fracture. Draw-bend tests which reproduce the conditions that promote “shear failure” were found to exhibit greatly varying formability depending on the strain rate in the test (Wagoner et al., 2009b; Wagoner et al., 2009d). Furthermore, temperatures of up to 100 deg. C in sheet regions away from the necking area were measured, much higher than with typical traditional steels with lower strength / ductility combinations. It became apparent that the role of temperature in the flow stress, almost universally ignored in standard sheet forming simulations and mechanical property determinations, might be a critical factor.

In order to consider these effects quantitatively, a reliable constitutive equation was needed in the range of strain, strain rate, and temperature encountered in such tests and forming operations.
2.2.1. Role of Deformation-Induced Heating in Plastic Localization

Unlike the uniformly elevated temperature imposed in a warm forming process (usually applied to increase the inherent ductility of the alloy), the heat caused by plastic deformation in a local region of high strain has a detrimental effect on mechanical formability because most alloys soften with increasing temperature, and the softening occurs at the eventual fracture location (Kleemola and Ranta-Eskola, 1979; Ayres, 1985; Lin and Wagoner, 1986; Gao and Wagoner, 1987; Lin and Wagoner, 1987; Wagoner et al., 1990; Ohwue et al., 1992). The phenomenon can be seen as a kind of “reverse strain-rate sensitivity,” because the higher strain rate in an incipient or developed neck tends to increase the flow stress via strain-rate sensitivity (and thus delay failure) but also tends to decrease the flow stress via the material’s temperature sensitivity (thus promoting earlier failure) (Wagoner and Chenot, 1997; Ghosh, 2006).

The deformation-induced thermal effects are important at strain rates high enough that heat transfer out of the neck region is limited. At sufficiently low rates the deformation becomes quasi-isothermal and temperature sensitivity has a negligible effect. At sufficiently high rates the deformation is quasi-adiabatic and the heat stays where it is generated. Typical sheet-forming strain rates in the automotive industry are approximately 10/s (Fekete, 2009), which is close to the adiabatic limit for typical steels. Therefore, the role of temperature sensitivity of flow stress may be important if it is sufficiently high and the work of deformation (related to flow stress times ductility) is sufficient to raise the temperature significantly. For DP steels with high strength and ductility, an accurate description of flow stress incorporating temperature may be
essential. Such a description, if verified, may also be useful for a large range of other materials and applications.

2.2.2. Plastic Constitutive Equations

The number, range and complexity of plastic constitutive equations proposed for metals are formidable. They take many forms, depending on application and intent. The application of interest in the current work is metal forming, sheet metal forming to be more precise. Typical sheet metal formability is related to the resistance to tensile necking, which is in turn related to the evolution of plastic flow stress in biaxial tensile stress states encountered during the progressive forming. Through-thickness stress is typically close to zero (i.e. plane stress) until failure is imminent.

Multi-axial aspects of the plastic constitutive response (yield surface shape, size, and location evolution) are often separated for convenience from one-dimensional (1-D) aspects that are obtainable from tensile tests carried out at various strain rates and temperatures. The current interest is in 1-D aspects. Since many material elements undergo strain paths close to proportional, strain reversal effects such as the Bauschinger effect can be ignored for many applications.

The preponderance of sheet metal forming is carried out at moderate strain rates (up to approximately 10/s (Fekete, 2009)) and nominally room temperature (although excursions of the order of up to 100 deg. C are possible because of heat generated deformation and friction during forming (Wagoner et al., 2009b; Wagoner et al., 2009d)). For most sheet-formed metals of commercial interest (e.g. steel, aluminum, copper,
titanium) under these conditions, strain hardening is the primary material factor resisting necking, with strain-rate and temperature sensitivity of flow stress being secondary.

<table>
<thead>
<tr>
<th></th>
<th>Un bounded-stress at large strain</th>
<th>Strain hardening change with temperature</th>
</tr>
</thead>
<tbody>
<tr>
<td>Brown-Anand</td>
<td>N</td>
<td>Y</td>
</tr>
<tr>
<td>MTS</td>
<td>N</td>
<td>Y</td>
</tr>
<tr>
<td>Modified Bodner-Partom</td>
<td>N</td>
<td>Y</td>
</tr>
<tr>
<td>Lin-Wagoner</td>
<td>N</td>
<td>Y</td>
</tr>
<tr>
<td>Zirilli-Armstrong</td>
<td>Y</td>
<td>N</td>
</tr>
<tr>
<td>Khan-Huang-Liang</td>
<td>Y</td>
<td>N</td>
</tr>
<tr>
<td>Rusinek-Klepaczko</td>
<td>Y</td>
<td>Y</td>
</tr>
</tbody>
</table>

Table 2.2: Classification of integrated 1-D plastic constitutive equations that incorporate the effects of strain, strain-rate, and temperature

As listed in Appendix A, there are integrated constitutive equations relating plastic flow stress with strain, strain rate, and temperature (which is the focus of the current work). The strain hardening representations (at constant strain rate and temperature) are of two main types (Table 2.2): those that approach a saturation stress at large strain (e.g. Brown-Anand (“BA”) model (Anand, 1982, 1985; Brown et al., 1989), MTS model (Kocks, 1976; Mecking and Kocks, 1981; Follansbee and Kocks, 1988), Modified Bodner-Partom (“BP”) model (Bodner and Partom, 1975; Chen et al., 2008) and Lin-Wagoner (“LW”) Model (Lin and Wagoner, 1987)) and those that are unbounded at large strain (e.g. Zirilli-Armstrong (“ZA”) model (Zerilli and Ronald, 1987), Rusinek-

---

1 For the classification shown in Table 2.2, whenever the flow stresses for a given strain at two temperatures differ only by a constant ratio or difference, then the strain hardening rate is deemed not to depend on temperature.
Klepaczko (“RK”) model (Klepaczko, 1987; Rusinek and Klepaczko, 2001; Rusinek et al., 2007) and Khan-Huang-Liang (“KHL”) model (Khan and Huang, 1992; Khan and Liang, 1999; Khan and Zhang, 2000, 2001; Khan et al., 2004; Khan et al., 2007)). The first group may be called, in short, “saturation” or “Voce” models (Voce, 1948; Follansbee and Kocks, 1988) while the second may be called “Hollomon” or “power-law” models (Hollomon, 1945); the proper names refer to the simplest empirical versions of these strain hardening forms appearing in the literature, with 3 parameters or 2 parameters, respectively.

The saturation-type laws are typically found suitable for materials at higher homologous temperatures (and most face-centered cubic (FCC) metals such as aluminum and copper (Mishra et al., 1989; Choudhary et al., 2001) at room temperature) while power-law-type models are more suitable for body-centered cubic (BCC) metals and at low homologous temperatures. For example, iron alloys are known to continue strain hardening at strains up to at least 3 (Johnson and Holmquist, 1988).

Integrated constitutive equations may be subdivided in another way: those that accommodate strain hardening rate changes with temperature changes (e.g. BA model, MTS model, Modified BP model, LW Model and RK model) and those that do not (e.g. ZA model and KHL model), Table 2.2. As shown in Table 2.2, only the saturation-type integrated laws incorporate strain hardening rates that are sensitive to temperature (beyond fixed multiplicative or additive functions), with one exception: the RK model. The RK model (2001) incorporates power-law strain hardening with a power, n, that varies in a prescribed manner with temperature as shown in Appendix A. It is similar in concept to the LW model (1987), where strain hardening parameters in a saturation-like
law are allowed to vary linearly with temperature, thus using 6 material parameters (Appendix A).

In addition to the integrated constitutive equations mentioned above, there are myriad ways to choose and combine otherwise independent basis functions of strain, strain rate and temperature, which may be generally represented as $f(\varepsilon, T)$, $g(\dot{\varepsilon})$, and $h(T)$, respectively, where $\varepsilon$ is true tensile strain, $\dot{\varepsilon}$ is true strain rate, and $T$ is temperature. (See Appendix B for a list of typical functions if this kind. The total number of multiplicative combinations for the choices presented there alone is 126.) The basis functions may be combined multiplicatively (Hutchison, 1963; Kleemola and Ranta-Eskola, 1979; Johnson and Cook, 1983; Lin and Wagoner, 1986), additively (Wagoner, 1981b; Ghosh, 2006), or by some combination (Zerilli and Ronald, 1987) thereof. Each material model of these types prescribes the strain hardening at each temperature as a simple multiple or addition (or combination), rather than allowing for the character or rate of strain hardening to vary with temperature, hence the definition adopted for Table 2.2 to decide whether strain hardening is a function of temperature or not. Common choices for the basis functions $f(\varepsilon, T)$, $g(\dot{\varepsilon})$, and $h(T)$, are presented in Appendix B. Several of these combinations have been reviewed and fit to tensile data in the literature (Lin and Wagoner, 1986).

2.2.3. Purpose of the Current Work

The purpose of the current work is to develop and verify an empirical plastic constitutive model that meets the following conditions while introducing the minimum number of undetermined parameters:
1) it reproduces strain hardening accurately at large strain from fitting in the uniform
tensile range,

2) it captures both extreme kinds of strain hardening forms as well as intermediate
cases: bounded (power law) and unbounded (saturation),

3) it captures the change of strain hardening character and rate depending on
temperature, and

4) it is capable of predicting strain localization and failure under typical sheet forming
conditions of strains, strain rates, and temperatures.

In order to accomplish these goals, a new multiplicative type phenomenological
constitutive equation, the H/V model, is introduced. The H/V model is a linear
combination of the Hollomon and Voce strain hardening equations with a temperature-
dependent proportion. The temperature-dependent proportion represents the change of
strain hardening character and rate as temperature changes. Standard multiplicative forms
for strain rate sensitivity and thermal softening are incorporated for fitting and testing.
The initial application, the one for which testing is conducted, is for sheet metal forming
at rates up to 10/s with deformation-induced heating to temperatures up to 100 deg. C.
These are conditions normally encountered in nominally room-temperature press-forming
operations; the application that inspired the current work. Nonetheless, the H/V model
may have application outside of this regime with or without changes in forms to
accommodate larger ranges of strain rate and temperature.
2.3. **H/V Constitutive Model**

A new empirical work hardening constitutive model, H/V model, is proposed as three multiplicative functions, Eq. 2.1.

\[ \sigma = \sigma(\varepsilon, \dot{\varepsilon}, T) = f(\varepsilon, T) \cdot g(\dot{\varepsilon}) \cdot h(T) \]  

Eq. 2.1

Functions \( f, g, \) and \( h \) together represent the effects of strain, strain rate, and temperature, respectively, on the tensile flow stress \( \sigma \). \( g \) and \( h \) are chosen from any of several standard forms (see Appendix B), but the strain hardening function, \( f \), is novel: it incorporates the temperature sensitivity of strain hardening rate via a linear combination of Voce (saturation) (1948) and Hollomon (power-law) (1945) strain-hardening forms.

### 2.3.1. **Strain Hardening Function, \( f(\varepsilon, T) \)**

The function \( f(\varepsilon, T) \) represents the major departure and contribution of the new development. The motivation for the development of \( f(\varepsilon, T) \) is illustrated in Figure 2.1(a), which shows strain hardening curves for a dual-phase steel, DP780, at three temperatures. To compare the strain hardening curve shapes, the stresses are normalized by dividing by the yield stresses at each temperature, Figure 2.1 (b). Clearly the strain hardening rate varies with temperature, being lower at higher temperatures. This behavior, which is seldom captured by existing constitutive models, is one of the two principal motivations for the current development.
Figure 2.1: Stress-strain response of DP780 at three temperatures: (a) experimental curves, (b) the same data, replotted with the stress divided by the initial yield stress to reveal strain-hardening differences.

In this study, a strain hardening function $f(\varepsilon, T)$ of the following form is proposed:
\[ f(\varepsilon, T) = \alpha(T) f_H + (1 - \alpha(T)) \cdot f_V \]  
\text{Eq. 2.2}

\[
\begin{cases}
\alpha(T) = \alpha_1 - \alpha_2 (T - T_0) \\
f_H = H_{HV} e^{n_{HV}} \\
f_V = V_{HV} (1 - A_{HV} e^{-B_{HV} \varepsilon})
\end{cases}
\text{Eq. 2.3}
\]

where \( T_0 \) is a reference temperature (298K for simplicity), and \( \alpha_1, \alpha_2, H_{HV}, n_{HV}, V_{HV}, A_{HV}, \) and \( B_{HV} \) are material constants. The function \( \alpha(T) \) allows a more Voce-like curve at higher temperatures, and a more Hollomon-like curve at lower temperatures or vice versa, depending on the sign of \( \alpha_1 \). If \( \alpha(T) = 1 \), the H/V model becomes a pure Hollomon model and if \( \alpha(T) = 0 \) it becomes a pure Voce model. If an intermediate hardening rate is sought that does not depend on temperature, \( \alpha \) may be chosen to be constant: \( 0 < \alpha_0 < 1 \).

2.3.2. Strain Rate Sensitivity Function \( g(\dot{\varepsilon}) \), Temperature Sensitivity Function \( h(T) \)

The functions \( g(\dot{\varepsilon}) \) and \( h(T) \) within the H/V model are not novel; the exact form may be selected from among those appearing in the literature, as listed in Appendix B, or as otherwise devised. It is anticipated that the choice of form of these functions will not be important over the range of strain rates (up to \( 10^{-1}/s \)) and temperatures (25-100 deg. C) encountered in the testing carried out in the current work. For extended ranges, the choice of \( g(\dot{\varepsilon}) \) and \( h(T) \) may become more clear. The choices of terms \( g(\dot{\varepsilon}) \) and \( h(T) \) for the implementation of the H/V model tested in the current work will be explained in more detail in the next Section.
2.4. Experimental Procedures

The experiments were chosen to correspond to typical sheet forming practice applied to three DP steels representing a range of strengths typical of automotive body applications. Constitutive models were fit using standard tensile tests up to the uniform strain. The quality of the fits was then probed using tensile tests and hydraulic bulge tests to failure.

2.4.1. Materials

DP steels exhibit good formability and high strength derived from a microstructure that is a combination of a soft ferrite matrix and a hard martensite phase “islands”. Two grades of DP steels of nominally 1.4mm thickness, DP590 and DP780, and one grade of DP steels of nominally 1.45mm thickness, DP980, were provided by various suppliers, who requested not to be identified. DP590 was supplied without coating, DP780 with hot-dipped galvanized coating (HDGI), and DP980 with hot-dipped galvanneal coating (HDGA). The chemical compositions were determined utilizing a Baird OneSpark Optical Emission Spectrometer (HVS-OES) based on ASTM E415-99a(05) and standard tensile tests were carried out according to ASTM E8-08 at a crosshead speed of 5mm/min. Both kinds of tests were conducted at General Motors North America (GMNA, 2007). The chemical compositions and standard ASTM standard tensile properties appear in Table 2.3 and 2.4, respectively. In order to distinguish normal anisotropies measured from sheets of the original thickness and those thinned for balanced biaxial testing (as will be discussed in Section 2.4.5), symbols $r_1$ and $r_2$ in Table 2.4 refer to sheets of original thickness and thinned thickness, respectively.
Table 2.3: Chemical composition of dual-phase steels in weight percent (balance Fe)\(^1\)

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Cr</th>
<th>Al</th>
<th>Ni</th>
<th>Mo</th>
<th>Nb</th>
<th>Ti</th>
<th>V</th>
<th>B</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>DP590</strong></td>
<td>0.08</td>
<td>0.85</td>
<td>0.009</td>
<td>0.007</td>
<td>0.28</td>
<td>0.01</td>
<td>0.02</td>
<td>0.01</td>
<td>&lt;.01</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
</tr>
<tr>
<td><strong>DP780</strong></td>
<td>0.12</td>
<td>2.0</td>
<td>0.020</td>
<td>0.003</td>
<td>0.04</td>
<td>0.25</td>
<td>0.04</td>
<td>&lt;.01</td>
<td>0.17</td>
<td>&lt;.003</td>
<td>&lt;.003</td>
<td>&lt;.003</td>
<td>&lt;.002</td>
</tr>
<tr>
<td><strong>DP980</strong></td>
<td>0.10</td>
<td>2.2</td>
<td>0.008</td>
<td>0.002</td>
<td>0.24</td>
<td>0.04</td>
<td>0.02</td>
<td>0.35</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
</tr>
</tbody>
</table>

Table 2.4: Mechanical properties of dual-phase steels

<table>
<thead>
<tr>
<th></th>
<th>Thick</th>
<th>0.2% YS</th>
<th>UTS</th>
<th>(e_u) (%)</th>
<th>(e_t) (%)</th>
<th>(n^2)</th>
<th>(\Delta r^3)</th>
<th>(r_1^2)</th>
<th>(r_2^4)</th>
<th>(a^3)</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>DP590</strong></td>
<td>1.4</td>
<td>352</td>
<td>605</td>
<td>15.9</td>
<td>23.2</td>
<td>0.21</td>
<td>0.30</td>
<td>0.84</td>
<td>0.98</td>
<td>1.83</td>
</tr>
<tr>
<td><strong>DP780</strong></td>
<td>1.4</td>
<td>499</td>
<td>815</td>
<td>12.7</td>
<td>17.9</td>
<td>0.19</td>
<td>-0.11</td>
<td>0.97</td>
<td>0.84</td>
<td>1.86</td>
</tr>
<tr>
<td><strong>DP980</strong></td>
<td>1.4</td>
<td>551</td>
<td>1022</td>
<td>9.9</td>
<td>13.3</td>
<td>0.15</td>
<td>-0.23</td>
<td>0.76</td>
<td>0.93</td>
<td>1.90</td>
</tr>
</tbody>
</table>

2.4.2. Material Variation

In order to assure uniform, reproducible material properties in this work, a study was undertaken to determine variations of mechanical properties across the width of a coil. DP steels can exhibit more significant variations in this regard (as compared with traditional mild or HSLA steels) because of the complex thermo-mechanical treatment they undergo during production, Figure 2.2. The standard deviation of the ultimate tensile stress for DP590 was 25 MPa, a scatter greater than 4% of the average UTS.

\(^1\) Chemical analysis was conducted at General Motors North America (GMNA, 2007).
\(^2\) \(n\) value was calculated from an engineering strain interval of 4% to 6%.
\(^3\) \(\Delta r\) and \(r_1\) were calculated at the uniform strain of original thickness material using \(\Delta r = r_0 - 2r_45 + r_90/2\) and \(r = r_0 + 2r_45 + r_90/4\), respectively.
\(^4\) \(r_2\) value was calculated at the uniform strain of thinned materials using \(r = r_0 + 2r_45 + r_90/4\) and \(a\) was found by the fit of bulge test result to tensile test data (using Eq. 2.9) considering anisotropy \(r_2\) of thinned materials.
In order to quantify the spatial distribution of material properties and with a goal to minimize such effects on subsequent testing, numerous rolling direction (RD) tensile tests were conducted with simplified rectangular (non-shouldered) tensile specimens of DP590 having gage regions between the grips of 125mm x 20mm. (As will be shown later, see Figure 2.4, use of the rectangular specimens introduces no significant errors up to the uniform elongation.) In order to minimize the effect of sheared edge quality, the specimen edges were smoothed with 120 grit emery cloth (a practice that was followed for all tensile and draw-bend tests used in this work). The results from these specimens are summarized in Figure 2.3: specimens within 360mm of the coil edge (and thus more than 390 mm from the coil center line) exhibit systematic and significant variations in thickness and UTS, while those in the central region do not. A similar set of tests was performed for a DP980 steel (not the one used elsewhere in the current work) with the
coil width of 1500mm. The central region more than 300mm from the edges had uniform properties. The central region more than 300mm from the edges had uniform properties. For the three materials used in the current work, DP590 (coil width 1500mm), DP780 (coil width 1360mm), DP980 (coil width 1220mm), only material at least 360 mm from the coil edges was used.

![Graph](image)

Figure 2.3: Variation of sheet thickness and ultimate tensile strength with position from the edge of a coil (coil width of 1500 mm).

### 2.4.3. Tensile Testing

ASTM E8-08 sheet tensile specimens with 0% and 1% width taper were initially used for tensile testing, but failures occurred frequently outside of the gage region, thus making the consistent measurement of total elongation problematic. The failures outside of the gage region were more prevalent at elevated temperatures. In order to obtain consistent test results, the width taper was increased to 2%, which proved sufficient to
insure failure at the center of the specimen for all tests. As expected (Raghavan and Wagoner, 1987), the increased taper reduced the total elongation, but there was no significant effect on stress-strain measurement up to the uniform elongation, Figure 2.4.

![Figure 2.4: Effect of tensile specimen geometry on measured stress-strain curves.](image)

The consistency and reproducibility of tensile results after these two improvements, that is, using material from the center of the coil and specimens with 2% width tapers, are illustrated in Figure 2.5. The standard variation of UTS is less than 2 MPa (less than 0.3 % of the average value for each material), and the variation of the total elongation is less than 0.006 (less than 2.5% of the average value for each material). In view of these results, specimens with 2% tapers cut from the central region of the coil were used for all subsequent tensile testing.
For finding the parameters for the selected constitutive equations, isothermal tensile tests were conducted at 25 deg. C, 50 deg. C, and 100 deg. C using a special test fixture designed for tension-compression testing (Boger et al., 2005; Piao et al., 2009). In this application, the side plates that are pressed against the surface of the specimen (with a force of 2.24kN) were not required to stabilize the specimen mechanically against buckling but served instead to heat the specimen for the 50 deg. C and 100 deg. C tests and to maintain near-isothermal conditions throughout the contact length and testing time for all tests. For strain rates up to $10^{-3}$/s, the temperature measured in the gage length of the specimen using embedded thermocouples was maintained within 1 deg. C of the original one up to the uniform strain for all three materials. This method has several advantages compared with performing tensile tests in air in a furnace, including rapid
heat-up (about 3 minutes to achieve 250 deg. C), uniform temperature along the length, and direct strain measurement using a laser extensometer.

The data was corrected for side plate friction (which was minimized using Teflon sheets) and the slight biaxial stress (which was about 1.5 MPa) using procedures presented elsewhere (Boger et al., 2005). The friction coefficient was determined from the slope of a least-squares line through the measured maximum force vs. four applied side forces (0, 1.12, 2.24 and 3.36kN). The friction coefficients obtained in this way were 0.06, 0.05 and 0.06 for DP590, DP780 and DP980, respectively.

![Graph](image)

Figure 2.6: Variation of strain hardening and failure elongation of DP590 with test temperature.

Examples of the tensile testing data at a strain of $10^{-3}/s$ illustrate the effect of test temperature on the stress-strain curves (Figure 2.6), the yield and ultimate tensile stresses (Figure 2.7), and the uniform and total elongations (Figure 2.8). The reduced elongation to failure of higher temperatures is a direct consequence of the lower strain hardening.
Figure 2.7: Variation of yield stress and ultimate tensile stress with test temperature: (a) DP590, (b) DP780, (c) DP980.
Figure 2.8: Ductility change with temperature: (a) uniform elongation ($e_u$), (b) total elongation ($e_f$).
For the purpose of comparison with FE simulation, non-isothermal tensile tests were also conducted at a strain rate of $10^{-3}$/s in air at room temperature.

2.4.4. Strain Rate Jump Tests

For materials with limited strain rate sensitivity relative to strength, even small differences of strength from specimen-to-specimen introduce large relative errors. Jump-rate tensile tests remove specimen-to-specimen variations and thus are preferred under these conditions. Strain rate jump-down tests, i.e. abruptly changing from a higher strain rate to lower strain rate, were employed in this study. (Stress-relaxation tests can also remove specimen variations but usually can only be used for very low strain rates, less than $10^{-6}$/s (Wagoner, 1981b).) The strain rate is changed by controlling crosshead speed ranging from 2400mm/min to 0.5mm/min. Fourteen down-jump rate changes, with one jump per tensile test (Saxena and Chatfield, 1976), were conducted at engineering strains of 0.1, 0.08, 0.06 for DP590, DP780, and DP980 steels, respectively, using the following pairs of rates: $0.5/s \rightarrow 0.1/s$, $0.5/s \rightarrow 0.05/s$, $0.1/s \rightarrow 0.01/s$, $0.1/s \rightarrow 0.001/s$, $0.05/s \rightarrow 0.01/s$, $0.05/s \rightarrow 0.005/s$, $0.01/s \rightarrow 0.001/s$, $0.001/s \rightarrow 0.0001/s$. The jumps $0.1/s \rightarrow 0.01/s$, $0.01/s \rightarrow 0.001/s$ and $0.001/s \rightarrow 0.0001/s$ were repeated three times. It has been shown that strain rate sensitivity is nearly independent of strain for steels (Wagoner, 1981a), which makes jump tests at a single intermediate strain sufficient.

For each down jump, a logarithmic strain rate sensitivity value was determined from the flow stresses $\sigma_1$ and $\sigma_2$ at the two strain rates $\varepsilon_1$ and $\varepsilon_1$, respectively, for an average strain rate of the two tested strain rates as shown:
\[
\frac{\sigma_2}{\sigma_1} = \left( \frac{\dot{\varepsilon}_2}{\dot{\varepsilon}_1} \right)^m \rightarrow m = \frac{\ln(\sigma_2 / \sigma_1)}{\ln(\dot{\varepsilon}_2 / \dot{\varepsilon}_1)} \tag{Eq. 2.4}
\]

\[
\dot{\varepsilon}_{\text{average}} = \sqrt{\dot{\varepsilon}_1 \dot{\varepsilon}_2} \tag{Eq. 2.5}
\]

Figure 2.9: Schematic illustrating the procedure for calculating of the strain rate sensitivity, m, for a down jump from $10^{-1}$/s to $10^{-2}$/s), for DP590 at an engineering strain of 0.1 (true strain of 0.095).

The stress after the jump shows a transient response that can be minimized by extrapolating both the higher rate curve and the lower rate curve to a common true strain 0.005 beyond the jump strain, at which point $\sigma_1$ and $\sigma_2$ are found (Wagoner, 1981a). The procedure is illustrated in Figure 2.9 for the case of DP590 with a jump from $10^{-1}$/s to $10^{-2}$/s. The response at the higher first strain rate over a true strain range of 0.02 is fit to a fourth-order polynomial and extrapolated to the common strain. The response at the
second lower strain rate is similarly fit to a strain range of 0.02 starting at a true strain 0.01 higher than the jump strain, and is extrapolated to the common strain. The stresses from these extrapolated curves at the common strain are used to determine the m value from Eq. 2.4.

2.4.5. Hydraulic Bulge Tests

Hydraulic bulge tests were conducted at the Alcoa (ATC, 2008). Because of force limits of the ATC hydraulic bulge test system, the materials were machined from one side, from an as-received thickness of 1.4mm to 0.5mm. The stress-strain response of thinned materials was within standard deviations of 5 - 10 MPa of the original thickness materials depending on the material. The die opening diameter was 150mm and the die profile radius was 25.4mm. For materials that exhibit in-plane anisotropy, the stress state near the pole is balanced biaxial tension (Kular and Hillier, 1972) with through-thickness stress negligible for small thickness/bulge diameter ratios (Ranta-Eskola, 1979). In-plane membrane stress, $\sigma_b$, and the magnitude of thickness strain, $\varepsilon_t$, are given by Eq. 2.6 and Eq. 2.7, respectively.

\[
\sigma_b = \frac{pR}{2t} \quad \text{Eq. 2.6}
\]

\[
\varepsilon_t = 2 \ln\left(\frac{D}{D_0}\right) \quad \text{Eq. 2.7}
\]

where $p$ is pressure, $R$ is a radius of bulge, $t$ is current thickness, $D$ is current length of extensometer, and $D_0$ is the initial length of the extensometer, 25.4mm. The radius of
curvature $R$, is measured using a spherometer with each leg located at a fixed distance of 21.6mm from the pole. More detailed information for the Alcoa testing machine has appeared elsewhere (Young et al., 1981).

For an isotropic material, the von Mises effective stress and strain are equal to $\sigma_b$ and $\varepsilon_i$, respectively. Because the DP steels tested here exhibit normal anisotropy, corrections were made for the known normal anisotropy parameters, $r_2$ (the normal anisotropy was measured using thinned samples at the uniform strain), shown in Table 2.4 and best-fit values of the material parameter anisotropy parameter $a$ corresponding to the Hill 1979 non-quadratic yield function (Hill, 1979):

$$2(1+r_2)\sigma^a = (1+2r_2)[\sigma_1 - \sigma_2]^a + [\sigma_1 + \sigma_2]^a$$  \hspace{1cm} \text{Eq. 2.8}

The equations for obtaining the appropriate tensile effective stresses and strains for a balanced biaxial stress state have been presented (Wagoner, 1980):

$$\bar{\sigma} = \frac{2\sigma_1}{[2(1+r_2)]^{1/a}}, \quad \bar{\varepsilon} = \varepsilon_1[2(1+r_2)]^{1/a}$$  \hspace{1cm} \text{Eq. 2.9}

where $\varepsilon_1$ is half of the absolute value of the thickness strain. The value of $a$ was found for each material using Eq. 2.9 and equating the effective stress and strain from a tensile test and bulge test at an effective strain equal to the true uniform strain in tension for each material, Table 2.4.
2.4.6. Coupled Thermo-Mechanical Finite Element Procedures

The proposed constitutive equation was tested using a thermo-mechanical FE model of a tensile test with the same specimen geometry as in the experiments. ABAQUS Standard Version 6.7 (ABAQUS, 2007) was utilized for this analysis. One half of the physical specimen is shown in Figure 2.10 with thermal transfer coefficients, but only

Figure 2.10: FE model for tensile test: (a) schematic of the model with thermal transfer coefficients, (b) central region of the mesh before deformation, (c) central region of the mesh after deformation.
one-quarter of the specimen was modeled, as reduced by mirror symmetry in the Y and Z directions. Eight-noded solid elements (C3D8RT) were used for coupled temperature-displacement simulations with 2 element layers through the thickness. The grip was modeled as a rigid body.

A von Mises yield function and the isotropic hardening law were adopted for simplicity in view of the nearly proportional uniaxial tensile stress path throughout most of the test and the normal anisotropy values near unity for the DP steels, Table 2.4. The 2% tapered specimen geometry required for experimental reproducibility automatically initiates plastic strain localization at the center of the specimen without introducing numerical defects for that purpose. The total elongation \( (e_t) \) from each simulation was defined by the experimental load drop at failure. The simulation accuracy was evaluated based on the comparison of total elongation with experiments at this same load.

### 2.4.7. Determination of Thermal Constants

For a thermal-mechanical FE simulation, various thermal constants are required. Thermal expansion coefficient, heat capacity and thermal conductivity were determined using JMatPro (Sente-Software, 2007) based on chemical composition of DP590: thermal expansion coefficient: linear variation from \( 1.54 \times 10^{-6} \) 1/K at 25 deg. C to \( 1.58 \times 10^{-6} \) 1/K at 200 deg. C, heat capacity: linear variation from 0.45 J/gK at 25 deg. C to 0.52 J/gK at 200 deg. C, and thermal conductivity: piecewise linear variation of 36.7 W/mK at 25 deg. C, 36.9 W/mK at 70 deg. C, 36.8 W/mK at 100 deg. C and 36 (W/mK) at 200 deg. C. The coefficients were almost same for other DP steels, so the values found from DP590 were employed for DP780 and DP980. Heat transfer coefficient of metal-air contact was
measured as 20W/m²K by comparing temperature changes of a block of steel which was heated to 250 deg. C with FE simulation and heat transfer coefficient of metal-metal contact were taken from the literature as 5KW/m²k (Burte et al., 1990).

A fraction of the plastic work done during deformation is converted to heat, with the remainder stored elastically as defects (Hosford and Caddell, 1983). The relationship between adiabatic temperature rise and plastic work is as follows:

\[ \Delta T = \frac{\eta}{\rho C_p} \int \sigma d\varepsilon_p \]  

where \( \eta \) is a fraction of heat conversion from plastic deformation, \( \rho \) is density of the material and \( C_p \) is heat capacity at constant pressure. In order to measure \( \eta \) for DP steels, temperatures were recorded during tensile tests using three thermocouples that were capacitance discharge welded onto specimens on the longitudinal centerline at locations at the center and ±5mm from the center. The specimen was wrapped with glass fiber cloth to establish quasi-adiabatic conditions at a strain rate of 5×10⁻²/s. Figure 2.11 shows the temperature change of results for DP590 and the acceptable correspondence to a value of 0.9 for the parameter \( \eta \). \( \eta \) was measured only for DP590, but the value 0.9 was used for each of the three materials.
Figure 2.11: Flow stress and temperature increase during a tensile test at a nominal strain rate of $5 \times 10^{-2}$ (1/s), DP590.

2.5. Results and Discussion

2.5.1. Determination of Best-Fit Constitutive Constants

As shown by the experimental points shown in Figure 2.12, the strain rate sensitivity index, $m$, is not independent of strain rate for any of the steels, so the power-law model (Eq. B.8) and Johnson-Cook rate law (Eq. B.9) were rejected as inadequate. The Wagoner rate law (Eq. B.10) and a simplified version the Wagoner law with a linear variation of $m$ with logarithmic value of strain rate, Eq. 2.11 below, represent the data equally over this strain rate range as shown in Figure 2.12, so the simpler law was selected:

$$\sigma = \sigma_0 \left( \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right)^{0.5 + \eta (1/2) \log(\dot{\varepsilon}/\dot{\varepsilon}_0)}$$  \hspace{1cm} \text{Eq. 2.11}
where $\sigma_{\varepsilon_0}$ is a stress at $\dot{\varepsilon}_0$ and $\gamma_1$ and $\gamma_2$ are material constants.

Figure 2.12: Measured strain rate sensitivities, $m$, and their representation by strain rate sensitivity laws.
The linear variation law, Eq. 2.11, had slightly better correlation coefficients ($R^2=0.94-0.99$) than those for the Wagoner rate law ($R^2=0.82-0.97$). The best-fit values for the parameters for each model are shown in Table 2.5.

<table>
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<tr>
<th></th>
<th>Power law($m_{avg}$)</th>
<th>Wagoner rate law</th>
<th>Linear law</th>
</tr>
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<tr>
<td></td>
<td>$m_1$</td>
<td>$m_0$</td>
<td>$\gamma_1$</td>
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<td>0.0062</td>
<td>0.162</td>
<td>0.0125</td>
</tr>
<tr>
<td>DP780</td>
<td>0.0050</td>
<td>0.378</td>
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<td>DP980</td>
<td>0.0041</td>
<td>0.289</td>
<td>0.0148</td>
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</table>

Table 2.5: Parameters describing the strain rate sensitivity according to three laws.

The remainder of the H/V law was fit using results from continuous, isothermal tensile tests in the uniform strain range conducted at a strain of $10^3$/s and at 3 temperatures: 25, 50, and 100 deg.C, Figure 2.13 (Note that the H/V model reproduces the variation of strain hardening rate in the application range 25 deg. C – 100 deg. C, one of the key objectives of this work). In order to choose a suitable form for the thermal softening function $h(T)$, three forms were initially compared as shown in Figure 2.14: Linear (Eq. (B.11)), Power Law 1 (Eq. (B.12)), and Johnson-Cook (Eq. (B.14)). As shown in Figure 2.14, and not unexpectedly, there is no significant difference in the accuracy of the fits of these three laws over the small temperature range tested. The Linear model (Eq. (B.11)) was chosen for the current work as perhaps the more common and simpler choice, but no advantage is implied. For application over a larger temperature range or for other materials, presumably one of the forms shown in
Appendix B would provide a measurable advantage over the others and thus could be adopted.

Figure 2.13: Comparison of isothermal test data and best fit H/V model for DP590, DP780 and DP980.

Figure 2.14: Comparison of temperature dependent functions h(T) for DP980.
With the form of the thermal function determined, the 8 optimal coefficients ($\alpha_1, \alpha_2, H_{HV}, n_{HV}, V_{HV}, A_{HV}, B_{HV}, \beta$) may be determined for the combined function, again using results from the 3 tensile tests (at three temperatures) for each material:

$$f(\varepsilon, T) \cdot h(T) = [\alpha(T)(H_{HV} \cdot \varepsilon^{n_{HV}}) + (1 - \alpha(T))V_{HV}(1 - A_{HV}e^{-B_{HV}\varepsilon})] \cdot [1 - \beta(T - T_0)] \quad \text{Eq. 2.12}$$

Using the method of least squares, high and low values of each variable were selected as initial values, producing $2^8$ sets of starting sets of parameters. The approximate range of parameters were known with Hollomon and Voce fit with the material data, therefore the initial values of each parameter were selected based on the range. If fit values are out of the range, new initial values were selected and fit again. The initial values are shown in the Table 2.6. Best-fit coefficients were found for each starting set when the absolute value of the difference between the norm of the residuals (square root of the sum of squares of the residuals), from one iteration to the next, was less than 0.001. The set exhibiting the minimum $R^2$ value was chosen as optimal, as reported in Table 2.6. The accuracy of the fit is shown in the Table 2.6 as a standard deviation between experiment and fit line and illustrated graphically in Figure 2.14.
### Table 2.6: Trial and best-fit coefficients of the H/V model

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Table 2.6: Trial and best-fit coefficients of the H/V model
Using identical procedures, except for the RK model which was fit according to a procedure recommended by its originators (Rusinek, 2009), several alternative constitutive models were fit to the same tensile test and jump test data, as follows:

1. H/V model with $\alpha=1$ (Hollomon strain hardening at all temperatures)
2. H/V model with $\alpha=0$ (Voce strain hardening at all temperatures)
3. H/V model with best-fit constant $\alpha=\alpha_0$ (H/V strain hardening, independent of temperature)
4. Lin-Wagoner (LW) model (Eq. (A.7)) (Voce hardening dependent on temperature)

These choices were made to be representative of the major classes of constitutive models, as reviewed in Section 2.1.2. Models 1 and 2 are power-law and saturation hardening models, respectively, that are independent of temperature (except for a multiplicative function to adjust flow stress with temperature). Models 3 and 4 allow strain hardening rates to vary with temperature within the frameworks of saturation/Voce or power-law/Hollomon forms, respectively.

The material parameters for the RK model, Eq. (A.10), were found using a procedure consisting of 6 steps recommended by the first author (Rusinek, 2009). Some fundamental constants in RK model: maximum strain rate, minimum strain rate and melting temperature were first to be taken from the literature (Larour et al., 2007) and the elastic modulus was taken to be constant because this study only covered the small
homologous temperature range. The remaining 6 steps followed for fitting the RK model are as follows:

Step 1: $\sigma^*(\dot{\varepsilon}_p, T)$ in Eq. (A.11) is zero at low strain rate and at critical temperature. In this study $\sigma^*(\dot{\varepsilon}_p, T) = 0$ at the strain rate of $10^{-4}$/s and the temperature of 300K since those were the lowest values for strain rate and temperature in this work. Using this strain rate and temperature combination, $C_1$ was calculated.

Step 2: Eq. (A.10) was fit to material data at the strain rate of $10^{-4}$/s and temperature of 300K, in which $\sigma^*(\dot{\varepsilon}_p, T) = 0$, thus allowing determination of $A(10^{-4}, 300K)$ and $n(10^{-4}, 300K)$.

Step 3: The strain rate sensitivity term, $\sigma^*(\dot{\varepsilon}_p, T)$, was fit to jump test results in order to determine $C_0$ and $m$.

Step 4: $A_j(\dot{\varepsilon}_i, T_j)$ and $n_j(\dot{\varepsilon}_i, T_j)$ were found at each strain rate and temperature using Eq. (A.10).

Step 5: $A(\dot{\varepsilon}, T)$ was fit to $A_{ij}$ values to find $A_0$ and $A_1$.

Step 6: $n(\dot{\varepsilon}, T)$ was fit to $n_{ij}$ values to find $B_0$ and $B_1$.

The optimal fit coefficients, standard deviations, and $R^2$ values of all models are shown in Table 2.7 and 2.8.
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<td>$f(\varepsilon)h(T)$</td>
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<td>S.D.(MPa)</td>
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<tr>
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<td>$H_1$: 1234</td>
<td>0.997</td>
<td>$n_1$: 0.175</td>
<td>0.997</td>
<td>$V_0$: 940.9</td>
</tr>
<tr>
<td></td>
<td>$n_1$: 0.148</td>
<td>S.D.=3.2</td>
<td>$V_0$: 1521</td>
<td>S.D.=2.8</td>
<td>$A_0$: 0.370</td>
</tr>
<tr>
<td></td>
<td>$\beta$: 1.0×10^{-3}</td>
<td></td>
<td>$A_0$: 0.289</td>
<td></td>
<td>$B_0$: 17.19</td>
</tr>
<tr>
<td></td>
<td>$\alpha_0$: 0.850</td>
<td></td>
<td>$B_0$: 37.7</td>
<td></td>
<td>$\beta$: 1.0×10^{-3}</td>
</tr>
<tr>
<td></td>
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<td></td>
<td></td>
<td></td>
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<tr>
<td>DP980</td>
<td>$H_1$: 1486</td>
<td>0.996</td>
<td>$n_1$: 0.125</td>
<td>0.999</td>
<td>$V_0$: 1111</td>
</tr>
<tr>
<td></td>
<td>$n_1$: 0.135</td>
<td>S.D.=3.9</td>
<td>$V_0$: 2154</td>
<td>S.D.=1.3</td>
<td>$A_0$: 0.353</td>
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<tr>
<td></td>
<td>$\beta$: 7.4×10^{-4}</td>
<td></td>
<td>$A_0$: 0.9</td>
<td></td>
<td>$B_0$: 24.33</td>
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<tr>
<td></td>
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<td>$B_0$: 52.3</td>
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<tr>
<td></td>
<td>$H_1$: 1332</td>
<td></td>
<td></td>
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</tr>
</tbody>
</table>

Table 2.7: Best-fit coefficients of H/V model for $\alpha(T)=1$, $\alpha(T)=0$, and $\alpha(T)=\alpha_0$.

1 For $\alpha(T)=1$ and $\alpha(T)=0$, the same initial values in Table 2.5 were used.

2 For $\alpha(T)=\alpha_0$, the same initial values in Table 2.5 were used except $V_0$, $A_0$ and $B_0$ which were found out of the range. For these 3 parameters, 500 and 2500 MPa, 0.1 and 0.9, and 10 and 60 were used for initial values, respectively.
<table>
<thead>
<tr>
<th></th>
<th>LW</th>
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<td>Trial Values</td>
<td>$f(\varepsilon)h(T)$</td>
<td>$R^2$</td>
<td>$S.D.$ (MPa)</td>
<td>$f(\varepsilon)h(T)$</td>
</tr>
<tr>
<td><strong>DP590</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$A$: 500, 1500</td>
<td>$A$: 764.5</td>
<td>0.999</td>
<td>S.D.=2.5</td>
<td>$A_0$: 1070</td>
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<td>$B$: 0.1, 0.6</td>
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<td></td>
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<td>$A_1$: 0.0738</td>
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<tr>
<td>$C_1$: -5, -40</td>
<td>$C_1$: -14.3</td>
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<td></td>
<td>$\varepsilon_0$: 0</td>
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<tr>
<td>$C_2$: -0.2, 0</td>
<td>$C_2$: -0.010</td>
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<td>$B_0$: 0.184</td>
</tr>
<tr>
<td>$\beta$: -0.9, -0.1</td>
<td>$\beta$: -0.229</td>
<td></td>
<td></td>
<td>$B_1$: -0.0479</td>
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<td></td>
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<td>$C_0$: 89.1</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>$C_1$: 0.53</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>m: 1.01</td>
</tr>
<tr>
<td><strong>DP780</strong></td>
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<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$A$: 500, 1500</td>
<td>$A$: 960.4</td>
<td>0.998</td>
<td>S.D.=2.7</td>
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<td>$B_1$: -0.0389</td>
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<td>$C_0$: 91.7</td>
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<td></td>
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<td>$C_1$: 0.53</td>
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<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>m: 1.02</td>
</tr>
<tr>
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<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>$A$: 500, 1500</td>
<td>$A$: 1118</td>
<td>0.999</td>
<td>S.D.=1.8</td>
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<td>$B$: 0.1, 0.6</td>
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<td></td>
<td>$A_1$: 0.1373</td>
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<td>$\varepsilon_0$: 0</td>
</tr>
<tr>
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<td>$C_2$: -0.032</td>
<td></td>
<td></td>
<td>$B_0$: 0.134</td>
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<tr>
<td>$\beta$: -0.9, -0.1</td>
<td>$\beta$: -0.311</td>
<td></td>
<td></td>
<td>$B_1$: -0.0286</td>
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<tr>
<td></td>
<td></td>
<td></td>
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<td>$C_0$: 83.9</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>$C_1$: 0.53</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>m: 1.01</td>
</tr>
</tbody>
</table>

Table 2.8: Trial and best-fit coefficients of LW and RK models.
Figure 2.15: Comparison of selected constitutive models at various temperatures, DP780: (a) 25 deg.C, (b) 50 deg.C, (c) 100 deg.C.

The true stress-strain curves for each of these fits for DP780 are shown in Figure 2.15. While the differences are small in the uniform tensile range (consistent with the
small standard errors of fit shown in the Table 2.6, 2.7 and 2.8), the differences become apparent when extrapolated to higher strains. Strain hardening at strains beyond the tensile uniform range is particularly important for sheet forming applications with high curvature bending, where bending strains can be large. The transition of strain hardening type of the H/V law from more Hollomon-like at room temperature to progressively more Voce-like behavior at 50 and 100 deg. C can be seen by comparing Figure 2.13(a), Figure 2.13(b), Figure 2.13(c).

Table 2.6, 2.7 and 2.8 show that the H/V model provided the lowest standard deviations of fit for each material, followed progressively by the three H/V models with constant $\alpha$, the LW model, and then the RK model. The difficulty of fitting RK model to the DP steel data has a fundamental origin: the data showed the strain hardening increases with increasing strain rate but decreases with increasing temperature, while the RK model requires that the effects of increasing strain rate and temperature have the same sign of effect on the strain hardening rate.

2.5.2. Comparison of Constitutive Model Predictions with Hydraulic Bulge Tests

Figure 2.16 compares the various constitutive models established above from tensile data with results from room-temperature (25 deg. C) hydraulic bulge tests. It is apparent visually and from the standard deviations computed over the strain range 0.03-0.7, 0.03-0.34, and 0.03-0.23 for DP590, DP780 and DP980, respectively, that the full H/V model and H/V model with the best-fit constant $\alpha_0$ value (independent of temperature) extrapolated to higher strains reproduces the hydraulic bulge test results with much greater fidelity than either Hollomon or Voce hardening models. Note that the
experimental stress-strain behavior for these materials is intermediate between standard Voce and Hollomon laws, and further that the H/V law fit from the tensile predicts the high-strain behavior much better than the standard ones. This result illustrates the achievement of one of the principal objectives of the current development predicting large strain, stress-strain curves accurately. It would be desirable to have data like that shown Figure 2.16 at other temperatures. Unfortunately, the authors are unaware of any facility capable of elevated temperature balanced biaxial testing of high strength steels.

2.5.3. Comparison of H/V Model Predictions with Tensile Tests to Failure.

In order to assess the accuracy and usefulness of the proposed constitutive model on plastic localization and failure, three kinds of tensile tests to failure were simulated using finite element procedures described in the Chapter 2.3.6. For the first set of tests, standard non-isothermal tensile tests were conducted in air at room temperature, and were compared with thermo-mechanical FE simulations using selected constitutive models. The results, Figure 2.17 (a), show that the H/V model predicts post-uniform straining accurately as compared with other models, for all three materials. The average percentage error between the measured and simulated elongation to failure, $e_f$, is 3% for the H/V model vs. 16% for the Hollomon ($\alpha =1$) law, 25% for the Voce ($\alpha = 0$) law, 5% for $\alpha(T) = \alpha_0$, 21% for LW and 33% for RK.
Figure 2.16: Comparison of constitutive models with bulge test results: (a) DP590, (b) DP780, (c) DP980.
Figure 2.17: Comparison of nonisothermal tensile test data and FE simulation using selected constitutive models: (a) at strain rate=$10^{-3}$/s, (b) at strain rate=1/s.
Similar tensile tests were conducted at strain rate of 1/s (the maximum strain rate available to the authors). Figure 2.17 (b) confirms that the H/V model predicts well the ultimate tensile strength as well as total elongation for all three DP steels at the highest rate for which tests were available, well outside of the fit range of rates.

For the third kind of assessment of strain localization and failure, tensile tests were conducted isothermally at three temperatures, 25, 50, and 100 deg.C, and were simulated using isothermal FEM. That is, the temperature was maintained at the initial temperature throughout the simulation and test. The isothermal tests and simulations using various laws are compared in Figure 2.18 for one material, DP590, and the corresponding differences of total elongation obtained. A summary of combined errors for each law and each material over all three temperatures appears in Table 2.9. The H/V model predicts the development of post-uniform necking over a range of temperatures with much better accuracy than other such models, typically better by an order of magnitude.

<table>
<thead>
<tr>
<th></th>
<th>LW</th>
<th>RK</th>
<th>( \alpha(T)=1 )</th>
<th>( \alpha(T)=0 )</th>
<th>( \alpha(T)=\alpha )</th>
<th>H/V</th>
</tr>
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<tr>
<td>DP590</td>
<td>19%</td>
<td>30%</td>
<td>23%</td>
<td>19%</td>
<td>8%</td>
<td>3%</td>
</tr>
<tr>
<td>DP780</td>
<td>21%</td>
<td>47%</td>
<td>21%</td>
<td>22%</td>
<td>11%</td>
<td>5%</td>
</tr>
<tr>
<td>DP980</td>
<td>18%</td>
<td>50%</td>
<td>29%</td>
<td>23%</td>
<td>13%</td>
<td>6%</td>
</tr>
<tr>
<td>Avg.</td>
<td>19%</td>
<td>42%</td>
<td>24%</td>
<td>21%</td>
<td>11%</td>
<td>5%</td>
</tr>
</tbody>
</table>

Table 2.9: Difference\(^1\) of total elongations of isothermal tensile tests conducted at 25, 50, and 100 deg. C and FE simulations for various constitutive models

\(^1\) The percentage error of predicted failure elongation is computed by \((e_{\text{f,FE}} - e_{\text{f,exp}}) / e_{\text{f,exp}} \times 100(\%)\) where \(e_{\text{f,FE}}\) is the predicted elongation at the time step when the predicted load matches the measured load at \(e_{\text{f,exp}}\). The percentage error shown in the table is the average of the absolute values of these percentage errors for three temperatures: 25, 50, and 100 deg.C.
Figure 2.18: Comparison of isothermal tensile test data and FE simulation using selected constitutive models for DP590: (a) at 25 deg.C, (b) at 50 deg.C, (c) at 100 deg.C. The percentage error of predicted failure elongations (defined at the measured engineering stress from experiment) is shown, and is summarized in Table 2.9.
Figure 2.19: Comparison of isothermal and non-isothermal FE simulations of tensile test using the H/V model at two nominal strain rates: (a) strain rate=10/s, (b) strain rate=10^{-3}/s.

Figure 2.19 shows the simulated differences of post-uniform straining for isothermal and non-isothermal tests. As expected, the differences at a nominal strain rate of 10^{-3}/s are small because there is sufficient heat flow to approximate isothermal condition. For
the simulated tests conducted at a nominal strain rate of 10/s, the effect of deformation-induced heating is greater because the heat flow is restricted for the shorter test time, thus approaching adiabatic conditions. The standard deviations for H/V predictions are factors of 2 to 8 less than those for the other models.

2.6. Summary and Conclusions

Standard jump tests and isothermal tensile tests of DP590, DP780 and DP980 steels in the uniform strain range at 25, 50 and 100 deg. C have been used to fit constitutive equations from the literature and as proposed in the current work (“H/V Model”). The accuracy of these laws was compared using tensile tests to failure (isothermal and standard) and balanced biaxial bulge tests, and parallel simulations. The following conclusions were reached:

1. The H/V Model provides a natural form to incorporate the transition of strain hardening from power-law-type (Hollomon-like) at low homologous temperatures (and for many bcc alloys) to saturation-type (Voce-like) at higher homologous temperatures (and for many bcc alloys).

2. The H/V Model, fit to the uniform strain range of tensile data, provides more accurate predictions of large-strain stress-strain behavior than existing models in the literature.

3. The H/V Model, fit to the uniform strain range of tensile data, predicts tensile failure strains more accurately, by factors of 2 to 8, than existing models in the literature.

4. Deformation-induced heating at normal industrial strain rates affects the strain hardening of DP steels significantly, thus promoting strain localization and failure.
5. The accuracy of predicted failure strains and large-strain stress-strain curves using the H/V Model, and the large differences between isothermal and non-isothermal predictions at industrial strain rates, suggest that damage is not a critical factor in the tensile failure of DP steels.

6. The H/V Model can be simplified by setting the linear combination coefficient equal to a constant: $\alpha = 1$ is a power-law/Hollomon model, $\alpha = 0$ is a saturation/Voce model, and $\alpha = \alpha_o$ exhibits a fixed character intermediate between Hollomon and Voce models. The last simplification provides most of the advantages of the full H/V model over the small range of temperatures investigated in the current work.

7. A heat conversion of efficiency of 0.9 was measured for DP590 steel.

8. Tapered tensile specimens with a 2% taper are sufficient to insure failure at the center for DP steels. Parallel and 1%-tapered specimens are not sufficient.

9. DP steels were found to have varying properties in the edge regions of the coils. The central regions were very uniform. The dividing line between the two regions was measured as 300-360mm from the coil edge for DP 590 and DP 980 steels having total coil widths of 1500mm.

10. Many of the common plastic constitutive laws (1-D) incorporating strain, strain rate, and temperature, have been identified and their forms presented. Selected representative ones that incorporate varying strain hardening as a function of temperature have been implemented and compared.
CHAPTER 3:

THE FORMABILITY OF DP STEELS IN DRAW-BEND FORMABILITY (DBF) TESTS

NOTE: The remainder of Chapter 3 (following this note) is a manuscript in preparation to submit the ASME journal of Engineering Materials and Technology. It represents the other of two major contributions described in this dissertation, namely development of new draw-bend formability (DBF) test and disclosure of the origin of the inability to predict the shear fractures for AHSS. The only differences between the manuscript and this chapter are Figure 3.3, which is waiting for the permission from Interlaken for publishing to a peer-reviewed journal (the permission for this dissertation was obtained) and organizational ones. That is, the figure numbers, section numbers and so on. have been changed to fit into the dissertation format. Addition deemed appropriate for a dissertation but not a journal paper are added in Appendix J to M:

Appendix J (determination of maximum displacement rate of draw-bend system for DP steels): Because the draw-bend system was designed for conventional steels and Aluminum, the capabilities of the machine in terms of load and displacement rate were investigated using DP and TRIP steels to determine the test conditions.

Appendix K (the results of draw-bend formability test): In the paper, only limited data were shown for the compactness of the paper. All data from DBF tests were listed in this Appendix.
Appendix L (strain measurement of fractured DBF sample): Strain contour of tested specimen was obtained in cooperation with General Motors. This Appendix explains the detail of procedure, results and conclusions.

Appendix M (plane strain fracture criteria): This Appendix introduces an analytical model which can explain the fracture at small R/t.

3.1. Abstract

Sheet forming failures of dual-phase (DP) steels occur unpredictably in regions of high curvature and with little apparent necking. Such failures are often referred to as “shear fractures”. In order to reproduce such fractures in a laboratory setting, and to understand their origin and the inability to predict them, a novel draw-bend formability (DBF) test was devised using dual displacement rate control. DP steels from several suppliers, with tensile strengths ranging from 590 to 980 MPa, were tested over a range of rates and bend ratios (R/t) along with a TRIP (Transformation Induced Plasticity) steel for comparison. The new test reliably reproduced three kinds of failures identified as Types I, II, and III, corresponding to tensile failure, transitional failure, and shear fracture, respectively. The type of failure depends on R/t and strain rate. Two critical factors influencing the lack of accurate failure prediction were identified. The dominant one is deformation-induced heating, which is particularly significant for advanced high strength steels because of their high heat dissipation. Temperature rises of up to 100 deg. C were observed. This factor causes reduced formability at higher strain rates, and a transition in failure from Type I to Type III. The second factor is related to microstructural features. This was observed in only one material in one test direction (one DP980(D) in the
transverse direction). Alternate measures for assessing the formability of materials, including the effect of bending, were introduced and compared. They can be used to rank the formability of competing materials and to detect processing problems that lead to unsuitable microstructures.

3.2. Background

Advanced high strength steels (AHSS) including dual-phase (DP) and transformation induced plasticity (TRIP) steels collectively offer impressive combinations of strength and ductility that can reduce the mass and improve the crash-worthiness of sheet-formed automotive parts and vehicles. These AHSS are being intensively studied by the automotive industry and its suppliers to address the demand for tighter standards such as: increased structural strength for vehicle safety, a required decrease in vehicle emissions, and increasingly stringent standards of fuel economy (Horvath and Fekete, 2004; Demeri, 2006; Opbroek, 2009). Aluminum and magnesium alloys, though attractive from a weight-savings perspective, present challenges in terms of high cost, low formability, low weldability and high primary-production emissions (Cole and Sherman, 1995; Johnson, 1995; Hall, 2008). Ducker Worldwide (Schultz and Abraham, 2009) projected that the use of flat rolled AHSS will increase threefold by 2020 as the result of their replacement of conventional and high strength steels in the North American production of light curb-weight vehicles.

The forming-limit diagram (FLD), generated with a large hemisphere punch and various widths of sheet, is based on the localized necking approach (Keeler and Backofen, 1964; Marciniak and Kuczynski, 1967b; Goodwin, 1968; Keeler, 1969) and has been
successfully used to characterize the sheet formability of a material (Embury and Duncan, 1981; Burford and Wagoner, 1989; Graf and Hosford, 1990; Bleck et al., 1998; Rees, 2001). However, depending on the application and the grade, forming failures in AHSS are not always predictable by the usual forming simulation and application of FLD (Sriram and Urban, 2003; Wagoner, 2006; Wu et al., 2006).

As shown in Figure 3.1, very large splits occurred at the small die radii in the forming of DP780 sheet. The failures were unpredicted and unexpected (Stoughton et al., 2006). This type of failure, often referred to as “shear fracture” (Walp et al., 2006; Sklad, 2008; Chen et al., 2009), occurs with little or no obvious through-thickness necking. It is observed for AHSS at the radii of forming tools, where the sheet undergoes bending and unbending under tension (Sriram and Urban, 2003; Haung et al., 2008; Kim et al., 2009a). This behavior contrasts with most experience for traditional low carbon steel alloys that
typically fail in response to stretching over gentle radii (Damborg, 1998). Significant recent effort has been applied to predict shear fractures by the use of analytical methods (Bai and Wierzbicki, 2008; Hudgins et al., 2010; Kim et al., 2010a), numerical methods (Krempaszky et al., 2007; Larour et al., 2007; Bai and Wierzbicki, 2008; Choi et al., 2009b; Kim et al., 2009b; Sung et al., 2009; Wagoner et al., 2009c; Kim et al., 2010a) and experimental methods (Sriram and Urban, 2003; Walp et al., 2006; Shih et al., 2009; Wagoner et al., 2009c; Hudgins et al., 2010). Conventional wisdom in the sheet forming area has attributed the occurrence of shear fracture to a postulated special internal damage mechanism related to the microstructures of DP steels: large 9-grain-scale) islands of high-strength martensite in a matrix of softer ferrite.

The angular stretch-bend test measures the formability limit (punch stroke to fracture) under combined bending and stretching deformation mode (Demeri, 1981; Narayanaswamy and Demeri, 1983; Sriram and Urban, 2003; Hudgins et al., 2007), but does not reproduce large draw distances and the consequent bending and unbending, of many sheet forming operations. A second limitation is the inability to control sheet tension separately from draw distance. This test ranks the formability of TRIP steel as much higher than DP steel over a range of punch radius to sheet thickness (R/t) ratios. (Sriram and Urban, 2003).

A tension-controlled draw bend fracture test was developed (Damborg et al., 1997) based on a friction testing design (Demeri, 1981; Vallance and Matlock, 1992; Wenzloff et al., 1992; Haruff et al., 1993) in order to address the shortcomings of the angular stretch-bend test. In the original, the back force (force at the back grip, Figure 3.2) was increased linearly with time while the front grip had a constant speed. The test mimics
the mechanics of deformation of sheet metal as it is drawn, stretched, bent and straightened over a die radius entering a typical die cavity. Thus it closely represents drawing of a sheet over a die radius in a forming operation, but with the added capability of careful control and measurement of sheet tension force in dependent of friction and draw distance. The original test can produce normal plastic localization/necking and shear fracture, depending on materials, sheet orientation, and die radii (Vallance and Matlock, 1992; Wenzloff et al., 1992; Haruff et al., 1993; Damborg et al., 1997; Damborg, 1998; Hudgins et al., 2007; Hudgins et al., 2010)

![Figure 3.2: The draw-bend test and its relationship to drawing over a die radius in sheet forming: (a) the DBF test equipment, specimen, and principal parts, (b) the equipment mechanics of draw over a die radius.](image)

Figure 3.2: The draw-bend test and its relationship to drawing over a die radius in sheet forming: (a) the DBF test equipment, specimen, and principal parts, (b) the equipment mechanics of draw over a die radius.

The current research was aimed at understanding the origins of sheet fracture, why it is difficult to predict using standard industrial methods, and how to predict it accurately.
Experiments were first conducted using the original draw-bend fracture test, but limitations were soon revealed. In particular, controlling the back force after necking begins leads to a reversal of sheet draw direction over part of the tool surface and makes interpretation of fracture difficult. A new draw-bend fracture test (designated "DBF test" here) was devised with constant speeds applied to both grips thus insuring a consistent draw direction and much better reproducibility. Using the new test with a wide range of R/t, draw speeds, grip speed ratios, and materials revealed the nature of shear fracture of DP steels and suggested ways of dealing with the problem.

3.3. EXPERIMENTAL PROCEDURES

3.3.1. Materials

A new draw-bend formability ("DBF") test was used to produce shear and tensile failures for a range of AHSS. AHSS exhibits good formability and high strength derived from a microstructure that is a combination of a soft ferrite matrix and a hard martensite phase “islands” for dual-phase (DP) steels, and from transformation of the retained austenite to martensite with plastic deformation for transformation-induced plasticity (TRIP) steels. Seven commercially available AHSS, three grades of DP steels of nominally 1.4mm thickness, DP590(B), DP780(D), DP980(A), DP980(D), and DP980(E), one grade of DP steel of nominally 1.2mm thickness, DP980(F), and one grade of TRIP steel (for the purpose of comparison) of nominally 1.6mm thickness, TRIP780(D), were provided by various suppliers, who requested not to be identified. (A), (B), (D), (E) and (F) indicate the different suppliers. DP590(B) and DP980(A) were supplied without
coating, DP780(D), DP980(E) and DP980(F) with Hot-Dip Galvanized (HDGI) coating, and DP980(D), TRIP780(D) with Hot-Dip Galvannealed (HDGA) coating.

<table>
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<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Cr</th>
<th>Al</th>
<th>Ni</th>
<th>Mo</th>
<th>Nb</th>
<th>Ti</th>
<th>V</th>
<th>B</th>
</tr>
</thead>
<tbody>
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<td>0.85</td>
<td>0.009</td>
<td>0.007</td>
<td>0.28</td>
<td>0.01</td>
<td>0.02</td>
<td>0.01</td>
<td>&lt;.01</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
</tr>
<tr>
<td>DP780(D)</td>
<td>0.12</td>
<td>2.0</td>
<td>0.020</td>
<td>0.003</td>
<td>0.04</td>
<td>0.25</td>
<td>0.04</td>
<td>&lt;.01</td>
<td>0.17</td>
<td>&lt;.003</td>
<td>&lt;.003</td>
<td>&lt;.003</td>
<td>&lt;.003</td>
</tr>
<tr>
<td>DP980(D)</td>
<td>0.10</td>
<td>2.2</td>
<td>0.008</td>
<td>0.002</td>
<td>0.24</td>
<td>0.44</td>
<td>0.06</td>
<td>&lt;.01</td>
<td>0.35</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
</tr>
<tr>
<td>DP980(A)</td>
<td>0.14</td>
<td>1.2</td>
<td>0.01</td>
<td>0.006</td>
<td>0.29</td>
<td>0.02</td>
<td>0.05</td>
<td>&lt;.01</td>
<td>&lt;.003</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
</tr>
<tr>
<td>DP980(E)</td>
<td>0.09</td>
<td>2.0</td>
<td>0.008</td>
<td>0.003</td>
<td>0.61</td>
<td>0.01</td>
<td>0.03</td>
<td>0.01</td>
<td>0.1</td>
<td>0.01</td>
<td>0.015</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
</tr>
<tr>
<td>DP980(F)</td>
<td>0.07</td>
<td>2.3</td>
<td>0.017</td>
<td>0.003</td>
<td>0.08</td>
<td>0.99</td>
<td>0.03</td>
<td>0.02</td>
<td>0.03</td>
<td>0.049</td>
<td>0.016</td>
<td>&lt;.002</td>
<td>&lt;.002</td>
</tr>
<tr>
<td>TRIP780(D)</td>
<td>0.15</td>
<td>2.1</td>
<td>0.016</td>
<td>0.003</td>
<td>0.06</td>
<td>0.12</td>
<td>1.5</td>
<td>&lt;.01</td>
<td>0.09</td>
<td>&lt;.003</td>
<td>&lt;.003</td>
<td>&lt;.003</td>
<td>&lt;.003</td>
</tr>
</tbody>
</table>

Table 3.1: Chemical composition of experimental AHSS in weight percent.

The chemical compositions were determined with a Baird OneSpark Optical Emission Spectrometer (HVS-OES) based on ASTM E415-99a, and standard tensile tests were carried out according to ASTM E8 at a crosshead speed of 5mm/min. Both kinds of tests were conducted at General Motors North America (GMNA, 2007). The chemical compositions and ASTM standard tensile properties appear in Table 3.1 and Table 3.2, respectively. Figure 3.3 compares tensile curves of tested materials.

Detailed constitutive equations expressing flow stress as a function of strain, strain rate, and temperature have been presented for these alloys. The critical aspect of the measured behavior is a decrease of strain hardening rate with temperatures up to at least 100 deg. C (Sung et al., 2009).
Table 3.2: Mechanical properties of the selected materials in weight percent

<table>
<thead>
<tr>
<th>Material</th>
<th>Thickness (mm)</th>
<th>UTS (MPa)</th>
<th>0.2%YS (MPa)</th>
<th>e_u (%)</th>
<th>e_t (%)</th>
<th>n²</th>
<th>r³</th>
<th>YS/UTS</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP590(B)</td>
<td>RD 1.39</td>
<td>605</td>
<td>352</td>
<td>15.9</td>
<td>23.2</td>
<td>0.21</td>
<td>1.02</td>
<td>0.58</td>
</tr>
<tr>
<td></td>
<td>TD 1.39</td>
<td>616</td>
<td>359</td>
<td>15.8</td>
<td>23.2</td>
<td>0.21</td>
<td>1.25</td>
<td>0.58</td>
</tr>
<tr>
<td>DP780(D)</td>
<td>RD 1.40</td>
<td>815</td>
<td>499</td>
<td>12.7</td>
<td>17.9</td>
<td>0.19</td>
<td>0.87</td>
<td>0.61</td>
</tr>
<tr>
<td></td>
<td>TD 1.40</td>
<td>810</td>
<td>486</td>
<td>12.9</td>
<td>17.2</td>
<td>0.18</td>
<td>0.69</td>
<td>0.60</td>
</tr>
<tr>
<td>DP980(D)</td>
<td>RD 1.43</td>
<td>1022</td>
<td>551</td>
<td>9.9</td>
<td>13.3</td>
<td>0.15</td>
<td>0.82</td>
<td>0.54</td>
</tr>
<tr>
<td></td>
<td>TD 1.43</td>
<td>1021</td>
<td>584</td>
<td>9.4</td>
<td>13.3</td>
<td>0.13</td>
<td>0.80</td>
<td>0.57</td>
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<tr>
<td>DP980(A)</td>
<td>RD 1.45</td>
<td>1018</td>
<td>665</td>
<td>9.5</td>
<td>14.2</td>
<td>n/a</td>
<td>1.13</td>
<td>0.65</td>
</tr>
<tr>
<td></td>
<td>TD 1.45</td>
<td>1010</td>
<td>653</td>
<td>9.5</td>
<td>13.5</td>
<td>n/a</td>
<td>0.92</td>
<td>0.65</td>
</tr>
<tr>
<td>DP980(E)</td>
<td>RD 1.39</td>
<td>1025</td>
<td>683</td>
<td>8.3</td>
<td>13.4</td>
<td>0.10</td>
<td>0.74</td>
<td>0.67</td>
</tr>
<tr>
<td></td>
<td>TD 1.39</td>
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<td>687</td>
<td>8.3</td>
<td>13.4</td>
<td>0.09</td>
<td>1.05</td>
<td>0.65</td>
</tr>
<tr>
<td>DP980(F)</td>
<td>RD 1.19</td>
<td>989</td>
<td>736</td>
<td>6.8</td>
<td>10.6</td>
<td>0.07</td>
<td>0.82</td>
<td>0.74</td>
</tr>
<tr>
<td></td>
<td>TD 1.19</td>
<td>1019</td>
<td>722</td>
<td>6.5</td>
<td>11.1</td>
<td>0.07</td>
<td>0.92</td>
<td>0.71</td>
</tr>
<tr>
<td>TRIP780(D)</td>
<td>RD 1.60</td>
<td>857</td>
<td>471</td>
<td>14.9</td>
<td>19.2</td>
<td>0.22</td>
<td>0.81</td>
<td>0.55</td>
</tr>
<tr>
<td></td>
<td>TD 1.57</td>
<td>860</td>
<td>501</td>
<td>13.9</td>
<td>18.4</td>
<td>0.22</td>
<td>1.20</td>
<td>0.58</td>
</tr>
</tbody>
</table>

Figure 3.3: Comparison of stress-strain curves

1 Chemical composition was analyzed at the GMNA Materials Laboratory based on ASTM E415-99a.
2 n value calculated from 4% to 6% strain.
3 r-value was calculated at end of uniform elongation.
4 Tests were conducted at GMNA Materials Lab, GMNA, 2007, GMNA Materials Lab., 660 South Blvd., Pontiac, MI, USA.
3.3.2. Draw-Bend Formability (DBF) Test

A specially-designed 90-degree draw-bend test system (Figure 3.4) was used for this study. The system is a fully closed-loop control system for two linear actuators oriented on axes 90 degrees from each other. The maximum tension capability of each actuator is 44.5 kN, the maximum speed is 150 mm/s depending load, and each grip can travel up to 250 mm. A roller is positioned at the intersection of the two linear actuators and connected to a rotating actuator which can be controlled independently, that is, it may be locked in place or rotated at a given rate. Roller sets with eight choices of radii ranging from 3 to 17 mm are available, as shown in Figure 3.5.
In the original force-controlled test the front actuator was controlled at a constant speed \( (V_1) \), typically 25 mm/s, and the back actuator was controlled to increase force linearly with time. As reported in the literature, this test reproduced tensile type failure and shear failure, depending on \( R/t \), the maximum restraining force and the load rate (Damborg et al., 1997; Damborg, 1998; Hudgins et al., 2007; Hudgins et al., 2010). However, the back portion of the specimen reversed displacement direction at some point in the test, as the control algorithm altered to imposing continuously increasing back forces which were not physically possible. Thus, the actual motion and force on the back actuator were not consistent or predictable and in some cases fracture occurred on the
back portion of the specimen (Figure 3.6). This indeterminate behavior made interpretation of fracture results difficult or impossible.

Therefore, a new test method was devised to maintain constant speeds of both grips, $V_1$ and $V_2$, with $V_2/V_1 = \alpha$, a constant. In concept, the test is similar to a tensile test (i.e. constant extension rate axially) with superimposed drawing over a radius to produce bending of the specimen. Contrary to the original back-force-controlled test, the DBF guarantees drawing in one direction and failures that occur on the front side of the specimen (i.e. beyond the contact center point). A typical stress-time curve and

Figure 3.6: Fracture types and stress-displacement curves in load rate control DBF test.

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displacement-time curve is shown in Figure 3.7. Preliminary tests reproduced three types of fractures, depending on $V_1$, $\alpha$ and $R/t$, as presented in the next section.

![Stress and displacement curves](image)

Figure 3.7: Stress and displacement curves of front and back grips for speed control mode on both grips.

The DBF, like all bending, involves a wide range of strain rates that occur simultaneously. The maximum strain rate prior to strain localization (necking/shear fracture) occurs on the outer fibers of the sheet during bending and unbending with superimposed tension approximately equal to the flow stress. The maximum true strain that occurs under that conditions, $\varepsilon_{\text{max}}$, is equal to the bending strain for bending under the neutral axis is located at the inner surface of the sheet:

$$\varepsilon_{\text{max}} = \ln(1 + t / R)$$

Eq. 3.1
The time to attain this strain is less clear; it is related to the draw rate over the tool radius (essentially $V_1$) and the draw distance needed to establish the bend radius $d_{bend}$:

$$\dot{\varepsilon}_{\text{max}} \approx \frac{\varepsilon_{\text{max}}}{(d_{bend} / V_1)}$$  \hspace{1cm} \text{Eq. 3.2}

Finite element simulations of the DBF test show that $d_{bend} \approx 3t$ ($t=$initial sheet thickness) (Wagoner et al., 2009a) under a wide range of conditions, so $\dot{\varepsilon}_{\text{max}}$ takes the explicit form:

$$\dot{\varepsilon}_{\text{max}} = \frac{\varepsilon_{\text{max}} V_1}{3t} = \frac{V_1}{3t} \ln(1 + t / R)$$  \hspace{1cm} \text{Eq. 3.3}

The maximum stretching strain rate experienced in industrial sheet forming operations is approximately 10/s (Fekete, 2009). In order to attain this rate, even using the smallest tool radius available (3.2mm), $V_1$ must exceed 100mm/s. For this reason, initial testing was performed at rates up to 125mm/s for DP590(B). However, for DP980(D), the higher $V_1$ that could be obtained reliably was 51mm/s, so this maximum rate was used for subsequent testing. Alternate pull rates of 2.5mm/s and 13mm/s were used to explore the effect of strain rate. Similarly grip speed ratios of $\alpha=0$ and 0.3 were used. $\alpha=0.3$ is the largest ratio that was able to produce fracture for the most formable materials given the limited pull distance available.

The effect of the friction condition between roller and sheet was studied using three different friction conditions with DP980(D): driven/unlubricated, fixed/lubricated and fixed/unlubricated rollers. The friction coefficients of three cases were estimated as 0.05,
0.1, and 0.15 using a constitutive model, H/V model, (Sung et al., 2009) and a coupled thermo-mechanical FE model (Kim et al., 2009a) proposed by authors in other places. For subsequent tests, the fixed/lubricated condition was consistently used. The fixed roller was lubricated on the contact side with a normal stamping lubricant, Parco Prelube MP-404.

<table>
<thead>
<tr>
<th>Material</th>
<th>Orientation</th>
<th>Material Orientation</th>
<th></th>
<th>V₁ (mm/s)</th>
<th>φ</th>
<th>Radii of Roller (mm)</th>
<th>Friction Condition</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP590(B)</td>
<td>RD</td>
<td>2.5, 13 and 51</td>
<td>0</td>
<td>0.3</td>
<td>3.2, 4.8, 6.4, 7.9, 9.5, 11.1, 14.3, 19</td>
<td>Fixed/Lub</td>
<td></td>
</tr>
<tr>
<td>DP780(D)</td>
<td>RD</td>
<td>2.5, 13 and 51</td>
<td>0</td>
<td>0.3</td>
<td>3.2, 4.8, 6.4, 7.9, 9.5, 11.1, 14.3, 19</td>
<td>Fixed/Lub</td>
<td></td>
</tr>
<tr>
<td>TRIP780(D)</td>
<td>RD</td>
<td>2.5, 13 and 51</td>
<td>0</td>
<td>0.3</td>
<td>3.2, 4.8, 6.4, 7.9, 9.5, 11.1, 14.3, 19</td>
<td>Fixed/Lub</td>
<td></td>
</tr>
<tr>
<td>DP980(D)</td>
<td>RD, TD</td>
<td>2.5, 13 and 51</td>
<td>0</td>
<td>0.3</td>
<td>3.2, 4.8, 6.4, 7.9, 9.5, 11.1, 14.3, 19</td>
<td>Fixed/Lub/Driven</td>
<td></td>
</tr>
<tr>
<td>DP980(A)</td>
<td>RD, TD</td>
<td>2.5, 13 and 51</td>
<td>0</td>
<td>0.3</td>
<td>3.2, 4.8, 6.4, 7.9, 9.5, 11.1, 14.3, 19</td>
<td>Fixed/Lub</td>
<td></td>
</tr>
<tr>
<td>DP980(E)</td>
<td>RD, TD</td>
<td>51</td>
<td>0</td>
<td>0.3</td>
<td>3.2, 4.8, 6.4, 7.9, 9.5, 11.1, 14.3, 19</td>
<td>Fixed/Lub</td>
<td></td>
</tr>
<tr>
<td>DP980(F)</td>
<td>RD, TD</td>
<td>51</td>
<td>0</td>
<td>0.3</td>
<td>3.2, 4.8, 6.4, 7.9, 9.5, 11.1, 14.3, 19</td>
<td>Fixed/Lub</td>
<td></td>
</tr>
</tbody>
</table>

Table 3.3: Test matrix of draw-bend tests for selected materials.

Sheet samples were mainly cut parallel to the rolling direction (RD), and with some samples (DP980s only) cut parallel to the transverse direction (TD) for comparison. The strips were sheared to 25-mm width and 660-mm length with width varying less than 0.2mm along the entire length. The sheared specimen edges were smoothed with 120-grit SiC emery cloth to remove burrs and reduce edge effects. The strips were first mounted into the back grip, then manually bent over the roller and mounted into the front grip.
Each test continued until the sample failed, either by tensile plastic localization/necking or by shear fracture. The detail DBF test matrix was shown in Table 3.3.

3.4. Results and Discussions

3.4.1. Fracture Types

DBF tests reproduced three visually-identifiable types of fracture, Figure 3.8, depending on three process parameters: draw speed \((V_1)\), draw speed ratio \((\alpha)\), and die-radius to sheet-thickness ratio \((R/t)\).

![Figure 3.8: Examples of three types of fracture types with the dual displacement rate controlled test.](image)

TYPE I is a standard plastic localization, or necking, as found in a standard tensile test of material. As shown in Figure 3.9 (a) and (d), it can be identified unambiguously because it occurs in the front specimen leg, away from any material that has been drawn
over the tool radius. Usually the fracture occurred along an angle of 60 and 65 degrees to the pulling direction; but in some materials, e.g. DP980(D), the fractures were perpendicular to the pulling direction.

Figure 3.9: Fracture types of DBF test.

TYPE III is what is often called “shear fracture.” It always occurs either over the roller, or at the exit tangent point, and it propagates in a direction perpendicular to the strip axis, Figure 3.9 (b) and (d). The shear fracture can be divided into two sub-types (Figure 3.10): one occurs on a sharp radius and the other is found at the tangent point of
the roller (Haung et al., 2008; Shih et al., 2009). However, it is difficult to consistently distinguish between the fracture sub-types in experiments; consequently no distinction will be made in the current work.

![Figure 3.10: Two types of Type III.](image)

(a) Failure at tangent exit of roller  (b) Failure over the roller

TYPE II is a transitional-mode fracture: It initiates at a specimen edge in a manner and location similar to TYPE III, but propagates at an angle like TYPE I, in material that has been drawn over the tooling, Figure 3.9 (c) and (d). Because the initiation is like TYPE III and it occurs in the bent-unbent region((Figure 3.9 (d)), it is treated as a transitional kind of shear fracture, likely related to finite width of the specimen. TYPE II is unlikely to occur in typical industrial forming, where tight-radius features are usually very long relative to the sheet thickness and die radius.

3.4.2. Effect of Friction Condition

Figure 3.11 (a) compares the displacement-force curves of DBF tests with three friction conditions: driven/unlubricated, fixed/lubricated and fixed/unlubricated rollers. In
DBF tests, the front drawing force ($F_1$) is the sum of the back restraining force ($F_2$), the bending and unbending force ($F_b$) and the friction force ($F_f$), Eq. (4) (Damborg, 1998).

$$F_1 = F_2 + F_b + F_f$$  \hspace{1cm} \text{Eq. 3.4}$$

$F_1$ does not change with the friction force, so $F_2$ decreases as the friction force decreases. Figure 3.11 (a) shows $F_2$ decreases as friction is lowered. The low friction coefficient permits the sheet to flow over the roller more easily, resulting in more deformation prior to fracture as shown in Figure 3.11 (a).

![Figure 3.11: Effect of roller condition: (a) force-displacement curves at various roller conditions, (b) Failure type maps with fixed and free rollers](image)

However, more plastic deformation of the sheet produces more heat generated by plastic deformation to the sheet, especially near the roller, therefore the specimen became more susceptible to TYPE III fracture. Figure 3.11 (b) shows the comparison of fracture type maps using the free roller and the fixed roller for DP780(D). The shifting of the $(R/t)^*$ line from right to left shows explicitly that the lower friction coefficient gives more susceptibility to TYPE III fracture. The direction of friction effect in the fracture map
agreed with FE simulation of a coupled thermo-mechanical model (Sung and Wagoner, 2008).

All other tests in this study, except this comparison, use the lubricated fixed roller to represent the industry forming condition.

3.4.3. Normalized Maximum Force Curves

Figure 3.12 shows a typical range of DBF results for various R/t for DP780(D) with \( V_1 = 51\text{mm/s} \), and \( \alpha = 0 \) (a), and \( =0.3 \) (b), respectively. The ordinate is the normalized maximum force \( (F'_{\text{max}}) \), i.e. the maximum force measured in a test divided by the maximum force for the same-sized specimen tested in tension at a strain rate of 0.7/s (the maximum strain rate attainable in tensile tests in our laboratory); The abscissa is the R/t ratio (roller radius to thickness). The dotted curve is a fit line of the results using a function of \( F'_{\text{max}} = A + B(1 - C^{R/t}) \).

As shown in the Figure 3.12(a), \( F'_{\text{max}} \) of most TYPE III fractures is less than those of TYPE I fractures. This phenomenon has been explained by analytical and FE models (Kim et al., 2009b; Hudgins et al., 2010; Kim et al., 2010a). Bending over the roller creates a condition close to plane strain condition, even though the width is narrow compared to the length. The maximum tensile force that a sheet can take during plane-strain bending-under-tension can be significantly reduced (even smaller than the UTS) at small R/t, which results in TYPE III failure (Kim et al., 2010a). Therefore, a critical R/t value can be defined - above which a material fails by tension, and below which a material fails by bending and tension at lower stress than UTS of a material (Hudgins et al., 2010).
Figure 3.12: Normalized max. force vs. R/t: DP780(D)-1.4mm: (a) $\alpha=0$, (b) $\alpha=0.3$. 

<table>
<thead>
<tr>
<th>Type</th>
<th>(R/t)*</th>
</tr>
</thead>
<tbody>
<tr>
<td>I</td>
<td>6</td>
</tr>
<tr>
<td>II</td>
<td></td>
</tr>
<tr>
<td>III</td>
<td></td>
</tr>
</tbody>
</table>
Two critical R/t values, (R/t)*1 and (R/t)*2, can also be found experimentally: (R/t)*2 is defined as the R/t value at which d F'_{\text{max}} /d (R/t) \approx 0, and has been reported as the critical R/t values by many groups (Walp et al., 2006; Hudgins et al., 2007; Shih et al., 2009; Hudgins et al., 2010). However, the value is difficult to be defined quantitatively, especially for \( \alpha = 0.3 \) because F'_{\text{max}} keeps increasing. Recently, a FE simulation result was reported that specimens were fractured as TYPE III at all ranges of R/t under the assumption of plane strain condition, but still there was a R/t value at which the maximum stress did not increase with increase of R/t, which indicated the (R/t)*2 was not so proper measure for shear fracture of AHSS. Therefore, (R/t)*1, which is where the fracture type changes, is used as the critical R/t value and designated as (R/t)* in this study.

The (R/t)* value (a vertical solid line in Figure 3.12) represents the susceptibility to shear fracture. That is, the larger the (R/t)*, the more susceptible to shear fracture the material. (R/t)* was chosen as the closest integer of the mid-point of two rollers having different fracture types. There are only fractures of TYPE I and TYPE III, when \( \alpha = 0 \), because of the limited drawn area on the wall. Therefore, the (R/t)* locates between TYPE III and TYPE I for \( \alpha = 0 \), and between TYPE II and TYPE I for \( \alpha = 0.3 \).

As shown in Table 3.4 and 3.5, (R/t)* varied from 2 to 9 when \( \alpha = 0 \), and from 5 to 16 when \( \alpha = 0.3 \), depending upon the material, and the values of \( V_1 \) and \( \alpha \). To explain the difference of (R/t)* at different \( \alpha \), it is necessary to explain phenomenon occurred in the bent-unbent region (see Figure 3.9 (d)) during sheet flows over the roller. The bent-unbent region was strengthened by strain hardening and softened by deformation-induced heat and thinning. Therefore, if the bent-unbent region is softened as a result of the
competition between those strengthening and softening factors, TYPE II fracture happens; and if it is strengthened, TYPE I happens. The softening becomes smaller as R/t gets larger, therefore, $F'_{\text{max}}$ is increasing as R/t is increasing, Figure 3.12 (b).

Figure 3.13: Comparison of normalized maximum forces of DP steels: (a) $\alpha=0$, (b) $\alpha=0.3$. 

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Figure 3.13 compares \((R/t)^*\) and \(F'_{\text{max}}\) curves of three DP steels. The \(F'_{\text{max}}\) of TYPE III decreased as \(R/t\) decreased, and as the strength of the material increased. The decrease of \(F'_{\text{max}}\) with material strength can be explained by an analytical model based on instability (Hudgins et al., 2010). The maximum applicable tension force of a material, over a roller radius under plan strain condition, was reported as a function of \((R/t)\), \(n\)-value and net longitudinal strain; this results in lower maximum force for a stronger material at smaller \(R/t\) than \((R/t)^*\) (Hudgins et al., 2010).

<table>
<thead>
<tr>
<th>Materials</th>
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<th>13mm/s</th>
<th>2.5mm/s</th>
</tr>
</thead>
<tbody>
<tr>
<td>TRIP780(D)</td>
<td>4</td>
<td>2</td>
<td>2</td>
</tr>
<tr>
<td>DP590(B)</td>
<td>5</td>
<td>4</td>
<td>3</td>
</tr>
<tr>
<td>DP780(D)</td>
<td>6</td>
<td>4</td>
<td>3</td>
</tr>
<tr>
<td>DP980(D)</td>
<td>6</td>
<td>5</td>
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Table 3.4: Summary of formability order based on \((R/t)^*\) when \(\alpha = 0\), for RD.

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<td>12</td>
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<td>N/A</td>
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<tr>
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<td>12</td>
<td>7</td>
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<tr>
<td>DP980(F)</td>
<td>&gt;16</td>
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<td>N/A</td>
</tr>
</tbody>
</table>

Table 3.5: Summary of formability order based on \((R/t)^*\) when \(\alpha = 0.3\), for RD.
3.4.4. Fracture Type Map

With three draw speeds \((\text{V}_1=51, 13, 2.5 \text{ mm/s})\), a fracture type map can be constructed according to \(\text{V}_1\) and \(R/t\). Figure 3.14 shows the fracture type maps of DP590(B) at \(\alpha=0\) and 0.3. The lines on the figures indicate where the fracture type changes. The meaningful line in the figures is the line dividing TYPE III and TYPE I for \(\alpha=0\), and TYPE II and TYPE I for \(\alpha=0.3\), which corresponds to \((R/t)^*\) at different \(\text{V}_1\) for a material. Figure 3.15 compares the fracture type maps of all selected materials in one plot (DP980(E) and DP980(F) were not shown because \(\text{V}_1=51\text{mm/s}\) was the only displacement rate these materials were tested). It can be stated that generally the line moves toward increasing \(R/t\) values as the strength of the material increases. The order of the \((R/t)^*\) is provided in Table 3.4 and 3.5 for both draw-speed ratios, but more roller sets are necessary to establish refined and explicit results for \(\alpha=0.3\).

Interestingly, the \((R/t)^*\) varied with \(\text{V}_1\) for every material, i.e., the higher the value of \(\text{V}_1\), the larger the \((R/t)^*\). This indicates that higher \(\text{V}_1\) leads to a greater susceptibility to shear fracture. Since \(\text{V}_1\) affects to strain rate, fracture type and \(F'_{\text{max}}\) changes were plotted, in Figure 3.16, with the maximum strain rate calculated by Eq. 3.3 for DP590(B). Specifically, the fracture type changed from TYPE I to TYPE III as the maximum strain rate increased, when \(R/t=3.4\) and 4.5.
Figure 3.14: Failure type maps, DP590(B)-1.4mm
Figure 3.15: Comparison of (R/t)* line
There are two parameters which are related to strain rate; strain rate sensitivity and heat caused by plastic deformation. Usually the strain rate sensitivity has a strengthening effect with increasing strain rate; therefore, this result suggests that heat induced by plastic deformation can serve as a probable cause for the change in fracture type. Microstructure damage mechanic/void growth mechanism cannot explain this phenomenon.

![Figure 3.16: F'_max vs. strain rate](image)

**3.4.5. Temperature Change During Draw-Bend Test**

In order to understand the origin of the fracture type change with strain rate found in fracture type maps, temperature change was measured during DBF tests using an infrared camera, FLIR ThermoVision A40. Figure 3.17(a) shows the setup of the FLIR A40 infrared camera with the draw-bend test system. The lens of the infrared camera was
focused in a direction perpendicular to the specimen, at a distance of 0.8 m. Before beginning temperature measurement, the proper value of emissivity of the specimen needed to be determined. A specimen was kept in a bucket of ice cubes for 10 minutes to bring the temperature of the specimen to 0°C; then the temperature of the specimen was measured with the infrared camera, resulting in the emissivity of 0.75.

Figure 3.17 (b) shows one example of thermal images, taken from the infrared camera with emissivity of 0.75, for a TYPE II fracture. The material was DP980(D), at \( V_1 = 51 \text{mm/s}, \alpha = 0.3 \) and \( R/t = 4.4 \). The maximum temperature was significant, up to 93.7°C, at a location near, but not in, the area of localization. It is obvious that the temperature increase in the bending-unbending area, indicated as "a red triangle", was much higher than at other locations. The maximum temperature was ranged from 50-100°C, depending on material, fracture type, and \( V_1 \) during the DBF tests.

It has been reported that this range of temperature increase can reduce the total elongation of DP steels up to around 20% (Sung et al., 2010b). Recently, the authors reported (in other places) that FE simulations of tensile tests and draw-bend formability tests showed the impressive improvements of prediction accuracy by taking the effect of the heat induced by plastic deformation into account using a plastic constitutive equation, H/V model, incorporating the effects of strain hardening, temperature softening, and strain rate sensitivity {Sung, 2010 #1839} {Kim, 2010 #1836}. Accordingly, the major reason for discrepancy between predicted failures and observed failures in DP steels can be concluded as the lack of consideration of deformation-induced heating by using only isothermal FEA and low-speed, isothermal constitutive equations to predict the strains during forming, and by using FLD generated isothermally at very low rates.
3.4.6. **Displacement to Fracture (Uₙ)**

(R/t)* can be a measure of susceptibility to shear fracture of AHSS, but it is limited by the finite number of roller sets. For example, DP780(D) showed same (R/t)*
compared with DP980(D), at $V_1=51\text{mm/s}$ and $\alpha=0$, based on $(R/t)^*$. It was especially difficult to clearly state the formability order based on $(R/t)^*$, at $V_1=51\text{mm/s}$ and $\alpha=0.3$, with the limited roller sets. Therefore, another measure was defined -- front displacement to fracture ($U_f$) -- which is physically more realistic. Because the initial stretching of the specimen depends on specimen alignment in the grips and the "lash", $U_f$ is more precisely defined as “the difference between the displacements of the front grip at the yield strength of each material, compared to the displacement of the front grip at fracture.”.

Figure 3.18: Change of $F'_\text{max}$ and $U_f$ with R/t.

Figure 3.18 shows the $F'_\text{max}$ and $U_f$ changes simultaneously, with various R/t. The measured $U_f$ were fit using the same equation used for $F'_\text{max}$, i.e. $U_f = A + B(1 - C^{R/t})$, and shown with a dotted line in the figure. The $U_f$ changed very similarly with $F'_\text{max}$ as R/t changed. However, $U_f$ for $\alpha=0$ is hard to be compared with that of $\alpha=0.3$. Therefore,
$U_1-U_2$ was compared as shown in Figure 3.19 and Figure 3.20 for various grades of DP and TRIP steels at various R/t, and for four DP980s provided by different suppliers, respectively. $U_1-U_2$ saturates because the fracture type is changing from TYPE III to TYPE I as R/t increases for $\alpha=0$; while $U_1-U_2$ keeps increasing, because the softening caused by thinning and deformation-induced heat decreases, as R/t increases for $\alpha=0.3$. However, $U_1-U_2$ had very similar values for $\alpha=0$ and $\alpha=0.3$ when R/t is large, i.e. when the specimen is fractured as TYPE I. The formability order based on $U_1-U_2$ was DP590(B) > TRIP780(D) > DP780(D) > DP980(D) > DP980(A) > DP980(E) > DP980(F) regardless of $\alpha$. These results agreed with the order of (R/t)* of materials, except TRIP780(D) and DP590(B) for $\alpha=0$; but this measure is more sensitive and explicit than (R/t)*. $U_1-U_2$ at three R/t values (3, 7, and 12) are listed in Table 3.6.

Figure 3.19: Comparison of $U_f$ for experiemental materials.
Figure 3.20: Comparison of $U_f$ with four DP980

3.4.7. Effect of Sheet Orientation

The distribution of martensite in the ferrite matrix of DP steels, and rolling process of sheet metal can cause the difference in formability between rolling direction (RD) and
transverse direction (TD). This is especially true for DP980 steels, which contain a high portion of martensite. Therefore, the formability of four DP980 steels in RD and TD were investigated using DBF tests to study the effect of sheet orientation.

Figure 3.21: Comparison of \( U_f \): RD vs. TD

Figure 3.21 compares \( F'_{\text{max}} \) of DP980(D) in RD and TD. \((R/t)^*\) was same in both directions, but \( F'_{\text{max}} \) in TD and RD were significantly different at small \( R/t \) under \((R/t)^*\) for DP980(D).

Figure 3.22 compares \( U_f \) in RD and TD for (a) DP980(D), and (b) DP980(A). While DP980(A) showed almost identical \( U_f \) in both RD and TD, smaller \( U_f \) in TD were observed for DP980(D). When \( U_f \) is compared between DP980(D) and DP980(A), DP980(D) had the better formability than DP980(A) in RD, however, the formability of DP980(D) was worse than that of DP980(A) in TD for small \( R/t \). Notably, the specimen
of DP980(D) failed immediately after the specimen began the deformation plastically for the smallest roller, R/t=2.2 (shown as "A" in Figure 3.22(a)). Another comparison of $U_f$ of four DP980s was shown in Figure 3.23. $U_f$ in TD and RD were almost same for all DP980s except DP980(D). The formability difference found in DP980(D) cannot be explained by any continuum mechanics because the mechanical properties in RD and TD are almost same, Table 3.2, indicating probable unsuitable microstructure caused by processing problems.

![Comparison of $U_f$: RD vs. TD](image)

Figure 3.22: Comparison of $U_f$: RD vs. TD

In this study the reason for this formability difference of DP980(D) with the sheet orientation was not investigated, but that difference can be another barrier to the wide use of DP steels.
3.4.8. Effect of Edge Condition

The edge splitting has been another big issue in sheet forming of AHSS (Wagoner, 2006). Therefore, it was necessary to investigate how the edge condition can affect to DBF test results. For this purpose, specimens were prepared with three methods: 1) original specimen cut with shear machine and ground with 120 grit emery cloth, 2) another specimen cut with shear machine and milled to 0.5mm from both sheared edges.
and ground by 120 grit emery cloth, and 3) the other specimen cut and milled to 1.5mm from both sheared edge and ground by 120 grit emery cloth.

Figure 3.24 shows (a) force-displacement curves, and (b) $U_f$ of three edge conditions. Three cases showed almost identical results for both comparisons, which addressed there was little edge effect to DBF results. It may be because tension in the edge area is not so significant in the DBF tests. It can be proved by two other works showing that the fracture begins from the middle of width in DBF tests by microscopic observation (see Section 4.2) and DIC technique.

![Figure 3.24: Effect of edge condition: (a) force-displacement curves for various edge conditions, (b) comparison of $U_f$ for various edge conditions.](image)

### 3.4.9. Measurement of Fracture Strain Based on Reduction of Area

Fracture strain can provide important information on the characteristics of fractured samples. The fracture strain can be determined from the reduction of area (RA) measurements. RA and fracture strain can be defined as Eq. 3.5 and Eq. 3.6.
\[ RA = \frac{A_0 - A_f}{A_0} \times 100 \]  
Eq. 3.5

\[ \varepsilon_f = \ln\left(\frac{A_f}{A_0}\right) \]  
Eq. 3.6

where \( A_0 \) and \( A_f \) are the original area and the final area, respectively. \( A_f \) was measured from the fractured section using a Clemax L 1.3C imaging analysis camera installed on an OLYMPUS SZH10 zoom stereo microscope with 3.5\( \times \) zoom. Figure 3.25 shows an example of the section of the fractured samples. \( A_0 \) and \( A_f \) were calculated as follows.

\[ A_0 = W_0 \times t_0 \]  
Eq. 3.7

\[ A_f = \left(\frac{t_{f1} + t_{f2}}{2}\right) \times W_f \]  
Eq. 3.8

where \( W_0 \) and \( W_f \) are the width of the sections of the original and fractured specimen, respectively. \( t_0 \) is the thickness of the original specimen and \( t_{f1} \) and \( t_{f2} \) the thicknesses at the center and one side of the section of the fractured specimen, respectively.

Figure 3.25: Measurement of area reduction from perspective view of section of fractured specimen.
Figure 3.26 (a) and (b) compare the fracture strains of DBF tests at different R/t for DP780(D) and DP980(D), respectively. For reference purpose, the average fracture strains of DP780(D) and DP980(D) were measured from four uniaxial tensile tests and
shown as a dotted line in the figures. The fracture strains are nearly the same for tensile failures and shear failures for all cases, except for DP980(D) tested in the TD direction at small R/t having TYPE III fractures. This result shows that the most DP steels in these strength ranges "shear failure" is not a different phenomenon than "tensile failure", except as influenced by differing mechanical and thermal constraints. That is, plastic localization is the principal failure mechanism for TYPE I, TYPE II and TYPE III failures. However, for DP980(D)-TD, the $\varepsilon_f$ is reduced markedly, to less than half the value in the RD direction. In this case, the limiting factor may be mainly microstructure and related to damage. Because the effect of microstructure is not within the scope of this present study, it will not be examined here.

3.5. Conclusions

A new novel draw-bend formability (DBF) test using displacement rate control of two actuators were developed in order to reproduce shear fracture in a laboratory setting and to understand the fundamental mechanisms of the fracture behavior of dual-phase steels. The following conclusions were reached:

1. The new DBF test guarantees drawing in one direction and failures on the front side of the specimen, thus reproducing industrial failure patterns with more fidelity, accuracy, and reproducibility as compared with back-force controlled draw bend tests.

2. The DBF test was able to reproduce three kinds of fractures for DP steels: TYPE I (tensile failure), TYPE II (transition failure), and TYPE III (shear failure).
3. The type of failure depends on bend ratio \((R/t)\), drawing speed ratio \((V_2/V_1)\), and strain rate (related mainly to \(V_1\) and \(R/t\)). Increased strain rate promotes TYPE III/shear fracture. The detrimental effect of deformation-induced heat is the main cause of this phenomenon. Temperatures up to 100\(^\circ\)C were observed outside of the final plastic localization zone.

4. Two new measures of formability for AHSS were introduced: \((R/t)^*\), representing the transition from tensile to shear failure, and \(\Delta U = U_1 - U_2\), a grip displacement-displacement difference to failure. Both measures can be used effectively to rank the formability of materials and to detect the processing problems, but \(\Delta U\) is more sensitive and physically more meaningful measure than \((R/t)^*\).

5. \((R/t)^*\) varied significantly depending on the drawing speed ratio \((V_2/V_1)\).

6. A lower friction coefficient increases the draw distance, but also the susceptibility to TYPE III fracture.

7. The edge quality of the specimen has little effect on the draw-bend formability. Fracture initiates at the middle of the width of the specimen for TYPE III fractures.

8. Local fracture strains, \(\varepsilon_f\), were the same in tensile and shear failures for all materials and directions except one combination. For that case, DP980(D)-TD, \(\varepsilon_f\) was reduced by more than half when TYPE III failures occurred (low \(R/t\)).
CHAPTER 4: DISCUSSION

NOTE: This chapter is summarizing two works to which the author contributed and collaborated, but for which the author did not have primary responsibility: "Finite Element Simulation of Shear Failure of Advanced High Strength Steels (Kim et al., 2010b)" and "Failure Analysis of Advanced High Strength Steels (AHSS) during Draw Bending (Kim et al., 2009a)". The author does not claim any part of these as a dissertation, but only for helping understand and making a linkage between two contributions in this dissertation: Chapter 2 and Chapter 3.

Figures and Tables are from the references aforementioned, unless others are referenced.

4.1. Draw-Bend Finite Element Analysis with the H/V model

The 1-D plastic constitutive model developed in Chapter 2 (H/V model) was tested with a coupled thermo-mechanical FE model of draw-bend formability (DBF) test; and the results were compared with the experiment data in terms of (R/t)* and U_f. Since the industry is limited to use the solid model due to high computation cost, an alternative, adiabatic model, was proposed.

4.1.1. Coupled Thermo-Mechanical Finite Element Procedure
A thermo-mechanical finite element model of the draw-bend test was developed in order to investigate the failure mechanisms of AHSS deformation, as shown in Figure 4.1. A symmetric (to Z direction) 3D solid model (C3D8RT) was used with 5 layers through thickness in Abaqus 6.7 Standard (ABAQUS, 2007). The mesh size was 0.5mm×1.5mm (x × y direction). A von Mises yield function and the isotropic hardening law were adopted for simplicity in view of the normal anisotropy values near unity for the DP steels. Friction coefficients of 0.06, 0.06, and 0.1 for DP590(B), DP780(D), and DP980(D) (the friction coefficient was based on a comparison of forces of front grip and back grip between FE simulation and DBF test using Coulomb friction law) were used,
respectively, and same thermal coefficients used for tensile test FE simulation in Chapter 2 were used.

The model accounts for deformation heating and heat transfer and is capable of representing softening and altered strain hardening of materials measured at elevated temperatures. Such capability was required because temperatures rise significantly beyond room temperature during draw-bend tests due to deformation heating as shown in DBF experiments.

4.1.2. Failure Types and Temperature

![Figure 4.2: Normalized stress vs. front displacement.](image)

Figure 4.2 compares front displacement-stress curves of thermo-mechanical FE simulation and isothermal FE simulation with that of the DBF test. Thermo-mechanical
FE simulation reproduced the experiment results in higher accuracy including the fracture type, while isothermal FE simulation over-predicted the test with wrong fracture type.

Figure 4.3: Temperature evolution during the draw-bend fracture test

Figure 4.3 compares the measured temperature increase with predicted temperature increase during DBF tests. The FE model predicted the temperature change well within 10 deg. C.

Figure 4.4 shows the deformed shape and temperature distribution at the maximum front force and 90% of the maximum front force for the Type I, Type II, and Type III failures. All three types of failures were predicted accurately (Type I – tensile, Type II – mixed, Type III – shear). And the predicted temperatures were within 10 °C from the measurements.
Figure 4.4: Deformed shape and temperature distribution at the maximum front force and 90% of the maximum front force for the Type I, Type II, and Type III failures for DP590(B) (Wagoner et al., 2009e).

The predicted (R/t)* lines (solid lines) are compared with measured (R/t)* lines (dotted lines) for the three materials in Figure 4.5. The detail measured and predicted failure types are summarized in Table 4.1. 68 cases among total 72 cases were predicted correctly in terms of fracture types. The different predictions were on the transition lines where the fracture type changed for each material, i.e. (R/t)* line, where DBF experiments also showed different results occasionally.
Figure 4.5: The measured and predicted (R/t)* lines (Wagoner et al., 2009e).

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Table 4.1: Comparison of the observed and predicted failure type for $\alpha=0$ (Wagoner et al., 2009e).
4.1.3. Effect of $R/t$

![Graph showing the comparison of measured and simulated maximum front forces for different $R/t$ ratios.](image)

(a) $\alpha = 0$

(b) $\alpha = 0.3$

Figure 4.6: Comparison of the measured and simulated maximum front forces (Wagoner et al., 2009e).
The normalized maximum forces (F'\text{max}) are compared for various R/t ratios in Figure 4.6. The definition of normalized maximum forces is same with the one in the section 2. For the cases with $\alpha = 0$, the FE model predicted the transition of failure type (I→III) and the maximum stress drop as the R/t ratio decreases, which were confirmed by the experiments. For the cases with $\alpha = 0.3$, the FE model predicted the maximum stress drop accurately as the R/t ratio decreases, but the transition of failure type (II→III) was not predicted. The failure type sometimes cannot be clearly identified in simulation, especially near the transition of Type II and III.

The displacements to necking are compared for various R/t ratios in Figure 4.7. The draw distances to necking were predicted in high accuracy within 10 percent of the measured ones where isothermal simulations significantly over-predicted the draw distances to necking.

Measured and predicted U_f are summarized in Table 4.2. In FE simulations, the fracture was decided based on fracture strain measured by reduction area in Chapter 3. There are still over-predictions to some extents using thermo-mechanical FE models, presumably related to other effects such as edge effects and inhomogeneity, not accounted in this model, but it is much less compared with isothermal FE results.
Figure 4.7: Comparison of the measured and simulated displacement to necking.
Table 4.2: Comparison of the measured and predicted $U_f$ for $\alpha = 0$ (unit: mm) (Wagoner et al., 2009e) (Values in the parentheses are predicted ones)

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4.1.4. Application to Industry: Adiabatic Constitutive Material Model

Commercial sheet-forming FE programs usually don't have a thermo-mechanical capability. Also shell models are dominantly used due to computational cost. Therefore, the limitation of a fully coupled thermo-mechanical finite element model is straightforward, even though it may provide accurate failure prediction capability. For computational efficiency at sufficiently high rates, an adiabatic constitutive equation may be used with standard isothermal FEA to approximate the full thermo-mechanical solution.
Figure 4.8: Adiabatic model calculation from H/V model (Wagoner et al., 2009c).

The full constitutive equation can be expressed with Eq. 2.2 and Eq. 2.3 in terms of strain, strain rate and temperature. Assuming no heat transfer in the material, the temperature rise for a uniaxial loading can be calculated from the plastic work. The adiabatic flow curve was obtained by adjusting the flow curve for the temperature rise using the full constitutive equation.

\[ \sigma_{\text{adiabatic}}(\varepsilon, \dot{\varepsilon}) = \sigma(\varepsilon, \dot{\varepsilon}, \Delta T) \]  

Eq. 4.1

The adiabatic flow curve was calculated as shown in Figure 4.8 under assumption that 90% of plastic work changed to heat (as shown in Chapter 2).
Figure 4.9 compares the displacement to failure for the thermo-mechanical, isothermal and adiabatic models. For both tensile and draw-bend FE simulations the displacements to failure of adiabatic simulations were close to those of thermo-mechanical simulations under the similar strain rate which industry experiences, 10/s., whereas isothermal simulations over-predicted the displacements to failure.

Figure 4.10 compares the displacement to failure for the thermo-mechanical, isothermal and adiabatic models. The displacements to failure of adiabatic simulations were close to those of thermo-mechanical simulations, whereas isothermal simulations over-predicted the displacements to failure.
4.2. Failure Analysis of AHSS

To understand the fracture mode, either brittle or ductile, the fractured B-pillar parts provided by A/SP teams and draw-bend fractured specimen tested in the laboratory setting were examined using the SEM (Scanning Electron Microscope) in Edison Welding Institute (EWI) as shown in Figure 4.11 and Figure 4.12.

SEM images of a fractured B-pillar part showed dimpled structures with a number of pores, Figure 4.11. The shear fractured specimen tested with DBF test also showed dimpled structure with many pores at the cross-section on fractured surface. It was found that fracture initiated in the middle of width, not from edges. The pores are a result of micro-void nucleation and coalescence. Although both fractured surfaces showed similar ductile dimple rupture, the draw-bend sample showed more directionality of dimples toward the shear loading direction. Fractured sections of DP steel parts showed
significant numbers of micro-voids. Therefore, the interrupted tensile tests were employed to monitor the growth and coalescence of micro-voids as the plastic stain increased.

![Figure 4.11: SEM results of failure in stretch bending area of B-pillar part (DP980(D), Courtesy of EWI)](image)

**Figure 4.11:** SEM results of failure in stretch bending area of B-pillar part (DP980(D), Courtesy of EWI)

**Figure 4.12:** Draw-bend sample (DP590(B)) and SEM results of fractured cross section (Courtesy of EWI)

### 4.2.1. Interrupted Tensile Test

The interrupted tensile test was conducted to quantify the growth and coalescence of micro-voids as the plastic stain increased as shown in Figure 4.13. Five different points were used to stop the tensile loading from UTS to the final failure point (i.e. 14% stress drop from UTS). The measured engineering stress-strain curves to different stops are
given in Figure 4.14. The cross-section areas of the tested specimens, both initially and at the stop point, are also listed there.

Figure 4.13: Interrupted tensile tests of DP590(B).

Figure 4.14: Engineering stress-strain curves obtained in interrupted tensile tests of DP590(B).
The micrographs of samples taken from the necking or failed areas were further examined in EWI as shown in Figure 4.15. Interestingly voids did not increase significantly up to stress drop of 12% from UTS. The voids and coalescence of voids were observed just before fracture as shown in Figure 4.15.

![Image: Increase of micro-voids from UTS to failure](Courtesy of EWI)

**4.2.2. Characterization of Micro-Void Formation**

The micrographs from interrupted tensile tests were analyzed more quantitatively. The area fraction of micro-voids of each micrograph (Figure 4.16) was calculated by using a commercial image processing and analysis software, Image-J, in EWI. The area fraction of micro-voids is plotted with engineering stress and strain, respectively, in Figure 4.16.

As shown in the Figure 4.16 (b) and (c), the average size and area fraction of micro-voids monotonically increased up to 88% UTS (i.e. 12% stress drop from UTS) and showed a very sharp increase of micro-voids near the failure point (i.e. 86% UTS). Figure 4.16(d) shows that voids are not increasing significantly before fracture. These results prove that damage (nucleation of voids, growth and coalescence) is unlikely to play a significant role in tensile localization and failure for DP steels.
Figure 4.16: Quantitative results of interrupted tests: (a) stop points, (b) average size of voids, (c) the area fractions of micro-voids with respect to stress, and (b) the area fractions of micro-voids with respect to strain.

4.3. **Connection to Chapter 2 and Chapter 3**

In this section, two works were summarized in order to help understand and make the link between Chapter 2 and Chapter 3, which are the two contributions of the candidate in this dissertation.

From Section 4.1, the following can be stated:

- The thermo-mechanical FE model using H/V constitutive law predicted the correct type of failure pattern in most cases (68 cases out of 78 cases for \( \alpha=0 \)) of DBF test without reliance on damage models or other non-continuum effects.
Thermally assisted strain localization has a major effect on the type of failure, when it occurs, and on the discrepancy between low-rate, isothermal FLD measurements and industrial practice. No damage mechanics is required to understand and predict the failure of DP590(B), DP780(D), and DP980(D) for rolling direction tests.

While taking into account thermal effects improves failure prediction greatly, the actual formability is still somewhat lower, presumably related to edge effects and inhomogeneities not accounted for in the modeling.

From Section 4.2, the following can be stated:

- SEM analysis of fractured sections of a B-pillar part and a draw-bend sample showed similar ductile dimple rupture, showing that shear fracture is not brittle fracture.
- Interrupted tensile tests proved that there was the nucleation and growth of micro-voids as plastic strain increased from UTS to fracture point, but the effect was little.
The main conclusion of this dissertation:

For most of the DP steels tested, heating induced by deformation was identified as the dominant effect in producing unpredicted fractures. This is a result of standard industrial techniques that do not take non-isothermal effect into account for simulation of sheet forming and measurement of failure limits. Microstructural damage can also contribute to shear fracture, but it was a secondary factor for all but one of the alloys tested, in one test direction.

Detailed conclusions for Chapter 2 are as follow:

Standard jump tests and isothermal tensile tests of DP590, DP780 and DP980 steels in the uniform strain range at 25, 50 and 100 deg.C have been used to fit constitutive equations from the literature and as proposed in the current work (“H/V Model”). The accuracy of these laws was compared using tensile tests to failure (isothermal and standard) and balanced biaxial bulge tests, and parallel simulations. The following conclusions were reached:

1. The H/V Model provides a natural form to incorporate the transition of strain hardening from power-law-type (Hollomon-like) at low homologous temperatures (and
for many bcc alloys) to saturation-type (Voce-like) at higher homologous temperatures (and for many bcc alloys).

2. The H/V Model, fit to the uniform strain range of tensile data, provides more accurate predictions of large-strain stress-strain behavior than existing models in the literature.

3. The H/V Model, fit to the uniform strain range of tensile data, predicts tensile failure strains more accurately, by factors of 2 to 8, than existing models in the literature.

4. Deformation-induced heating at normal industrial strain rates affects the strain hardening of DP steels significantly, thus promoting strain localization and failure.

5. The accuracy of predicted failure strains and large-strain stress-strain curves using the H/V Model, and the large differences between isothermal and non-isothermal predictions at industrial strain rates, suggest that damage is not a critical factor in the tensile failure of DP steels.

6. The H/V Model can be simplified by setting the linear combination coefficient equal to a constant: $\alpha = 1$ is a power-law/Hollomon model, $\alpha = 0$ is a saturation/Voce model, and $\alpha = \alpha_o$ exhibits a fixed character intermediate between Hollomon and Voce models. The last simplification provides most of the advantages of the full H/V model over the small range of temperatures investigated in the current work.

7. A heat conversion of efficiency of 0.9 was measured for DP590 steel.
8. Tapered tensile specimens with a 2% taper are sufficient to insure failure at the center for DP steels. Parallel and 1%-tapered specimens are not sufficient.

9. DP steels were found to have varying properties in the edge regions of the coils. The central regions were very uniform. The dividing line between the two regions was measured as 300-360mm from the coil edge for DP 590 and DP 980 steels having total coil widths of 1500mm.

10. Many of the common plastic constitutive laws (1-D) incorporating strain, strain rate, and temperature, have been identified and their forms presented. Selected representative ones that incorporate varying strain hardening as a function of temperature have been implemented and compared.

Detailed conclusions for Chapter 3 are as follow:

A new novel draw-bend formability (DBF) test using displacement rate control of two actuators were developed in order to reproduce shear fracture in a laboratory setting and to understand the fundamental mechanisms of the fracture behavior of dual-phase steels. The following conclusions were reached:

1. The new DBF test guarantees drawing in one direction and failures on the front side of the specimen, thus reproducing industrial failure patterns with more fidelity, accuracy, and reproducibility as compared with back-force controlled draw bend tests.

2. The DBF test was able to reproduce three kinds of fractures for DP steels: TYPE I (tensile failure), TYPE II (transition failure), and TYPE III (shear failure).

3. The type of failure depends on bend ratio (R/t), drawing speed ratio (V₂/V₁), and strain rate (related mainly to V₁ and R/t). Increased strain rate promotes TYPE
III/shear fracture. The detrimental effect of deformation-induced heat is the main cause of this phenomenon. Temperatures up to 100°C were observed outside of the final plastic localization zone.

4. Two new measures of formability for AHSS were introduced: \((R/t)^*\), representing the transition from tensile to shear failure, and \(\Delta U = U_1 - U_2\), a grip displacement-displacement difference to failure. Both measures can be used effectively to rank the formability of materials and to detect the processing problems, but \(\Delta U\) is more sensitive and physically more meaningful measure than \((R/t)^*\).

5. \((R/t)^*\) varied significantly depending on the drawing speed ratio \(V_2/V_1\).

6. A lower friction coefficient increases the draw distance, but also the susceptibility to TYPE III fracture.

7. The edge quality of the specimen has little effect on the draw-bend formability. Fracture initiates at the middle of the width of the specimen for TYPE III fractures.

8. Local fracture strains, \(\varepsilon_f\), were the same in tensile and shear failures for all materials and directions except one combination. For that case, DP980(D)-TD, \(\varepsilon_f\) was reduced by more than half when TYPE III failures occurred (low R/t).
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APPENDIX A: INTEGRATED CONSTITUTIVE EQUATIONS

There are many 1-D integrated constitutive equations proposed. This section lists these equations briefly.


\[
\dot{\varepsilon}^p = A \exp\left(-\frac{Q}{RT}\right) \left[\sinh\left(B \frac{\sigma}{s}\right)\right]^{1/m} \tag{A.1}
\]

Evolution equation:

\[
\begin{cases}
\dot{s} = \left\{C\left(1 - \frac{s}{S}\right) \right\}^{D} \text{sign}\left(1 - \frac{s}{S}\right) \dot{\varepsilon}^p \\
S^* = E \left[\frac{\dot{\varepsilon}^p}{A} \exp\left(-\frac{Q}{RT}\right)\right]^p
\end{cases} \tag{A.2}
\]

where \(\dot{\varepsilon}^p\) is the plastic strain rate, \(s\) is an internal variable, deformation resistance, \(R\) is a gas constant, \(T\) is the present temperature (K), and \(A, B, C, D, E, F, Q, m\) are parameters to be found.

A.2. MTS (Mechanical Threshold Stress) model (Kocks, 1976; Mecking and Kocks, 1981; Follansbee and Kocks, 1988)

\[
\sigma = \sigma_s + (\sigma - \sigma_s) \left\{1 - \left[\frac{kT}{\mu b^2 g_0} \ln\left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0}\right)\right]^{1/a}\right\}^{1/b} \tag{A.3}
\]
Evolution equation:

\[
\frac{d\hat{\sigma}}{d\varepsilon} = C \left[ 1 - \frac{\hat{\sigma} - \sigma_a}{\sigma_s(T, \hat{\varepsilon}) - \sigma_a} \right]
\]

\[
\ln \left( \frac{\hat{\varepsilon}}{\hat{\varepsilon}_{s0}} \right) = \frac{\mu b^3 D}{kT} \ln \frac{\sigma_s}{\sigma_{s0}}
\]

where \( \hat{\sigma} \) is an internal variable, the mechanical threshold stress, \( \mu \) is the shear modulus, \( b \) is the magnitude of Burgers vector, \( k \) is the Boltzmann constant, \( T \) is the present temperature (K), \( \hat{\varepsilon} \) is the plastic strain rate, \( \hat{\varepsilon}_{s0} \) is the reference strain rate, \( \sigma_s \) is the saturation stress and \( \sigma_a, \sigma_{s0}, A, B, C, D, \hat{\varepsilon}_{s0}, \sigma_{s0} \) are parameters to be found.

A.3. Modified Bodner and Partom model (Bodner and Partom, 1975; Chen et al., 2008)

\[
\hat{\varepsilon}^p = \frac{2}{\sqrt{3}} \left( \frac{\sigma}{|\sigma|} \right) A \exp \left[ -\frac{(B + 1)}{2B} \left( \frac{Z \exp(CT^\beta)}{\sigma} \right)^{2B} \right]
\]

\[
\begin{cases}
Z = Z_i + (Z_o - Z_i) \exp\left( -D \frac{\sigma d\varepsilon^p}{Z_o} \right) \\
T^* = \frac{T - T_0}{T_m - T_0}
\end{cases}
\]

where \( \hat{\varepsilon}^p \) is the plastic strain rate, \( T \) is the present temperature (K), \( T_m \) is the melting temperature of the material, \( T_0 \) is a reference temperature, and \( A, B, C, D, \beta, Z_o, Z_i \) are parameters to be found.

A.4. Lin-Wagoner model (Lin and Wagoner, 1987)
\[
\sigma = A \left\{ 1 - B \exp[(C_1 + C_2(T - T_0))\varepsilon] \right\} \left( \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right)^m \left( \frac{T}{T_0} \right)^\beta
\]

where \( T \) is the present temperature (K), \( T_0 \) is a reference temperature, \( \dot{\varepsilon} \) is the strain rate, \( \dot{\varepsilon}_0 \) is a reference strain rate, and \( A, B, C_1, C_2, m, \beta \) are material constants.

**A.5. Zerilli-Armstrong model (Zerilli and Ronald, 1987)**

\[
\begin{cases}
BCC: \sigma = A_{BCC} + B_{BCC} \exp[-(\beta_1 - \beta_2 \ln \dot{\varepsilon})T] + C_{BCC} \varepsilon^n \\
FCC: \sigma = A_{FCC} + (B_{FCC} + C_{FCC} \varepsilon^n) \exp(-\beta_1 T + \beta_2 T \ln \dot{\varepsilon})
\end{cases}
\]

where \( T \) is the present temperature (K), \( \dot{\varepsilon} \) is the strain rate, \( A_{BCC}, B_{BCC}, C_{BCC}, \beta_1, \beta_2, n \)
\( (A_{FCC}, B_{FCC}, C_{FCC}, \beta_1, \beta_2, n) \) are parameters to be found.

**A.6. Khan-Huang-Liang model (Khan and Huang, 1992; Khan and Liang, 1999; Khan and Zhang, 2000, 2001; Khan et al., 2004; Khan et al., 2007)**

\[
\sigma = \left\{ A + B \left[ \left( 1 - \frac{\ln(\dot{\varepsilon}_p)}{\ln \dot{\varepsilon}_{\text{max}}} \right) \right]^{n_0} \left( \frac{\dot{\varepsilon}_p}{\dot{\varepsilon}_0} \right)^{n_0} \left( \frac{T - T_0}{T_m - T_0} \right)^\beta \right\} \left( \frac{\dot{\varepsilon}_p}{\dot{\varepsilon}_0} \right)^m \left( \frac{T - T_0}{T_m - T_0} \right)^\beta
\]

where \( \dot{\varepsilon}_p \) is the plastic strain rate, \( \dot{\varepsilon}_0 \) is the reference plastic strain rate, \( \dot{\varepsilon}_{\text{max}} \) is a upper limit strain rate, \( T \) is the present temperature (K), \( T_m \) is the melting temperature of the material, \( T_0 \) is a reference temperature, and \( A, B, n, n_0, m, \beta \) are material constants.

**A.7. Rusinek-Klepaczko model (Klepaczko, 1987; Rusinek and Klepaczko, 2001; Rusinek et al., 2007)**
\[
\sigma = \frac{E(T)}{E_0} \left[ A(\dot{\varepsilon}_p,T)(\dot{\varepsilon}_0 + \varepsilon)^{n(\dot{\varepsilon}_p,T)} + \sigma^*(\dot{\varepsilon}_p,T) \right] \quad (A.10)
\]

\[
\begin{align*}
E(T) &= 1 - \frac{T}{T_m} \exp(T_c(1 - \frac{T_m}{T})) \\
A(\dot{\varepsilon}_p,T) &= A_0 \left[ \frac{T}{T_m} \log(\frac{\dot{\varepsilon}_{\max}}{\dot{\varepsilon}_p}) \right]^{A_i} \\
n(\dot{\varepsilon}_p,T) &= B_0 \left[ 1 - B_1 \frac{T}{T_m} \log(\frac{\dot{\varepsilon}_p}{\dot{\varepsilon}_{\min}}) \right] \\
\sigma^*(\dot{\varepsilon}_p,T) &= C_0 \left[ 1 - C_1 \frac{T}{T_m} \log(\frac{\dot{\varepsilon}_{\max}}{\dot{\varepsilon}_p}) \right]^{m}
\end{align*}
\]  

(A.11)

where \( \dot{\varepsilon}_p \) is the plastic strain rate, \( T \) is the present temperature (K), \( T_m \) is the melting temperature of the material, \( T_c \) is the characteristic homologous temperature, \( \dot{\varepsilon}_{\max} \) is the maximum strain rate, \( \dot{\varepsilon}_{\min} \) is the minimum strain rate, \( \dot{\varepsilon}_p \) is the plastic strain rate, \( E_0 \) is the young's modulus at 0K, and \( \varepsilon_0, T_c, A_0, A_i, B_0, B_1, C_0, C_1, m \) are parameters to be found.
APPENDIX B: COMPOSITE FUNCTIONS

This section covers the strain hardening, strain rate sensitivity, and thermal softening functions.

B.1. Strain hardening functions: \( f(\varepsilon) \)

1) Hollomon (Hollomon, 1945): \( \sigma = K \varepsilon^n \) \hspace{1cm} (B.1)

2) Swift (Swift, 1952): \( \sigma = K (\varepsilon + \varepsilon_0)^n \) \hspace{1cm} (B.2)

3) Ludwik (Ludwik, 1909): \( \sigma = \sigma_0 + K \varepsilon^n \) \hspace{1cm} (B.3)

4) Hartley (Hartley and Srinivasan, 1983): \( \sigma = \sigma_0 + K (\varepsilon + \varepsilon_0)^n \) \hspace{1cm} (B.4)

5) Ludwigson (Ludwigson, 1971): \( \sigma = K_1 \varepsilon^n + \exp(K_2 + n_2 \varepsilon) \) \hspace{1cm} (B.5)

6) Baragar (Baragar, 1987): \( \sigma = \sigma_0 + c \varepsilon^m + d \varepsilon^n + e \varepsilon^{1.2} \) \hspace{1cm} (B.6)

7) Voce (Voce, 1948): \( \sigma = \sigma_0 (1 - A \exp(B \varepsilon)) \) \hspace{1cm} (B.7)

B.2. Strain rate sensitivity functions: \( g(\dot{\varepsilon}) \)

1) Power law model (Kleemola and Ranta-Eskola, 1979; Hosford and Caddell, 1983)

\[
\sigma = \sigma_0 \left( \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right)^m
\] \hspace{1cm} (B.8)

2) Johnson-Cook model (Johnson and Cook, 1983)

\[
\sigma = \sigma_0 \left[ 1 + m \ln \left( \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right) \right]
\] \hspace{1cm} (B.9)
3) Wagoner model (Wagoner, 1981a)

\[
\sigma = \frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \left( \frac{T}{T_0} \right)^m
\]  

(B.10)

**B.3 Thermal softening functions: \( h(T) \)**

1) Linear model (Hutchison, 1963)

\[
h(T) = \sigma_0 (1 - \beta (T - T_0))
\]  

(B.11)

2) Power law model 1 (Zuzin et al., 1964; Misaka and Yoshimoto, 1969)

\[
h(T) = \sigma_0 \left( \frac{T}{T_0} \right) \beta
\]  

(B.12)

3) Power law model 2 (Lubahn and Schnectady, 1947)

\[
h(T) = \sigma_0 a \left( \frac{T}{T_0} \right)
\]  

(B.13)

4) Johnson-Cook model (1983)

\[
h(T) = \sigma_0 \left[ 1 - \left( \frac{T - T_0}{T_m - T_0} \right) \beta \right]
\]  

(B.14)

5) Khan model (Khan et al., 2004)

\[
h(T) = \sigma_0 \left( \frac{T_m - T}{T_m - T_0} \right) \beta
\]  

(B.15)

6) Exponential model 1 (Wada et al., 1978)

\[
h(T) = \sigma_0 a \exp \left( \frac{A}{BT} \right)
\]  

(B.16)

7) Exponential model 2 (Chen et al., 2008)

\[
h(T) = \sigma_0 a \exp \left( C \left( \frac{T - T_0}{T_m - T_0} \right)^m \right)
\]  

(B.17)
APPENDIX C: TEMPERATURE INCREASE OF DP STEELS DURING DEFORMATION

Temperature increase for DP and TRIP steels, generated from plastic work, is much higher than that of typical traditional steels. The temperature increase can be easily calculated in tensile test under assumption of adiabatic condition using Eq. C.1.

\[ \Delta T = \frac{\eta}{\rho C_p} \int \sigma d\varepsilon \]

where \( \eta \) is a fraction of heat conversion from plastic deformation, \( \rho \) is density of the material (7.87×10^{-3} g/mm^3) and \( C_p \) is heat capacity at constant pressure (0.46 J/gK).

Figure C.1: Comparison of true stress-strain curves
If fraction of heat conversion is chosen as 0.9 and stress-strain curves are given as Figure C.1, the temperature increase of various steels can be calculated up to UTS as shown in Figure C.2. As shown in the figure, temperature increase of DP980(D) is about 4 times more than that of DQSK at strain 0.1.

![Figure C.2: Temperature increase with plastic strain on tensile test.](image)

It can be claimed that the temperature increase in uniform elongation may be similar between DP steels and DQSK based on graphical inspection. However, if these materials are used in the bending under tension condition, e.g. draw-bending test, the effective strain over the roller is depending on the roller radius, than the material itself. Therefore, in small radius of roller, the difference of temperature increase between, for example, DP980(D) and AKDQ can be extreme. Temperature increase in draw-bend test can be
predicted by a coupled thermo-mechanical FE model as shown in Figure C.3. Figure C.3 compares temperature increase between DP980(D) and AKDQ in same thickness. As explained above, the effective strain at the maximum force of each test was about 0.5 for R/t=2.2. For both cases, the maximum temperatures were generated over the roller, and temperature of DP980(D) was about 3 times higher than that of AKDQ.

![Figure C.3: Maximum temperature at the maximum force in DBF FE simulation: (a) AKDQ, (b) DP980(D).](image-url)
APPENDIX D: MICROSTRUCTURE OF SELECTED AHSS

The microstructures of three DP steels, DP590(B), DP780(D) and DP980(A) and one TRIP steel, TRIP780(D), were revealed using 4% natal etching and presented in Figure D.1 in cooperation of Edison Welding Institute (EWI).

Figure D.1: Micrographs of AHSS: (a) DP590(B), (b) DP780(D), (c) DP980(A), and (d) TRIP780(D).

The micrographs illustrate that the DP steels contained martensite islands situated at the ferrite grain boundaries. The increased strength of DP980 compared to those of DP780...
and DP590 can be attributed to the greater amount of martensite. Retained austenite and bainite (and martensite) were observed in the ferrite matrix from micrograph of TRIP780(D), Figure D.1(d).

It has been shown that "banding of martensite" in microstructure of DP steels can affect to the anisotropy of formability in DP steels (Dykeman, 2009). In order to understand the big anisotropy of DP980(D) in formability, the microstructures in small magnification (×100) of DP980(A) and DP980(D) were compared as shown in Figure D.2. Unlike what it was expected, DP980(A) showed more explicit bandings of martensite than DP980(D). More microstructure studies would be necessary to find the source of the big formability difference of DP980(D) depending the sheet orientation, but it was not within the scope of the present study.

![Figure D.2: Micrographs of (a) DP980(D), and (b) DP980(A).](image-url)
Figure E.1: High temperature isothermal test device (Piao et al., 2009): (a) installation to MTS-810, (b) detail setup of heat plates on a test sample.
In order to study the effect of temperature on mechanical properties, a series of isothermal tensile tests were conducted at room temperature, 50 deg. C, and 100 deg. C using a special test fixture originally designed for tension-compression testing, Figure E.1 (Boger et al., 2005; Piao et al., 2009).

The fixture was installed to a MTS-810 test frame with 100KN hydraulic grips. In this application, the heated side plates pressed by compressed air against the surface of the specimen (with a constant force of 2.24kN) serve to heat the specimen and to maintain near-isothermal conditions along the length of the contact length. In order to reduce the frictional effect, Teflon sheets were applied between heat plates and a test sample. For measuring strain, a non-contact laser extensometer (Epsilon LE-05) was used. This method (heat plates) has several advantages compared with performing tensile tests in a furnace, including rapid heat-up (about 3 minutes to achieve 250 deg. C), isothermality along the length (at low strain rates), and direct strain measurement using a laser extensometer.

The data from the fixture need to be corrected with Eq. E.1 and Eq. E.2 for friction and bi-axial stresses, respectively, for the side plate contact.

\[
F_{\text{actual}} = F_{\text{raw}} - F_{\text{friction}} = F_{\text{raw}} - 2\mu F_{\text{side}} \tag{E.1}
\]

\[
\bar{\sigma} = \sqrt{\frac{1}{2} \left( (\sigma_1 - \sigma_3)^2 + \sigma_1^2 + \sigma_3^2 \right)} \tag{E.2}
\]

The friction coefficient was determined from the slope of the least-squares line through the measured maximum force vs. four applied side forces (0, 1.12, 2.24 and
3.36kN) as shown in Figure E.2. Friction coefficients for materials obtained in this way appear in Table E.1.

![Graph showing friction force vs side force](image)

Figure E.2: Measurement of friction coefficient in elevated temperature isothermal test device

<table>
<thead>
<tr>
<th>Material</th>
<th>DP590(B)</th>
<th>DP780(D)</th>
<th>TRIP780(D)</th>
<th>DP980(D)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Friction Coeff.</td>
<td>0.06</td>
<td>0.05</td>
<td>0.06</td>
<td>0.06</td>
</tr>
</tbody>
</table>

Table E.1: Friction coefficients of advanced high strength steels

Figure E.3 compares the stress-strain curves adjusted in this manner for a room-temperature test with a side force of 2.24kN and a uniaxial tensile test without side plate contact. The agreement confirms the use of the procedure.
Figure E.3: Comparison of stress-strain curves: Modified curve vs. uniaxial tensile test data

Tensile testing data at various temperatures were shown in Figure E.4. Interestingly DP steels showed decrease of both strength and ductility as temperature increased (UTS decrease of UTS: 3-8%, decrease of total elongation: 5-22%), while TRIP steel demonstrated decrease of strength and increase of ductility (decrease of UTS: 13%, increase of total elongation: 35%). It has been reported that conventional steels decrease the ductility in this temperature range (Lee, 2005).
Figure E.4: Comparison of stress-strain curves: Modified curve vs. uniaxial tensile test data
APPENDIX F: ANALYSIS OF HYDRAULIC BIAXIAL BULGE TEST

Recently, the balanced biaxial bulge test has been increasingly used for measuring the flow stress of a material. A major advantage of the balanced biaxial bulge tests compared to a uniaxial tensile tests is that the stress-strain curve can be obtained to a higher strain which is common in forming processes. In this study, the balanced biaxial bulge test is employed for three DP steels and one TRIP steel in order to study the strain hardening behavior at higher strain by comparing with the constitutive model.

The balanced hydraulic bulge tests were conducted at the Alcoa (ATC, 2008). The opening diameter was 150mm and the die profile radius was 25.4mm. Because of force limits of the ATC hydraulic bulge test system, the materials were machined from one side, from an as-received thickness of 1.4mm for DP590(B) and DP780(D), 1.45mm for DP980(D) and 1.6mm for TRIP780(D) to 0.5mm. A couple of specimens showed warping resulting from the machining, but most were able to be thinned without warping. Tensile tests were conducted with the thinned specimens to compare the material properties before and after machining, and the stress-strain response of thinned materials was within standard deviations of 5 - 10MPa of the original thickness materials depending on the material.

The thinned samples were successfully formed until the specimens were fractured, and it was roughly observed that the bulge height to failure decreased as the UTS increased with exception of TRIP780(D), Figure F.1. TRIP780(D) had bigger fracture
height than DP780 even though UTS of TRIP780(D) was bigger than that of DP780(D). Another rough observation with the samples was that all samples (2 samples per materials) were fractured along with rolling direction.

Figure F.1: AHSS specimens after hydraulic biaxial bulge tests

The stress state near the pole in a bulge test is balanced biaxial tension when a specimen is deformed with hydraulic fluid (Kular and Hillier, 1972) with through-thickness stress negligible for small thickness/bulge diameter ratios (Ranta-Eskola, 1979). In-plane membrane stress, $\sigma_b$, and the magnitude of thickness strain, $\varepsilon_b$, are given by Eq. F.1 and Eq. F.2, respectively.

$$\sigma_b = \frac{pR}{2t} \quad \text{(F.1)}$$

$$\varepsilon_b = 2 \ln\left(\frac{D}{D_0}\right) \quad \text{(F.2)}$$

where $p$ is pressure, $R$ is a radius of bulge, $t$ is current thickness, $D$ is current length of extensometer, and $D_0$ is the initial length of the extensometer, 25.4mm. The radius of curvature $R$ is measured using a spherometer with each leg located at a fixed distance of 21.6mm from the pole. More detailed information for the Alcoa testing machine has appeared elsewhere (Young et al., 1981).
For an isotropic material, the von Mises effective stress and strain are equal to $\sigma_e$ and $\varepsilon_e$, respectively. Figure F.2 are the comparison between biaxial bulge tests and uniaxial tensile tests under assumption of isotropy. The difference between two curves is due to anisotropy of the materials.
When anisotropy is considered, the in-plain membrane stress and the thickness strain must be transformed to effective stress and effective strain using the measured anisotropy. The effect of plastic anisotropy was considered using the ‘79 Hill’s yield criterion (Hill, 1979). The following is a summary of Hill’s 79 theory used for interpreting balanced biaxial bulge test.

**Stress-Strain Relations**

Hill’s 79 yield criterion is:

\[
2(1 + r)\bar{\sigma}^m = (1 + 2r)|\sigma_1 - \sigma_2|^m + |\sigma_1 + \sigma_2|^m
\]  

(F.3)

where \(\bar{\sigma}\) is the yield stress from tensile test, \(r\) is a normal anisotropy and \(m\) is a constant. The strain states of interests in these experiments lie between uniaxial tension and balanced biaxial tension, so \(\sigma_1 \geq \sigma_2 \geq 0\) was assumed. Under this assumption, Eq. (F.3) reduces to:

\[
2(1 + r)\bar{\sigma}^m = (1 + 2r)(\sigma_1 - \sigma_2)^m + (\sigma_1 + \sigma_2)^m
\]  

(F.4)

From plastic normality with assumption of positive and proportional loading,

\[
\frac{d\varepsilon_2}{d\varepsilon_1} = -\left(\frac{d\sigma_2}{d\sigma_1}\right)^{-1} = \frac{\varepsilon_2}{\varepsilon_1}
\]  

(F.5)

By implicit differentiation from (F-4),

\[
\frac{d\sigma_2}{d\sigma_1} = \frac{(1 + 2r)(\sigma_1 - \sigma_2)^{m-1} + (\sigma_1 + \sigma_2)^{m-1}}{(1 + 2r)(\sigma_1 - \sigma_2)^{m-1} - (\sigma_1 + \sigma_2)^{m-1}}
\]  

(F.6)

therefore,

\[
\frac{d\varepsilon_2}{d\varepsilon_1} = \frac{(\sigma_1 + \sigma_2)^{m-1} - (1 + 2r)(\sigma_1 - \sigma_2)^{m-1}}{(\sigma_1 + \sigma_2)^{m-1} + (1 + 2r)(\sigma_1 - \sigma_2)^{m-1}}
\]  

(F.7)
For uniaxial tension ($\sigma_2 = 0$),

$$\frac{\varepsilon_2}{\varepsilon_1} = \frac{-r}{1+r} \quad (F.8)$$

For balanced biaxial tension ($\sigma_1 = \sigma_2 = 0$),

$$\varepsilon_1 = \varepsilon_2 \quad (F.9)$$

**Effective Stress and Strain**

The effective stress and strain can be utilized in order to compare stress and strain state.

With proportional and positive loading under assumption of equivalence of plastic work ($\bar{\sigma} d\bar{\varepsilon} = \sigma_1 d\varepsilon_1 + \sigma_2 d\varepsilon_2$) from (F.10),

$$\bar{\sigma} = \left\{ \frac{1}{2(1+r)} \left[ (1+2r)(\sigma_1 - \sigma_2)^m + (\sigma_1 + \sigma_2)^m \right] \right\}^{1/m} \quad (F.10)$$

$$\bar{\varepsilon} = \left[ \frac{2(1+r)^{1/m}}{2} \left\{ \frac{1}{(1+2r)^{1/(m-1)}} (\varepsilon_1 - \varepsilon_2)^{m/(m-1)} + (\varepsilon_1 + \varepsilon_2)^{m/(m-1)} \right\} \right]^{(m-1)/m} \quad (E.11)$$

For uniaxial tension ($\sigma_2 = 0$),

$$\bar{\sigma} = \sigma_1, \quad \bar{\varepsilon} = \varepsilon_1 \quad (E.12)$$

For balanced biaxial tension ($\sigma_1 = \sigma_2 = 0$),

$$\bar{\sigma} = \frac{2\sigma_1}{[2(1+r)]^{1/m}}, \quad \bar{\varepsilon} = \varepsilon_1[2(1+r)]^{1/m} \quad (E.13)$$

It is well known that the bulge test results have limited accuracy in the low strain range because of the bending effect, which is varying with die geometry and sheet thickness. Sun and Wagoner (Sun and Wagoner, 2008) studied on the bending effect
using 3D solid FE simulation of bulge test for two die opening diameters (100mm and 150mm) and two die profile radii (25.4mm and 2mm).

Figure F.3: Stress relative error according to geometry of bulges (Sun and Wagoner, 2008).

<table>
<thead>
<tr>
<th>Material</th>
<th>DP590(B)</th>
<th>DP780(D)</th>
<th>TRIP780(D)</th>
<th>DP980(D)</th>
</tr>
</thead>
<tbody>
<tr>
<td>r²-value</td>
<td>0.84</td>
<td>0.97</td>
<td>0.96</td>
<td>0.76</td>
</tr>
</tbody>
</table>

Table F.1: Normal anisotropy values of tested materials obtained from thinned sheet.

Figure F.3 shows that the smallest stress differences between input stress and output stress (i.e. the stress calculated from FE bulge simulation) occurred at strain 0.12 for DP590(B) regardless of die opening diameter and die profile radius. The author reported that the smallest stress differences occurred at 0.09 and 0.06 for DP780(D) and DP980(D), respectively, therefore, the bulge test results were fitted to true strain and stress at strains of 0.12, 0.09, 0.09 and 0.06 with Eq. E.13 for DP590(B), DP780(D), TRIP780(D) and DP980(D), respectively.
The normal anisotropies of materials (r-value) are measured from thinned sheet, Table F.1. The value is slightly different with the r-value found in the materials with full thickness.

Figure F.4: Comparison of stress-strain curves: uniaxial tensile results vs. biaxial bulge results (after anisotropy correction) using Hill ’79 yield criterion.
Figure F.4 shows bulge test results after correction of anisotropy. The “a” values found from fitting are shown on each graph for each material. The bulge test results showed good agreements with uniaxial tensile test with much more extended elongation.
APPENDIX G: MESH OF FE MODEL OF TENSILE TEST

Figure G.1: FE model for tensile test: (a) schematic of the model with thermal transfer coefficients, (b) central region of the mesh before deformation, (c) central region of the mesh after deformation.

A thermo-mechanical FE model of a tensile test was developed for evaluating the constitutive models. One-quarter of the specimen was modeled as shown in Figure G.1, as reduced by mirror symmetry in the Y and Z directions. Eight-noded solid elements (C3D8RT) were used for coupled thermo-mechanical FE simulations.

For choosing the mesh size of the FE model, a series of FE simulations were conducted with various size of meshes, Figure G.2. The mesh size did not affect the
engineering stress-strain curves up to uniform elongation, but total elongation, Figure G.2 (a). The mesh size of 0.33mm×0.63mm×0.35mm (X×Y×Z) was chosen based on the computation cost and prediction accuracy, Figure G.2 (a)-(c). The grip was modeled as a rigid body.

Figure G.2: Prediction variations with mesh size in: (a) x-direction, (b) y-direction, (c) z-direction.
APPENDIX H: DETERMINATION OF THERMAL COEFFICIENTS

For thermal-mechanical FE simulation, various thermal constants are required. JMatPro(Sente-Software, 2007) was used for determining these thermal coefficients based on chemical composition with the cooperation of EWI. Figure H.1 shows thermal expansion coefficient (Figure H.1(a)), heat capacity (Figure H.1(b)) and thermal conductivity (Figure H.1(c)) of DP590(B) based on the chemical composition, Table 2.1.
For DP780(D) and DP980(D) steels, the same coefficients were used because the chemical composition of DP780(D) and DP980(D) steels are similar with that of DP590(B).
APPENDIX I: COMPARISON OF EXPERIMENT AND FE SIMULATION IN ISOTHERMAL TENSILE TEST

For the purpose of accessing strain localization and failure, tensile tests were conducted isothermally at three temperatures, 25, 50, and 100deg. C, and were simulated using isothermal FEM with various constitutive models. The difference of total elongation between isothermal FE simulation prediction and isothermal tensile test measurements were found at the three test temperatures, Figure I.1 for DP590(B), Figure I.2 for DP780(D), and Figure I.3 for DP980(D).

Those were summarized in Table I.1. First the errors were expressed at percentage errors for each temperature and for each constitutive model and the absolute values were averaged, Table I.1. The average percentage error between the measured and simulated elongation to failure, $e_f$, is 3-6% for the H/V model vs. 21-29% for the Hollomon ($\alpha$ =1) law, 19-23% for the Voce ($\alpha$ = 0) law, 8-13% for $\alpha(T) = \alpha$, 18-21% for LW and 30-50% for RK.

Consequently, H/V model predicts the development of post-uniform necking over a range of temperatures with significantly better accuracy than other such models.
Figure I.1: Comparison of isothermal tensile test data and FE simulation using selected constitutive models for DP590(B): (a) at 25 deg.C, (b) at 50 deg.C, (c) at 100 deg.C (The percentage error of predicted failure elongations is defined by \( \frac{(e_{f}^{FE} - e_{f}^{Exp})}{e_{f}^{Exp}} \times 100\% \) (\( e_{f}^{FE} \) is the simulated elongation at the time step when the simulated load matches the measured load at \( e_{f}^{Exp} \)).
Figure I.2: Comparison of isothermal tensile test data and FE simulation using selected constitutive models for DP780(D): (a) at 25 deg.C, (b) at 50 deg.C, (c) at 100 deg.C (The percentage error of predicted failure elongations is defined by \((\frac{\varepsilon_{f,FE}-\varepsilon_{f,exp}}{\varepsilon_{f,exp}})\times 100\%) (\varepsilon_{f,FE} is the simulated elongation at the time step when the simulated load matches the measured load at \(\varepsilon_{f,exp}\)).
Figure I.3: Comparison of isothermal tensile test data and FE simulation using selected constitutive models for DP980(D): (a) at 25 deg.C, (b) at 50 deg.C, (c) at 100 deg.C (The percentage error of predicted failure elongations is defined by \( \frac{(e_f^\text{FE}-e_f^\text{Expt})}{e_f^\text{Expt}} \times 100\% \) \( e_f^\text{FE} \) is the simulated elongation at the time step when the simulated load matches the measured load at \( e_f^\text{Expt} \).)
### Table I.1: Difference$^1$ of total elongations of isothermal tensile tests conducted at 25, 50, and 100 deg. C and FE simulations for various constitutive models

<table>
<thead>
<tr>
<th></th>
<th>LW</th>
<th>RK</th>
<th>$α(T)=1$</th>
<th>$α(T)=0$</th>
<th>$α(T)=α$</th>
<th>H/V</th>
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<td></td>
</tr>
<tr>
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<td>12%</td>
<td>-27%</td>
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<td>17%</td>
<td>-23%</td>
<td>-1%</td>
<td>-5%</td>
</tr>
<tr>
<td>100 °C</td>
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<td>2%</td>
</tr>
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<td><strong>Avg</strong></td>
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<td>19%</td>
<td>8%</td>
<td>3%</td>
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<td></td>
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</tr>
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<td>6%</td>
<td>5%</td>
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<td>50%</td>
<td>29%</td>
<td>23%</td>
<td>13%</td>
<td>6%</td>
</tr>
</tbody>
</table>

Table I.1: Difference$^1$ of total elongations of isothermal tensile tests conducted at 25, 50, and 100 deg. C and FE simulations for various constitutive models

---

$^1$ The percentage error of predicted failure elongation is computed by $(\varepsilon_{f,FE} - \varepsilon_{f,exp}) / \varepsilon_{f,exp} \times 100\%$ where $\varepsilon_{f,FE}$ is the predicted elongation at the time step when the predicted load matches the measured load at $\varepsilon_{f,exp}$. The percentage error shown in the table is the average of the absolute values of these percentage errors for three temperatures: 25, 50, and 100 deg. C.
APPENDIX J: DETERMINATION OF MAXIMUM DISPLACEMENT RATE OF DRAW-BEND SYSTEM FOR DP STEELS

Originally the draw-bend system was designed for friction and springback test of Aluminum and conventional steels. The energy required for forming DP and TRIP steels is much higher than that of conventional steels. Therefore, it was necessary to check if the draw-bend system can reach their designed maximum load and speed. For this purpose, the draw-bend system was tested with various grades of DP and TRIP steels with $V_1=130\text{mm/s}$ in order to apply various levels of loads. Figure J.1 shows that $V_1$ drops as the load increases for every case, and, especially, $V_1$ approached to 57mm/s when the load was about 38KN (DP780(D) with 33mm width) even though the set speed was 130mm/s. The system was seized when the load reached to 40KN. Based on these results, the maximum speed of front grip ($V_1$) was set as 51mm/s (2inch/s).
Figure J.1: The change of $V_1$ with displacement of front grip for various grades of steel.
APPENDIX K: THE RESULTS OF DRAW-BEND FORMABILITY (DBF) TEST

For the compactness of dissertation, only limited and important data were shown in the manuscript. In this appendix, the detail data were provided for the reference purpose.

K.1. Normalized Maximum Force Curve

In the initial stage of this research, DBF tests were conducted with one of ductile conventional steels, AKDQ-0.83mm, at small roller radii. The specimens were fractured as TYPE I even at the smallest roller radius, R/t=3.8, for both \( \alpha = 0 \) and 0.3, Figure K.1, indicating that shear fracture is not a critical issue for the conventional steels.

Figure K.2 to K.7 illustrates normalized maximum force curves of tested DP and TRIP steels.

![Figure K.1: Fracture type and maximum stresses at small R/t: AKDQ-0.83mm.](image-url)
Figure K.2: Front normalized maximum force for various R/t: DP590(B).
Figure K.3: Front normalized maximum force for various R/t: DP780(D).
Figure K.4: Front normalized maximum force for various R/t: DP980(D).
Figure K.5: Front normalized maximum force for various R/t: TRIP780(D).
Figure K.6: Front normalized maximum force for various R/t: DP980(E).
Figure K.7: Front normalized maximum force for various R/t: DP980(F).
K.2. Fracture Type Map

Figure K.8 to K.12 illustrate fracture type maps of experimental DP and TRIP steels.

(a) $\alpha = 0$

(b) $\alpha = 0.3$

Figure K.8: Failure type maps, DP590(B).
Figure K.9: Failure type maps, DP780(D).
Figure K.10: Failure type maps, DP980(D).
Figure K.11: Failure type maps, TRIP780(D).
Figure K.12: Failure type maps, DP980(A).
K.3. Effect of Sheet Orientation

DP980 steel is one of the strongest sheet materials among DP steels being researched actively by automotive company and their suppliers. The distribution of martensite in the ferrite matrix of DP steels and rolling process of sheet metal may cause the formability difference between rolling direction (RD) and transverse direction (TD), especially, for DP980 steels which contain a high portion of martensite. Therefore, the formability in RD and TD were compared for four DP980s provided by various suppliers based on $U_f$, Figure K.13-16. All DP980 except DP980(D) did not show the significant formability difference depending on sheet orientation. For DP980(D), the fracture happened as soon as it yields for both $\alpha=0$ and 0.3 as indicated by blue circle in Figure K.13.

Figure K.13: Comparison of $U_f$: DP980(D).
Figure K.14: Comparison of $U_\ell$: DP980(A).

Figure K.15: Comparison of $U_\ell$: DP980(E).
The significant difference of formability in RD and TD as shown for DP980(D) case can give rise to unexpected fracture during stamping processes because forming along only the RD is almost impossible. Therefore, characterization of material along both RD and TD is very important for DP steels to detect a certain processing problem during steel making.

Figure K.16: Comparison of \( U_f \): DP980(F).
APPENDIX L: STRAIN MEASUREMENT OF FRACTURED DBF SAMPLE

A couple of DBF tested samples were analyzed by Digital Image Correlation (DIC) technique in cooperation with GM (Hector, 2009). For taking image, a high speed camera (Phantom V of Vision Research) was set up as shown in Figure L.1 (a). Circle grids (stencil# 3L58271, LECTROETCH) were applied on DBF test samples using VT-45A(LECTROETCH) before the test (For generating high resolution strain contour, several marking methods were tried, but all of them (except etched circle grids) were took off due to high temperature and chock energy generated by fracture). Figure L.1 (b) and (c) show images before and after DBF test. By comparing both image, true strain contour was obtained as shown in Figure L.1 (d). The strain contour showed the localization and fracture began from the middle of the width of the specimen, and the maximum strain was about 0.45. This results prove the RA measurement are correct within measurement error, and explain the effect of edge condition explained in Section 3.4.8.
Figure L.1: Measurement of strain contour using DIC technique: (a) camera setup, (b) initial image before DBF test, (c) final image after DBF test, (d) true strain contour along direction DBF test (Hector, 2009).
APPENDIX M: PLANE STRAIN FRACTURE CRITERIA

Most industry forming is close to plain strain condition because the drawing and stretching occurs over a long cylindrical die radius. Bending over the roller in draw bend test creates a similar condition close to plane strain condition, even though the width is narrow compared to the length. Therefore, the behavior of sheet over the roller can be understood by considering the plane strain condition. Recently, an analytical model considering plane strain condition was proposed for draw-bending test (Kim et al., 2010a).

![Neutral line](image)

Figure M.1: Plane strain bending under tension (Kim et al., 2010a).

At a plane where the radius of curvature is $r$, the normal logarithmic strains in the $\theta$- and $r$-directions are given by

\[
\varepsilon_\theta = \frac{R}{R + r_n} \quad \text{and} \quad \varepsilon_r = \frac{r_n}{R + r_n}
\]
\[ \varepsilon_r = \ln \left( \frac{r}{r_c} \right), \quad \varepsilon_{\theta} = -\ln \left( \frac{r}{r_c} \right) \]  

(M.1)

where \( r_c \) is the radius of curvature of the neutral plane.

Ignoring elasticity and assuming plastic isotropy, the equivalent plastic strain is given by

\[ \bar{\varepsilon} = \sqrt{\frac{2}{3} \varepsilon_r \varepsilon_{\theta}} = \frac{2}{\sqrt{3}} \ln \left( \frac{r}{r_c} \right) \]  

(M.2)

The equilibrium equation is given by

\[ \frac{\partial \sigma_r}{\partial r} = \frac{\sigma_r - \sigma_{\theta}}{r} = \pm \frac{2}{\sqrt{3}} \bar{\sigma} \quad \text{(upper: } r > r_c, \text{ lower: } r < r_c) \]  

(M.3)

where \( \sigma_r \) and \( \sigma_{\theta} \) are the normal stresses in the \( r \)- and \( \theta \)-directions, respectively, and \( \bar{\sigma} \) is the effective stress. The relationship between \( \bar{\sigma} \) and \( \bar{\varepsilon} \) is given by Eq. 4.1 for nonisothermal condition and Eq. 2.2 at 25 deg.C for isothermal condition at a constant strain rate \( (\dot{\varepsilon} = 0.2 \text{ / s}) \). The boundary conditions are \( \sigma_r = 0 \) at the outer surface and \( \sigma_r = -T / R \) at the inner surface where \( T \) is the tensile force and \( R \) is the inner surface radius of curvature. The effect of friction during bending under tension is ignored.

By solving Eq. M.3 numerically, the tensile force is obtained for a given \( r_c \) and \( R \). The necking failure is assumed to occur when the tensile force reaches a maximum, i.e., where the sheet deformation becomes unstable.
Figure M.2: Comparison of the maximum stress obtained using the analytical model and the failure stress obtained using the FE model (Kim et al., 2010b).

Figure M.2 shows stress-based failure criteria (failure stress vs. R/t) were obtained using the finite element and analytical procedures in plane strain condition. Two cases were calculated: isothermal and nonisothermal cases. The failure criterion obtained using the finite element procedure (rate insensitive, frictionless) agreed well with that obtained using the analytical procedure. It was found that the maximum stress that a sheet could take during plane-strain bending-under-tension could be significantly reduced at small R/t, which results in TYPE III failure. Therefore, a critical R/t value can be defined - above which a material fails by tension, and below which a material fails by bending and tension at lower stress than UTS of a material.