CORRELATION OF MICROSTRUCTURE, TENSILE PROPERTIES
AND HOLE EXPANSION RATIO IN COLD ROLLED
ADVANCED HIGH STRENGTH STEELS

by

Oscar R. Terrazas
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Golden, Colorado

Date ______________________________

Signed: __________________________________

Oscar R. Terrazas

Signed: __________________________________

Dr. Chester J. Van Tyne
Thesis Advisor

Signed: __________________________________

Dr. Kip O. Findley
Thesis Advisor

Golden, Colorado

Date ______________________________

Signed: __________________________________

Dr. Ivar Reimanis
Professor and Interim Head
Department of Metallurgical and Materials Engineering
ABSTRACT

The demand for advanced high strength steels (AHSS) with higher strengths is increasing in the automotive industry. While there have been major improvements recently in the trade-off between ductility and strength, sheared-edge formability of AHSS remains a critical issue. AHSS sheets exhibit cracking during stamping and forming operations below the predictions of forming limits. It has become important to understand the correlation between microstructure and sheared edge formability. The present work investigates the effects of shearing conditions, microstructure, and tensile properties on sheared edge formability. Seven commercially produced steels with tensile strengths of 1000 ± 100 MPa were evaluated: five dual-phase (DP) steels with different compositions and varying microstructural features, one trip-aided bainitic ferrite (TBF) steel, and one press-hardened steel tempered to a tensile strength within the desired range.

It was found that sheared edge formability is influenced by the martensite in DP steels. Quantitative stereology measurements provided results that showed martensite size and distribution affect hole expansion ratio (HER). The overall trend is that HER increases with more evenly dispersed martensite throughout the microstructure. This microstructure involves a combination of martensite size, contiguity, mean free distance, and number of colonies per unit area. Additionally, shear face characterization showed that the fracture and burr region affect HER. The HER decreases with increasing size of fracture and burr region. With a larger fracture and burr region more defects and/or micro-cracks will be present on the shear surface. This larger fracture region on the shear face facilitates cracking in sheared edge formability. Finally, the sheared edge formability is directly correlated to true fracture strain (TFS). The true fracture strain from tensile samples correlates to the HER values. HER increases with increasing true fracture strain.
TABLE OF CONTENTS

ABSTRACT ................................................................................................................................ iii
LIST OF FIGURES ............................................................................................................................ vi
LIST OF TABLES ............................................................................................................................. xi
ACKNOWLEDGEMENTS ................................................................................................................. xii
CHAPTER 1: INTRODUCTION....................................................................................................... 1
CHAPTER 2: LITERATURE REVIEW ........................................................................................ 3
  2.1 Steel Characteristics .............................................................................................................. 3
    2.1.1 Dual Phase Steels ............................................................................................................ 3
    2.1.2 Trip-aided Bainitic Ferrite Steels .................................................................................. 6
    2.1.3 Press Hardened Steels ................................................................................................. 7
  2.2 Hole Expansion ................................................................................................................... 8
  2.3 Sheared Edge Formability ................................................................................................. 9
    2.3.1 Shearing Process ......................................................................................................... 10
    2.3.2 Influence of the Shear Affected Zone ........................................................................ 13
    2.3.3 Hole Expansion of Various AHSS .............................................................................. 17
    2.3.4 Correlation of HER and Tensile Properties .............................................................. 18
  2.4 Angular Stretch Bend ....................................................................................................... 22
CHAPTER 3: EXPERIMENTAL PROCEDURE ........................................................................ 24
  3.1 Material Selection .............................................................................................................. 24
    3.1.1 Tempering Study for 22MnB5 ...................................................................................... 25
  3.2 Material Characterization ................................................................................................. 25
    3.2.1 Volume Fraction Analysis .......................................................................................... 25
    3.2.2 X-ray Diffraction ....................................................................................................... 26
    3.2.3 Quantitative Stereology ............................................................................................ 27
  3.3 Sheared Edge Characterization ......................................................................................... 29
    3.3.1 Micro Hardness ......................................................................................................... 30
  3.4 Tensile Testing .................................................................................................................. 31
  3.5 Hole Expansion Testing ................................................................................................... 32
    3.5.1 Hole Expansion Crack Analysis ............................................................................... 33
  3.6 Angular Stretch Bend Testing ........................................................................................ 34
CHAPTER 4: RESULTS

4.1 Material Selection

4.1.1 Tempering Study Results

4.2 Microstructural Characterization

4.2.1 Volume Fraction Analysis

4.2.2 X-ray Diffraction

4.2.3 Quantitative Stereology

4.3 Sheared Edge Characterization

4.3.1 Shear Affected Zone (SAZ)

4.4 Tensile Properties

4.5 Hole Expansion

4.6 Angular Stretch Bend

CHAPTER 5: DISCUSSION

5.1 Influence of Microstructure on Sheared-Edge Formability

5.1.1 Influence of Microstructure on Tensile Properties

5.2 Influence of Sheared Edge Conditions on Sheared Edge Formability

5.3 Correlation Between Sheared Edge Formability and Tensile Properties

5.4 Correlation of Angular Stretch Bend Results to HER and Tensile Properties

5.5 Discussion Summary

CHAPTER 6: SUMMARY AND CONCLUSIONS

CHAPTER 7: FUTURE WORK

REFERENCES CITED

APPENDIX A: PRELIMINARY HOLE EXPANSION STUDY

APPENDIX B: INFLUENCE OF SHEAR AFFECTED ZONE ON ANGULAR STRETCH BEND TEST

APPENDIX C: INCONCLUSIVE CORRELATIONS
LIST OF FIGURES

Figure 2.1  SEM micrograph of a laboratory-produced DP steel. The larger, darker regions are ferrite and the smaller, lighter regions are martensite. Etched with 2% nital [5] .......... 4

Figure 2.2  Time versus temperature plots for the two primary processing methods to produce DP steels: (a) intercritically annealed and (b) as-hot-rolled [6]................................. 5

Figure 2.3  SEM micrograph of a commercially produced TBF steel. Etched with 2% nital. (Color Image see PDF Copy) ........................................................................................ 6

Figure 2.4  Time versus temperature plot for the primary processing method to produce TBF steels [7].......................................................................................................................... 7

Figure 2.5  Schematics of the (a) direct and (b) indirect hot stamping processing methods used to produce press hardened steels [10] ........................................................................... 8

Figure 2.6  Visual representation of hole expansion test samples. A flat-bottom punch was used on the left specimen and a conical punch was used on the right specimen. The crack observed in the conical punch specimen shows it was tested beyond failure [12] ....... 9

Figure 2.7  HER versus UTS for a variety of steel grades. Increasing ultimate tensile strength is associated with lower hole expansion ratios. Adapted from experimental work by Sadagopan et al. [15]. .................................................................................................... 10

Figure 2.8  Visual representation of the tooling used during the shearing process [17] ......... 11

Figure 2.9  Schematic illustrating steps in the shearing process. a) the rollover phase of shearing, b) an expanded view of rollover with corresponding flow lines, c) the burnishing phase of shearing, d) expanded view of burnishing with corresponding flow lines e) the fracture phase of shearing, and f) an expanded view of the initiation of fracture and location of the burr [17] ........................................................................... 12

Figure 2.10 Visual representation of sheared edge showing the four deformation regions. Roll-over is part of the free portion of the sheet that deformed plastically, burnish is the penetration of the punch in a vertical manner, the fracture region is where the crack initiated and propagated, and the burr is a protrusion of metal from the edge [18]..... 13

Figure 2.11 Average micro hardness profiles for DP600 and DP780, showing the depth of the work hardening effects from the SAZ. Adapted from experimental work performed by N. Pathak et al. [11] ................................................................................................................. 14

Figure 2.12 Average HER for a DP600 steel with various hole edge conditions. Adapted from experimental work performed by N. Pathak et al. [11].................................................... 15

Figure 2.13 Equivalent strains at failure obtained from hole expansion tests for boron steels. Plot from the experimental work by Butcher et al. [12] .................................................. 16
Figure 2.14  Circumferential strain versus radial strain plot showing the experimental strain path compared to the strain path from the simulation with and without the shear affected zone using the equivalent strain measurement methods [20]................................................. 17

Figure 2.15  Relationship between YS/UTS ratio and HER. Plot adapted from the experimental work by Xinyan et al. [27]........................................................................................... 19

Figure 2.16  Correlation between post-uniform elongation and hole expansion ratio. Adapted from experimental work by S. Sadagopan et al. [15] and re-plotted by S.K. Paul [30]............................................................................................................... 21

Figure 2.17  Correlation between plastic anisotropy and hole expansion ratio. Adapted from experimental work by S. Sadagopan et al. [15] and re-plotted by S.K. Paul [30].......... 21

Figure 2.18  Typical observed failure locations for the smallest and largest R/t ratio on angular stretch bend test samples [15]............................................................................................... 22

Figure 2.19  Height of failure as a function of R/t ratio. Most grades have a critical R/t ratio which causes transition from punch radius failure to sidewall failure [15]. (Color Image see PDF Copy).................................................................................................. 23

Figure 3.1  Sample XRD pattern for the TBF steel. The calculated refinement is overlaid on the experimental pattern. The difference between calculated and experimental results is shown below. (Color Image see PDF Copy)................................................ 27

Figure 3.2  Representative micrograph of (a) circle grid of known dimensions used for quantitative stereology analysis, and (b) definition of a martensite colony. (Color Image see PDF Copy)........................................................................................................ 29

Figure 3.3  Sample image of the sheared edge face of DP600 steel. Size of the different regions is represented as a fraction of the sheet thickness........................................................ 30

Figure 3.4  Schematic of the location of the micro hardness profiles .............................................. 31

Figure 3.5  Sample micro hardness profiles for DP600 steel measured in the fracture region of the sheared edge........................................................................................................... 31

Figure 3.6  Photograph of the punch tooling configuration on the Interlaken formability press. The punch remains stationary, while the bottom part of the tooling on the crosshead moves up with the sample. (Color Image see PDF Copy)....................... 33

Figure 3.7  Angular stretch bend set-up. (Color Image see PDF Copy)........................................ 34

Figure 4.1  Predicted tensile strength as a function of tempering temperature used to determine the desired heat treatment for 22MnB5 steel .......................................................... 36
Figure 4.2  LOM micrographs of DP980 LY (a), DP980 LY LSi (b), DP980 HY (c),
DP980 LCE (d), DP1180 (e), and TBF980 (f). Etched with 2 pct. nital. (*Color
Image see PDF Copy*) ................................................................. 37
Figure 4.3  LOM micrographs of 22MnB5 prior to heat treatment (a) and post-heat
treatment (b). Etched with 2 pct. nital.................................................. 38
Figure 4.4  SEM micrograph of TBF980 at 2500X magnification. Etched with 2 pct. nital.
(*Color Image see PDF Copy*) .......................................................... 39
Figure 4.5  XRD patterns of the TBF steel. Both as-received and after testing patterns are
shown. (*Color Image see PDF Copy*) .................................................... 40
Figure 4.6  Rollover, burnish, and fracture + burr region sizes as a function of sheet thickness... 41
Figure 4.7  Micro hardness profiles for (a) DP980 A, (b) DP980 B, (c) DP980 C, (d) DP980 D,
(e) TBF980, (f) DP1180, and (g) 22MnB5 measured in the fracture and burr region of the
sheared edge.................................................................................. 42
Figure 4.8  Representative engineering stress vs. engineering strain curves of a single test
specimen for all steels. (*Color Image see PDF Copy*).............................. 45
Figure 4.9  Average hole expansion ratios and standard deviations for all steels.................. 46
Figure 4.10  Representative images of hole expansion samples as viewed from the top with
multiple cracks. Images shown are (a) TBF980 and (b) DP980 A .................... 47
Figure 4.11  Representative images of hole expansion samples as viewed from the top with
partial cracks. Images shown are (a) DP980 A and (b) DP980 C...................... 47
Figure 4.12  Representative images of cracks propagating through the microstructure of hole
expansion samples. Images shown are for (a) DP1180 and (b) DP980 D ............... 48
Figure 4.13  Relationship between (a) R/t ratio and height at failure, and between (b) R/t ratio
and reduction in area for all experimental steels............................................. 49
Figure 5.1  Hole expansion ratio as a function of (a) martensite volume fraction and (b) carbon
content of martensite................................................................. 50
Figure 5.2  Hole expansion ratio as a function of (a) contiguity of martensite, (b) mean free
distance between martensite colonies, and (c) colony size of martensite for DP
steels ......................................................................................... 51
Figure 5.3  Hole expansion ratio as a function of the product of colony size, contiguity, and
mean free distance of martensite for DP steels ............................................. 52
Figure 5.4  Hole expansion ratio as a function of number of martensite colonies per unit area.... 53
Figure A.5  Relationship between (a) HER and YS/UTS and between (b) HER and post-uniform elongation for the preliminary set of steels ....................................................... 80

Figure A.6  Relationship between R/t ratio and height at failure for DP and TRIP steels. Open symbols represent samples that fractured at the drawbead ........................................... 81

Figure A.7  Relationship between R/t ratio and reduction in area for DP and TRIP steels ........ 81

Figure B.1  Average micro hardness profiles for DP600 (a) milled edge and (b) sheared edge conditions ..................................................................................................................... 83

Figure B.2  Representative load vs. displacement curves of a single test specimen for milled and sheared edge conditions. Specimens were tested with the 1.0 mm radius .......... 84

Figure B.3  Representative milled edge specimen tested with a 5.0 mm punch. (a) Top view and (b) side view. (Color Image see PDF Copy)................................................................. 85

Figure B.4  Representative sheared edge specimen tested with a 1.0 mm punch. (a) Top view and (b) side view. (Color Image see PDF Copy)................................................................. 85

Figure C.1  HER as a function of (a) product of grain size and mean free distance, (b) mean free distance to martensite volume fraction ratio, (c) martensite volume fraction to contiguity ratio, and (d) product of true fracture strain and contiguity...................... 86

Figure C.2  HER as a function of (a) n-value, (b) yield strength, (c) relative depth of SAZ, and (d) total elongation........................................................................................................ 87
LIST OF TABLES

Table 2.1  Typical Chemical Composition of 22MnB5 Steels (wt. pct.) [10] ............................... 7
Table 3.1  Chemical Compositions of Experimental Steels (wt. pct.) ......................................... 25
Table 4.1  Volume Fraction of Micro Constituents in Steels for Future Work (in wt. pct.) ......... 38
Table 4.2  Summary of Quantitative Stereology Results ............................................................ 40
Table 4.3  Summary of Shear Face Region Analysis as Percentage of Thickness ...................... 40
Table 4.4  Depth, Magnitude, and Relative Depth of the Shear Affected Zone ......................... 44
Table 4.5  Summary of Tensile Properties for Experimental Steels ............................................ 44
Table 4.6  Summary of HER for Experimental Steels ................................................................. 46
Table 5.1  Correlation Coefficients between HER and Quantitative Stereology Measurements 54
Table 5.2  Correlation Coefficients between HER and Tensile Properties ................................. 64
Table A.1  Chemistry of the Steels Used for the Preliminary Study (wt. pct.) ............................ 74
Table A.2  Size and Relative Depth of the Shear Affected Zone ............................................... 77
Table A.3  Tensile Properties for Preliminary Steels ................................................................. 77
Table A.4  Comparison of Various Work-Hardening Measurements ....................................... 78
Table A.5  HER of the Steels Used for the Preliminary Study .................................................... 79
Table B.1  Summary of ASB Results for DP600 ...................................................................... 84
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CHAPTER 1: INTRODUCTION

The use of advanced high strength steel (AHSS) is increasing in the automotive industry due to the fact that these steels meet the demands of higher strength while maintaining good formability. AHSS consists of various microstructural constituents such as ferrite, retained austenite, bainite, and martensite, which properly controlled can achieve favorable strength and ductility properties [1]. Formability of these steels is essential, especially when stamping of complex geometry parts is involved. Some automotive components have very tight geometrical tolerances, and the chance of higher springback and failure are greater with higher strength steels.

Many automotive stampings are formed from sheet steel that has been sheared from a larger coil. For these sheets there is a sheared edge which is formed during the shearing process and then goes through the stamping process. Traditional formability prediction methods such as forming limit diagrams (FLDs) serve as a useful guide for general formability but are often inaccurate in predicting sheared edge formability of AHSS. The common forming limit curve does not account for edge cracking, which occurs at lower strains than those normally predicted by FLD [2]. Some of the factors influencing sheared-edge formability include material microstructure, edge condition, stamping parameters, and forming parameters. Accurate understanding and measurement of the sheared-edge formability is currently one of the major challenges encountered by the sheet steel stamping industry.

The current study was initiated with the following objectives: a) to analyze the effect of microstructure of various AHSS on hole expansion ratio (HER), b) compare HER to mechanical properties obtained from tensile tests and, c) identify correlations between microstructure, tensile properties, and sheared edge formability. The scope of this project consisted of assessing and performing hole expansion tests on dual phase (DP), TRIP-aided bainitic ferrite (TBF) and tempered press-hardened steels, as well as tensile testing and microstructural analysis. The emphasis was on DP steels in order to provide a more focused set of results.

It is important to understand the correlation between microstructure and sheared-edge formability, with the goal of decreasing premature cracking and failure of complex parts when
stamped from a sheared sheet steel. Understanding how the size, morphology, volume fraction, and distribution of the micro-constituents that are present in the microstructure affect sheared-edge formability will provide insight into structure-property relationships for sheared edge formability.

The following chapter provides background on DP, TBF and press-hardened steels, as well as a background on sheared edge formability and sheared edge testing. The subsequent chapters describe the experimental methods used for this thesis work, followed by results of the material characterization, tensile properties, and sheared edge formability tests. A discussion of the results is presented, followed by a summary and conclusions of the findings. Future work recommendations are presented in the final chapter.
CHAPTER 2: LITERATURE REVIEW

Development of advanced high strength steel (AHSS) for automotive components is a continuing requirement in order to improve safety and reduce weight to improve fuel efficiency of vehicles. Automotive companies demand steels that can be shaped by various forming methods and with material tensile strengths greater than 1 GPa. Higher strength steels generally have poor formability compared to lower strength steels, especially during the stretching of sheared edges. Poor formability can result in problems during stamping or during assembly of vehicle parts and components. Forming limit diagrams (FLDs) are used to estimate the amount of deformation that can be imparted to a sheet metal before failure occurs during a forming operation. Typically, sheared edge formability is less than the FLD prediction.

The use of higher strength AHSS had been limited to automotive parts with simple geometries (such as door impact beams or bumper reinforcement), but lately they have been used in more complex geometries (such as seat frames or B-pillars), where ductility and sheared edge formability are important considerations [3]. While there have been major improvements recently in the trade-off between ductility and strength, sheared-edge stretching of AHSS remains a critical issue.

The present chapter reviews the microstructure and properties of steels in the same classes as the experimental steels, sheared edge formability, measurements of sheared edge formability, and the correlation of microstructure and mechanical properties to hole expansion ratio. This background information directed the development for the current study.

2.1 Steel Characteristics

2.1.1 Dual Phase Steels

Dual phase steels consist of a microstructure of ferrite and martensite, with varied volume fractions of each micro-constituent. Figure 2.1 shows a light optical microscopy (LOM) micrograph of an example microstructure for DP steels. The dispersion of martensite islands within a ferrite matrix provides a good combination of strength and ductility for dual phase
steels [4]. This combination of properties makes DP steels more suitable for applications where high formability is required. DP steels have relatively low alloying requirements compared to other AHSS, as well as the capability of being produced by multiple processing methods. The two primary methods to produce DP steels are hot-rolled and continuously annealed versus cold-rolled and intercritically annealed. Figure 2.2 shows the time-temperature profiles for these two different processing methods. In the hot-rolled and continuously annealed condition, the cooling profile is controlled based on alloying content, where a specific austenite fraction transforms to ferrite and the remaining austenite transforms to martensite. In the cold-rolled and intercritically annealed condition, a cold-rolled steel is initially heated to a temperature within the intercritical phase region (between \( A_1 \) and \( A_3 \)) where both austenite and ferrite are present. The steel is then cooled at a specific cooling rate to obtain the desired fractions of ferrite and martensite [4].

Figure 2.1  SEM micrograph of a laboratory-produced DP steel. The larger, darker regions are ferrite and the smaller, lighter regions are martensite. Etched with 2% nital [5].
The mechanical properties of DP steels are dominated by the volume fraction, morphology, and distribution of the martensite in the ferrite matrix. Factors affecting martensite volume fraction include intercritical temperature, cooling rate, and alloy content. A higher intercritical temperature results in greater austenite fractions, which transform to martensite upon rapid cooling [5]. Alloying elements can affect austenite stability. For example, manganese and nickel increase austenite stability and increase hardenability, allowing more of the austenite to transform into martensite at slower cooling rates.

Martensite morphology has been shown to have an effect on mechanical properties and failure behavior of DP steels [7]. There are various morphologies for martensite such as needle-like, granular, and equiaxed. All of these three morphologies result in increasing strength with increasing volume fraction of martensite. Aspects of the morphology can have preferential effects on the strength and ductility of DP steels. Bag et al. showed that long ribbons of martensite, separated by ferrite, had the greatest strength in the rolling direction at the expense of ductility and this morphology was achieved by heat treatments that ranged from 20 hours to 5 days followed by warm rolling. Decreasing martensite aspect ratios results in increased ductility at the expense of strength. An evenly distributed array of equiaxed martensite islands normally results in the best combination of strength and ductility for DP steels [7].
2.1.2 Trip-aided Bainitic Ferrite Steels

Trip-aided bainitic ferrite steels consist of a microstructure of ferrite, bainite, and retained austenite, with varied amounts of each micro-constituent. During deformation, the retained austenite undergoes phase transformation due to a transformation-induced plasticity (TRIP) effect. Figure 2.3 shows an example SEM micrograph for TBF steels. The dispersion of retained austenite islands that undergo the TRIP effect in a ferritic-bainitic matrix provides the enhanced strength and formability that characterize TBF steels [8].

Figure 2.3 SEM micrograph of a commercially produced TBF steel. Etched with 2% nital. (Color Image see PDF Copy).

TBF steels are produced primarily by using an interrupted ausforming process, similar to a one-step quench and partitioning (Q&P) type processing method. Figure 2.4 shows the time-temperature profile for this processing method. It consists of an austenitization step followed by a controlled quench down to the bainitic region, which is followed by an austempering hold, then oil quench to room temperature. This processing method results in a microstructure of ferrite, bainite, and retained austenite. However, variations have been explored recently where heating, holding, and cooling steps are slightly altered to optimize TBF steels. For example, Hausmann et al. quenched to the martensitic region and followed by an isothermal hold within that region which resulted in a microstructure of ferrite, bainite, martensite, and retained austenite [9]. This process results in a higher strength TBF steel.
2.1.3 Press Hardened Steels

Press hardened steels consist of a microstructure of martensite. This microstructure is achieved after the hot stamping process. Currently, most of the press hardened steels used in the automotive industry are 22MnB5 or similar grade, with a chemical composition very similar to the one shown in Table 2.1 [10]. Initially, the hot stamping process was not developed with the purpose to manufacture automotive steels. It was first used in manufacturing various blades but was later recognized as a potential processing method for automotive structural components.

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>Ti</th>
<th>Nb</th>
<th>V</th>
<th>Al</th>
<th>N</th>
<th>S</th>
<th>P</th>
</tr>
</thead>
<tbody>
<tr>
<td>22MnB5</td>
<td>0.2</td>
<td>1.18</td>
<td>0.22</td>
<td>0.12</td>
<td>0.16</td>
<td>N/R</td>
<td>0.04</td>
<td>N/R</td>
<td>0.03</td>
<td>0.005</td>
<td>≤0.005</td>
<td>≤0.03</td>
<td></td>
</tr>
</tbody>
</table>

The hot stamping process essentially makes use of the formability of austenite at high temperatures, stamping or forming into the actual component dimension during that step, then transforming the steel to martensite upon quenching. 22MnB5 steel grades contain boron, which increases hardenability, and titanium, which helps prevent boron precipitating with nitrogen. The as-received material can be hot-rolled or cold-rolled, and can be uncoated or coated with an aluminum-silicon or zinc layer. It commonly has a tensile strength of 600-800 MPa range prior to hot stamping and tensile strength of 1300-1800 MPa after hot stamping. The hot stamping
process can be done directly or indirectly on blanks from the as-received material. Figure 2.5 shows a process map for the direct and indirect hot stamping processes. The direct stamping process involves austenitization of the blank, transfer of the austenitized blank to the forming die, pressing or forming of the blank and immediate quenching of the part inside the die. The indirect hot stamping process goes through the same steps except there is an additional step at the beginning prior to austenitization where there is a cold pre-forming step.

Figure 2.5 Schematics of the (a) direct and (b) indirect hot stamping processing methods used to produce press hardened steels [10].

### 2.2 Hole Expansion

Stretch flangeability and sheared-edge stretching are usually evaluated by the hole expansion test. The hole expansion ratio (HER) is calculated from a hole expansion test by

\[
\text{Hole Expansion Ratio (\%)} = \left( \frac{D_f - D_o}{D_o} \right) \times 100
\]

where \(D_o\) is the initial diameter of the hole and \(D_f\) is the hole diameter when a through-thickness crack is first observed. Figure 2.6 provides a visual representation of common hole expansion samples after testing. Flat-bottom, hemispherical and conical punch set-ups are methods used for hole expansion testing [11].

When stretching a circular hole produced by shearing with a flat-bottom punch, it is assumed that deformation at the very edge of the hole follows a uniaxial tensile strain path. The principal stress component directions in this type of hole-expansion test are in the circumferential, radial, and thickness directions. When a conical punch is used for the test, the deformation is more complex since the sheet edge undergoes bending as well as stretching. When the punch moves up from the bottom, the upper side of the sheet experiences more
circumferential strain as compared to the side of the sheet that is in contact with the conical punch.

Figure 2.6 Visual representation of hole expansion test samples. A flat-bottom punch was used on the left specimen and a conical punch was used on the right specimen. The crack observed in the conical punch specimen shows it was tested beyond failure [12].

Sample position during testing affects the results. Once the hole is punched, there are different damage regions created on the punched face. These regions have different amounts of strain and affect crack initiation and propagation [13]. Details will be provided in the following section.

2.3 Sheared Edge Formability

Multiple microstructural properties influence the sheared-edge formability of AHSS. Most AHSS grades have microstructures with different volume fractions of soft/hard constituents, as well as different morphologies and dispersions of these constituents. The multiphase nature of AHSS produces a good combination of strength and ductility and has a strong influence on the mechanical properties, failure mechanisms, and sheared edge formability.

In addition to the material microstructure, work hardening also influences sheared edge formability. A low amount of work hardening effect minimizes the damage imparted on the sheared edge during the shearing process [14]. Work hardening strongly influences strain path. There is also the influence of ultimate tensile strength (UTS) to consider. Figure 2.7 illustrates the hole expansion ratio (HER), a standard measure of shear edge formability, for a variety of
steel grades as a function of UTS. As UTS increases, the HER decreases for most steel grades. However, DP steels do not follow the trend as well as the other steels. The lines overlaid on Figure 2.7 indicate that DP steels initially have HER decreasing with increasing UTS, and at approximately 780 MPa the HER increases with increasing UTS. This trend suggests that the microstructure in these steels affects sheared edge formability.

![Figure 2.7 HER versus UTS for a variety of steel grades. Increasing ultimate tensile strength is associated with lower hole expansion ratios. Adapted from experimental work by Sadagopan et al. [15].](image)

2.3.1 Shearing Process

Edge formability of sheet steel is often described as the ability of the steel to be stamped into a specific part without necking or fracture at the sheared edge [16]. Factors such as stress state, edge condition and quality, shearing process, volume fraction of hard phases, amount of carbon in the steel, previous deformation history, and micro-cleanliness can affect the sheared-edge formability limits. Figure 2.8 shows a schematic of the tooling used during the shearing process. During the shearing process, the sheet is held between an upper pad and a lower die through the use of pad pressure. The blade moves downward, shearing the sheet. Both the die and the shear blade have specific radii, with the radius of the blade usually smaller. A smaller radius gives rise to a sharper blade. The shear blade and die are separated by a distance that is
described as the clearance, which is usually quantified as a percentage of the sheet metal thickness.

A shear face is produced on the sheet after it is cut. In addition a zone of deformation is developed directly behind the shear face. This deformation zone is referred to as the shear affected zone (SAZ). Deformation during shearing happens at very high strain rates with a significant increase in local temperature caused by the deformation.

During the shearing process, several different regions are formed on the sheared face. Figure 2.9 provides a visual representation of how the regions are formed during the shearing process [17]. These regions are: the rollover, which is formed as the blade makes initial contact with the sheet and it plastically deforms; the burnish, which is formed when the punch blade penetrates the steel in a vertical manner; and the fracture region, which is formed when a crack initiates and propagates through the remaining thickness of the sheet. Often, there is also a burr, which is a protrusion of metal from the edge. Figure 2.10 gives a visual representation of these four regions [18]. The size of these regions depends on the sheet material and the shearing process itself. The clearance during shearing is often the primary parameter varied to make changes in the size of these regions. The material near the edge undergoes significant strain hardening, and voids are often present in the SAZ for multiphase steels [19]. Macroscopically it is observed that cracks often initiate in the fracture region close to the burr.
Microscopically, AHSS normally have a microstructure with various amounts of hard and soft phases. During deformation, the hard phase will often deform elastically initially while the soft phase deforms plastically. The hard phase can deform plastically too, but strain partitioning often occurs to the softer phase. Cracks grow along the interface between the soft and hard phases. Crack growth increases with increasing surface area of the interface between soft and hard constituent.

![Figure 2.9](image)

Figure 2.9 Schematic illustrating steps in the shearing process. a) the rollover phase of shearing, b) an expanded view of rollover with corresponding flow lines, c) the burnishing phase of shearing, d) expanded view of burnishing with corresponding flow lines e) the fracture phase of shearing, and f) an expanded view of the initiation of fracture and location of the burr [17].
2.3.2 Influence of the Shear Affected Zone

The influence of the resulting SAZ on sheared-edge formability can be attributed to three primary factors: 1) surface roughness and imperfections, 2) plastic deformation and work hardening, and 3) initiation and growth of microvoids or other damage due to plastic deformation during the shearing process. The hardened material in the SAZ facilitates premature nucleation of microvoids due to severe pre-straining. Voids nucleated during punching are expected to remain small because voids do not grow as fast under shear loading but rapidly grow when subjected to the tensile loading of the hole expansion test [11].

A study performed by Pathak et al. [11] showed that the size of the SAZ depends on the sheet thickness and steel grade. Figure 2.11 shows the micro hardness plotted against distance from the hole edge, measured in the fracture region, for two different steels: a DP780 with a 1.5 mm thickness and a DP600 with a 1.8 mm thickness. As the distance from the hole edge increases, the amount of work-hardening decreases and eventually the unaffected base material hardness is reached [11]. For the DP780, the base material hardness is reached at an approximate
distance of 0.5 mm from the hole edge. For the DP600, the base material hardness is reached at an approximate distance of 0.9 mm from the hole edge. The two DP steels have different depths of work hardened zones. In this case, the volume fractions of ferrite and martensite are different between the two conditions. The DP600 contains approximately 84.5% ferrite with the balance consisting of martensite and bainite. The DP780 contained 63% ferrite with the balance consisting of martensite and bainite.

Figure 2.11 Average micro hardness profiles for DP600 and DP780, showing the depth of the work hardening effects from the SAZ. Adapted from experimental work performed by N. Pathak et al. [11].

Since there is evidence that a severely work hardened SAZ limits sheared-edge formability, hole edge condition becomes critical. Pathak et al. [11] also reported that the effect of different hole edge conditions can result in different values of HER. Figure 2.12 shows the various hole edge conditions and their corresponding HER for a DP600 steel. The punched and then reamed holes have a significantly higher HER as compared to punched holes, because the SAZ was essentially removed after punching and therefore there is little initial damage around the hole edge. The drilled holes are subjected to some surface damage and work hardening at the hole edge, which results in a lower HER than the reamed holes. Figure 2.12 also shows that removal of the roughened sheared edge by polishing of the shear face only marginally improves HER. It had been thought that removal of the rough edge would make a significant improvement
in HER, but this study proves that the only way to significantly improve HER is to remove the SAZ.

Figure 2.12 Average HER for a DP600 steel with various hole edge conditions. Adapted from experimental work performed by N. Pathak et al. [11].

Figure 2.12 also shows a slightly higher HER for the burr-down position. This positioning means that the burr was in contact with the conical punch, and there is a burr-removal/flattening operation. There is a slight compression in the burr when it comes in contact with the punch, which leads to mitigation of void nucleation and growth. Also, the top surface of the hole expands freely in uniaxial tension during the upward movement of the conical punch. There is a strain gradient through the material thickness, with the top surface in greater tension than the bottom surface [11]. In the burr-up position, the formation of surface cracks and damage within the fracture zone of the hole is promoted in the region of greater tension, which results in a lower HER.

Pathak et al. [11] showed that the best edge formability was achieved with a reamed hole edge or an edge without the SAZ. To determine the fundamental cause for the decrease in HER values in punched steels, Butcher et al. [12] performed a normalizing heat treatment on a boron steel with various hole edge conditions to remove the work hardened effects of the SAZ while leaving the voids nucleated in the SAZ and the surface roughness intact. Figure 2.13 shows the effects of the heat treatments on equivalent strains at failure obtained from hole expansion tests. By removing the work hardening effects through a normalizing heat treatment, the formability of
the material with a sheared edge was restored to that of the reamed edge condition, even with the roughened surface and the nucleated voids still present on and near the edge.

It must be noted that Butcher et al. [12] used a boron steel for this study. Additional heat treatments would change the microstructure and properties. The martensite condition was achieved by soaking at 925°C for five minutes then quenching in water. The bainite condition was achieved by soaking at 925°C for five minutes then cooling to room temperature in still air. The normalized bainite was achieved by a second re-austenization step at 925°C for five minutes followed by cooling in still air. Normalizing sheared edges of automotive parts and components may not be practical, but it is interesting to note the drastic improvement of sheared-edge formability with such a heat treatment.

Figure 2.13  Equivalent strains at failure obtained from hole expansion tests for boron steels. Plot from the experimental work by Butcher et al. [12].

The large amount of deformation that is induced in the SAZ during shearing means that the metal in this zone has been work hardened extensively and is harder than the base sheet metal. The higher strength in the SAZ affects the strain path during the initial stage of stretching a sheared edge. The strain path during a flat punch hole expansion test is represented by a major strain (true circumferential strain) and a minor strain (true radial strain) [17]. Figure 2.14 shows the variation in strain path for a sheared edge with the SAZ and a sheared edge with the SAZ.
removed, as well as a line of experimental data from thickness measurements [20]. The initial strain path for the SAZ deviates from the uniaxial tensile path, since the material experiences a small amount of biaxial stretching prior to moving along the tensile strain path due to the presence of a highly work-hardened area.

Various factors can influence the value of HER determined during a hole expansion test. Removing the SAZ results in higher stretchability (i.e. higher values of HER), but would increase costs significantly if used in production. The difference in the HER value between a machined hole, where the SAZ has been removed, and a sheared hole is dependent on work-hardening [21]. The presence of a highly work-hardened zone limits local formability. Varying the sheared hole diameter in a hole expansion test usually results in no discernible differences in HER performance for high strength steels.

![Circumferential strain versus radial strain plot](image)

**Figure 2.14** Circumferential strain versus radial strain plot showing the experimental strain path compared to the strain path from the simulation with and without the shear affected zone using the equivalent strain measurement methods [20].

### 2.3.3 Hole Expansion of Various AHSS

Multiple micro-constituents within the same microstructure result in mechanical properties, including HER, that are dependent upon the volume fraction, morphology, and
dispersion of such micro-constituents. Karelova et al. reported that variations in HER between DP and complex phase (CP) steels of the same strength level can be explained by differences in their microstructure [22]. Individual constituent properties within a microstructure affect local formability. Hasegawa et al. showed that multiphase microstructures have a lower HER than single phase grades with similar strength levels and attributed the results to hardness differences between constituent phases [4]. Sugimoto et al. showed similar results where lower HER values for a multiphase microstructure were obtained as compared to a single phase material at the same strength level [23]. Taylor et al. also showed that HER decreased with increasing martensite hardness and martensite/ferrite hardness ratio for DP steels, attributing it to greater strain partitioning to the ferrite during plastic deformation and resulting in interface incompatibility, leading to decohesion [24].

Multiphase microstructures exhibit lower HER values than single phase grades that have similar strength levels, which are assumed to be related to hardness differences between the phases. For DP steels, a large amount of martensitic phase surrounding a ferritic phase can take significant amounts of plastic deformation, reducing localization in the ferritic phase and therefore delaying the material’s failure [25]. Plastic strain distribution within the microstructure directly affects the fracture behavior during hole expansion [26]. It has been observed in some high strength low alloy (HSLA) steels that small bands of coarse ferrite grains along with reduced bainite content have a negative effect on HER values. The presence of martensite and microstructural bands are also detrimental to HER values in such steels [16]. The exact relationship between HER value and microstructure can be rather complex.

2.3.4 Correlation of HER and Tensile Properties

There is interest in correlating tensile properties to both microstructure and HER, because it would be useful to predict HER performance from tensile tests, which are relatively easy to perform. The YS/UTS ratio has proven to be one of the more accurate correlations to HER. Figure 2.15 shows the relationship between YS/UTS ratio and HER. Jin et al. showed for various grades of quenched and partitioned (Q&P), DP, and TRIP steels that as the YS/UTS ratio increased, the HER increased as well [27]. In their study, Q&P and DP steels of the same strength level had significant differences in HER. The DP steels provided better HER results at
comparable YS/UTS ratios. The study is an example of how both the tensile properties and microstructural characteristics play a major role on the HER.

![Graph showing relationship between YS/UTS ratio and HER](image)

**Figure 2.15** Relationship between YS/UTS ratio and HER. Plot adapted from the experimental work by Xinyan et al. [27].

It is assumed that the strain path of stretching of a sheared edge is essentially equivalent to a tensile test. However, failure during hole expansion is different from a tensile failure since there is often no necking prior to failure (localized deformation). Hole expansion has been studied and related to the tensile strength, total elongation, transverse elongation, normal anisotropy, and strain hardening exponent. Some of these properties have a better correlation with HER than others. The hole expansion ratio increases almost linearly with increasing transverse total elongation. Hole expansion is higher in materials that have a higher total elongation, post-uniform elongation, and a higher normal anisotropy. The plastic strain ratio ($r$) is a measure of a material’s resistance to thinning and is defined as the ratio of the width strain to thickness strain during the stretching of the sheet in the length direction. The normal anisotropy ($r_m$) is an averaged value of the plastic strain ratio for different testing directions. The expression for normal anisotropy is

\[
\text{Normal Anisotropy} \ (r_m) = \frac{r_0 + 2r_{45} + r_{90}}{4} \quad (2.2)
\]
where 0, 45, and 90 indicate angular degrees of the tensile axis from the rolling direction. As the value of \( r_m \) increases, there should be more deformation and extension in the tensile direction before failure occurs [28].

Ultimate tensile strength (UTS) has one of the major correlations to the HER values of AHSS. Figure 2.7 shows that as the UTS for various sheet steels increases, the HER value decreases. It should be noted that HER decreases almost linearly with increasing tensile strength up until about 590 MPa. Steels with tensile strengths over 780 MPa have a relatively constant HER value. This constant value may be due to the hard phase reaching a certain volume fraction, which may limit the amount of deformation in the softer phase and therefore resulting in a non-changing HER as a function of strength level [29].

Post-uniform elongation can be correlated to the HER value of sheet steels including AHSS of various strength levels. Figure 2.16 shows the correlation. It is clear that there is almost a linear relationship between the two parameters. Increasing post-uniform elongation results in increasing HER. Figure 2.17 shows the effects of plastic anisotropy on the HER values of AHSS of various strength levels. It can be observed that hole expansion ratio increases with increasing normal anisotropy. In both figures the correlation is not as strong for DP steels individually, or with some single classes of steels.

Variations in sheared edge stretching limits exist among steel grades, and these variations are not fully explained by tensile properties. Factors such as chemical composition, processing routes, phases present, volume fractions of such phases, distribution of phases and their properties could be the reason behind these variations. HER has been correlated with yield strength, ultimate tensile strength, total elongation, post uniform elongation, and normal anisotropy with a range of success.
Figure 2.16  Correlation between post-uniform elongation and hole expansion ratio. Adapted from experimental work by S. Sadagopan et al. [15] and re-plotted by S.K. Paul [30].

Figure 2.17  Correlation between plastic anisotropy and hole expansion ratio. Adapted from experimental work by S. Sadagopan et al. [15] and re-plotted by S.K. Paul [30].
2.4 Angular Stretch Bend

The angular stretch bend (ASB) test provides insight on deformation of sheet steels that are subjected to a combination of stretching and bending simultaneously. The ASB test simulates, in a realistic manner, some of the effects of forming operations of sheet metal automotive parts and components. This test is considered to be close to static, or quasi-static, when the punch displacement speed is fairly low. Height at failure is often used as the comparative measurement instead of load. This is due to the failure location varying between the punch and sidewall. In most cases, failure location transitions from the sidewall to the punch radius as the punch radius decreases. Figure 2.18 shows the difference between a punch radius failure and a sidewall failure [15].

Figure 2.18 Typical observed failure locations for the smallest and largest R/t ratio on angular stretch bend test samples [15].

Figure 2.19 shows a plot of height at failure as a function of critical R/t ratio [15]. It can be observed that as the R/t ratio increases up to a certain limit the transition of failure location from punch to sidewall occurs. However, Sadagopan et al. showed in this study that this is true for most steels except DP steels. It was considered that the effect of bending severity on the ability to form a part might be minimal beyond a threshold R/t ratio. DP steels seem to have no transition of failure location.

For significantly high friction conditions at the interface between the punch and the sheet, there is no net displacement of the sheet over the punch. Lubrication conditions can be varied to reduce and/or control the friction during the test. There is a near plane-strain condition at the
sheet-die interface [31]. In order to achieve a deeper understanding of all the aspects of deformation during industrial practices, there needs to be a combination of results from different mechanical tests. The angular stretch bend test provides useful information that can complement tensile testing and HER testing.

![Figure 2.19](image-url)  
**Figure 2.19** Height of failure as a function of R/t ratio. Most grades have a critical R/t ratio which causes transition from punch radius failure to sidewall failure [15]. *(Color Image see PDF Copy).*
CHAPTER 3: EXPERIMENTAL PROCEDURE

This chapter presents the experimental materials and methods performed to evaluate the mechanical and microstructural properties to be correlated. The materials used in this study, their microstructural and sheared edge characterization, tensile testing, hole expansion testing, and angular stretch bend testing are presented in this chapter. The primary focus of the current study was to analyze the effect of microstructure on sheared edge formability and correlate mechanical properties to hole expansion ratios.

3.1 Material Selection

In order to focus on the microstructural effects on sheared edge formability, steels within a strength range of 1000 ± 100 MPa were selected for this study. Four commercially produced dual phase steels with varying chemical compositions, designated A-D, a DP1180 grade, as well as one commercially produced trip-aided bainitic ferrite steel (TBF980) were obtained for the current study. One press hardened steel, designated 22MnB5, was also obtained. All of these steels were cold-rolled prior to annealing and are uncoated with the exception of TBF980, which has an electro-galvanized coating. Table 3.1 lists the corresponding thickness and nominal compositions for each steel. The DP steels contain micro-alloying additions of Nb and V for microstructural refinement.

All material was received as sectioned panels from the coil. The material was labeled and divided into sample sets for each type of test for each steel upon machining. For each sample set, five samples were allotted for evaluation of tensile properties, fifteen for hole expansion ratio evaluation, ten for angular stretch bend evaluation, and five for initial characterization of microstructure and sheared edges. Remaining samples were kept as backup for additional testing, if required.
Table 3.2 – Chemical Compositions of Experimental Steels (wt. pct.)

<table>
<thead>
<tr>
<th>Steel</th>
<th>Thickness (mm)</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP980 A</td>
<td>1.45</td>
<td>0.162</td>
<td>2.04</td>
<td>1.46</td>
</tr>
<tr>
<td>DP980 B</td>
<td>1.00</td>
<td>0.116</td>
<td>1.77</td>
<td>0.51</td>
</tr>
<tr>
<td>DP980 C</td>
<td>1.45</td>
<td>0.101</td>
<td>2.07</td>
<td>1.50</td>
</tr>
<tr>
<td>DP980 D</td>
<td>1.40</td>
<td>0.140</td>
<td>1.86</td>
<td>0.50</td>
</tr>
<tr>
<td>TBF980 (EG)</td>
<td>1.00</td>
<td>0.155</td>
<td>2.15</td>
<td>1.41</td>
</tr>
<tr>
<td>DP1180</td>
<td>1.10</td>
<td>0.167</td>
<td>2.13</td>
<td>1.34</td>
</tr>
<tr>
<td>22MnB5</td>
<td>1.10</td>
<td>0.257</td>
<td>1.24</td>
<td>0.21</td>
</tr>
</tbody>
</table>

3.1.1 Tempering Study for 22MnB5

The 22MnB5 steel was heat treated to obtain a tensile strength within the desired range. This heat treatment consisted of a full austenitization step (880 °C for 30 s) in a salt pot followed by a water quench, then a tempering step (475 °C for 1 h) and a water quench. This tempering condition was chosen after a tempering study was performed in order to achieve the desired strength range of 1000 ± 100 MPa.

The tempering study consisted of 25.4 x 25.4 mm (1x1 in) coupons going through the same full austenitization and quenching step then tempered at different temperatures for one hour. Initial hardness measurements were taken using the Rockwell C scale, which were converted to tensile strength predictions.

3.2 Material Characterization

Microstructures of the steels were evaluated initially in both a quantitative and qualitative manner. Light optical microscopy (LOM), scanning electron microscopy (SEM), and x-ray diffraction (XRD) were the methods used to image and quantify these microstructures. Volume fraction analysis was performed to quantify the micro-constituents present.

3.2.1 Volume Fraction Analysis

Characterization of the steels required an evaluation of the volume fraction of the micro-constituents present in the corresponding microstructure. LOM micrographs were used for
manual point counting according to the method of the ASTM E562-11 standard [32]. Samples for volume fraction analysis were prepared using standard metallographic procedures following ASTM E3 [33]. All steels were etched with 2% nital for approximately 10 seconds. A circle grid was overlaid on multiple micrographs where the phase at grid intersections was recorded and a fraction was calculated from these point intersections. Five fields were analyzed for each steel, with a total point count of 2500 per steel. The average carbon content of martensite was calculated for DP steels using a basic rule of mixtures formula shown below:

\[ C_{\alpha} = \frac{C_{\text{Bulk}}}{MVF} \]  

(3.1)

Where \( C_{\alpha} \) is the carbon content (wt. pct.) of martensite, \( C_{\text{Bulk}} \) is the carbon content (wt. pct.) of each steel, and with the assumption that there is no carbon in the ferrite and all of the carbon is in the martensite.

3.2.2 X-ray Diffraction

XRD was conducted on samples of the TBF steel to determine the retained austenite volume fraction. Samples were prepared by grinding and polishing down to a 6 µm surface finish. In order to remove the deformed surface layer, the samples were etched with a solution consisting of 50 parts water, 50 parts hydrogen peroxide, and 1 part hydrofluoric acid for approximately 10 minutes. XRD was conducted using a Cr-\( \alpha \) source, with a scan range between 50° and 150° 2θ with a scan speed of 1.833°/min. Retained austenite measurements were made from XRD patterns using a Rietveld refinement, which involves refining a theoretical line until it matches the measured pattern. This analysis was performed using the general structure analysis system (GSAS) program developed by Los Alamos National Laboratory [34, 35]. Figure 3.2 shows a sample of the XRD pattern and refinement.
Figure 3.1 Sample XRD pattern for the TBF steel. The calculated refinement is overlaid on the experimental pattern. The difference between calculated and experimental results is shown below. *(Color Image see PDF Copy)*.

The calculated line in Figure 3.1 shows the calculations by the program based on the Rietveld refinement approach, which is then used to determine the volume fraction of retained austenite in the specimen. The difference line shows how close the calculated pattern and the experimental pattern match to each other.

### 3.2.3 Quantitative Stereology

Characterization of the steels required quantification of martensite size and dispersion. In order to achieve this characterization, different types of measurements were taken using LOM micrographs. Several assumptions were made for this quantitative metallography. Using LOM micrographs limits resolution of martensite boundaries, and it was assumed that all martensite packets in contact with each other were a single martensite colony or particle for measurements. Another assumption made was that these colonies were of a spherical shape. The measurements include mean free distance between martensite colonies, which is determined using the following equation defined by Fullman [36]:

\[
\lambda = \frac{1-V_f}{N_L} \text{mm}
\]  

\[ (3.2) \]
where $\lambda$ is the mean free distance between martensite colonies, $V_f$ is the volume fraction of martensite, and $N_L$ is the number of interceptions per unit length of test lines with martensite colonies. Another measurement was contiguity, which is defined as the fraction of the total interface area of martensite that is shared by particles (or in this case colonies) of martensite. The contiguity of martensite was determined using the following equation defined by Gurland [37]:

$$
C_{mm} = \frac{4(P_L)_{mm}}{4(P_L)_{mm} + 2(P_L)_{mf}}
$$

(3.3)

where $C_{mm}$ is the contiguity of martensite, and $(P_L)_{mm}$ and $(P_L)_{mf}$ are the number of intersections per unit length of test line with martensite-martensite and martensite-ferrite interfaces.

Figure 3.2a shows a representative micrograph with a circle grid of known dimensions where counts were made for quantitative stereology analysis. Martensite colony size was determined using the following equation defined by Fullman [36]:

$$
r = \frac{3V_f}{4N_L}_{mm}
$$

(3.4)

where $r$ is the martensite colony radius, $V_f$ is the volume fraction of martensite, and $N_L$ is the number of interceptions per unit length of test lines with martensite colonies. Figure 3.2b shows a representative micrograph with martensite colonies that are circled. A martensite colony was defined as such when multiple packets were in contact with each other that had large interface contact and were of a spherical shape. Some colonies consisted of smaller packets with large boundary contact between them and with some ferrite present as shown with the arrows in Figure 3.2b. Differentiation between colonies that were close or slightly in contact was defined by obvious discontinuity between them. Another measurement was number of martensite colonies per unit area, which consisted of a manual count of such colonies in a specified area of several LOM micrographs.
3.3 Sheared Edge Characterization

The face of the sheared edge was characterized, along with the shear affected zone (SAZ), in order to evaluate the effects of microstructure on sheared edge formability. The distinct features of the regions on the sheared face were characterized via fractography. Figure 3.3 shows an example of a characteristic shear face with the distinct regions easily discernible for analysis. The size of the two dimensional projection of the roll-over, burnish, fracture, and burr regions were measured using ImageJ, an image analysis software. Measurements were made in the cross section of the sheared edge for each individual region, with a total of ten individual measurements per steel. An average and standard deviation was calculated from these measurements.
3.3.1 Micro Hardness

Vickers micro hardness was performed on the sheared edge of various specimens to quantify the depth of the SAZ for each grade. The micro hardness tests were performed following the ASTM E384 standard [38]. Samples containing a cross-section view of the sheared edge were ground and diamond polished to 1 µm. Three micro hardness profiles were performed for each specimen from the fracture region of the shear face into the base material. Figure 3.4 shows a schematic of the location of these micro hardness profiles. The depth of the SAZ was calculated as the average depth between the deepest point where the hardness is the base material hardness and the next deepest point where work hardening is observed. From this average, an uncertainty was calculated as well as a standard deviation. The data are presented as a micro hardness profile with respect to location on the sheared edge. Figure 3.5 shows a sample micro hardness profile from DP600 steel.
3.4  **Tensile Testing**

Tensile tests were performed for all steels following the ASTM E8 standard [39]. The tests were performed on a 89 kN (20,000 lb) screw-driven load frame at a crosshead displacement rate of 2.54 mm/min (0.1 in/min) with the gauge section displacement monitored.
with an extensometer. The tests were performed with the tensile axis parallel to the rolling direction. The properties of yield strength (YS), UTS, uniform elongation, total elongation, as well as strain-hardening exponent and strength coefficient for the Holloman power law hardening equation were obtained from each test and averaged, with the standard deviation used as uncertainty. The Holloman power law equation is the following:

\[ \sigma = K \epsilon^n \]  

(3.1)

where \( \sigma \) is the flow stress, \( K \) is the strength coefficient, \( \epsilon \) is the plastic strain, and \( n \) is the strain-hardening exponent.

### 3.5 Hole Expansion Testing

Hole expansion tests were performed on the steels listed in Table 3.1. Sample blanks for hole expansion were machined with 101.6 x 101.6 mm (4x4 in) dimensions. These blanks were then punched at the center. Blank punching was performed on an Interlaken formability press at a constant displacement rate of 25.4 mm/s (1 in/s). The hole expansion blanks had a 10 mm hole punched with a clearance within a 12.5-15 % range. The punch tooling consisted of a 10 mm punch mounted to the top of the frame and a coordinating \( \gamma \) die plate. Figure 3.6 shows the punch tooling set-up. The die plate was interchangeable in order to provide the clearance within the specified range. Clearance is expressed as a percentage and is a function of the punch diameter, die diameter, and sheet thickness according to the following equation:

\[ \text{Clearance (\%)} = \left( \frac{D_{\text{Die}} - D_{\text{Punch}}}{2t} \right) \times 100 \]  

(3.2)

where \( D_{\text{Die}} \) is the die diameter, \( D_{\text{Punch}} \) is the punch diameter, and \( t \) is the thickness of the steel sheet. Punched holes were visually inspected in order to assess quality of the sheared edge and maintain uniform conditions by preventing punching with damaged tooling that could negatively affect hole expansion performance.
The hole expansion tests were conducted on an Interlaken formability press. All samples were tested using a conical punch (top angle of 60°) in the burr up position, relative to the punch coming up from the bottom. A constant hold-down force of 445 kN (100 kips) was used on the perimeter of the sample to prevent draw-in. A constant punch displacement rate of 50.8 mm/min (2 in/min) was used for all tests. Visual aid was provided by a monitor screen connected to a digital camera directed on top of the expanding hole. The hole in each sample was measured using a caliper in three different directions (0°, 45°, 90°) after expansion to the point of a through thickness fracture so that the HER of each steel could be calculated. Ten samples were tested for each condition.

3.5.1 Hole Expansion Crack Analysis

An area surrounding the propagating crack was sectioned from the tested hole expansion samples. The samples were then prepared using standard metallographic procedures following ASTM E3 [33]. Micrographs were acquired using a FEI Quanta 600I environmental scanning electron microscopy (E-SEM) at 20kV accelerating voltage, 10 mm working distance, and a spot size of 4.0. This was performed in order to analyze at higher magnifications how the cracks were propagating through the microstructure. Cracks were analyzed for all DP steels.
3.6 Angular Stretch Bend Testing

Angular stretch bend (ASB) tests were performed on the steels listed in Table 3.1. Sample dimensions were 180 x 25 mm. These tests were performed on a 100 kip servo-hydraulic frame with a punch displacement rate of 5.08 mm/s (0.2 in/s). Figure 3.7 shows the angular stretch bend test set-up. Samples were held down by a clamp down force of 151 kN (34,000 lbs) and locking the sample in place with drawbeads. The tests were performed under high lubricant conditions with the purpose of reducing friction as much as possible. The lubricant conditions consisted of a PTFE (Teflon) film coated with LPS2 lubricant that was placed between the steel specimen and the tooling (i.e. the punch nose). The tests were performed for three different punch radii of 1.0 mm, 2.5 mm, and 5.0 mm. These radii were chosen under the principle that fracture should occur for the sheet on the punch nose when there is a sharp radius. Height at failure and reduction in area measurements near the fracture were evaluated and compared against the punch radius to material thickness ratio (R/t). Three samples were tested per radii per steel. The results from the three tests were averaged, with the standard deviation used as uncertainty.

Figure 3.7 Angular stretch bend set-up. (Color Image see PDF Copy).
CHAPTER 4: RESULTS

The initial goal of this project was to evaluate the effect of sheared edge formability based on microstructure and tensile properties. A set of different microstructures that provide a strength level of 1000 ± 100 MPa were selected. In this chapter, the results obtained following the experimental methods described in the previous chapter are presented. Some of the results will focus on the DP steels in the current study in order to understand the role ferrite-martensite microstructures better. The DP980 steels have similar martensite volume fractions, which allow the study of the effect of morphology and dispersion of martensite on the performance. Appendix A contains results from a preliminary study performed with steels varying in grade, strength, thickness, and microstructural characteristics. The preliminary study was performed to become familiar with the testing methods. Nevertheless, some interesting data, which are captured in Appendix A, were obtained during this preliminary testing.

4.1 Material Selection

4.1.1 Tempering Study Results

Initial tempering heat treatments on the press hardened steel were performed in 50 °C increments. Once those results were analyzed, the temperature range was narrowed between 450 – 500 °C, and additional heat treatments were performed in 10 °C increments at 465 °C, 475 °C, and 485 °C. Figure 4.1 shows the predicted tensile strength, based on hardness measurements, as a function of tempering temperature. The 475 °C heat treatment resulted in a predicted tensile strength of 1038 MPa; therefore, it was decided to use that tempering temperature for the press hardened steels in the study.
Figure 4.1  Predicted tensile strength as a function of tempering temperature used to determine the desired heat treatment for 22MnB5 steel.

4.2 Microstructural Characterization

Microstructural characterization was performed via light optical microscopy (LOM). Figure 4.2 shows the microstructures of the DP and TBF steels. It can be observed in Figures 4.2a – 4.2e that the DP microstructures consist of a ferritic matrix (light colored regions) with martensite islands (darker colored regions) dispersed throughout. Figure 4.3f shows the microstructure of the TBF steel where the microstructure consists mostly of a ferritic/bainitic matrix with some martensite and retained austenite dispersed throughout. Figure 4.3 shows the microstructure of 22MnB5 steel prior to heat treatment (Figure 4.3a), which consists of ferrite-pearlite, and post-heat treatment (Figure 4.3b), which consists of tempered martensite.
Figure 4.2  LOM micrographs of DP980 A (a), DP980 B (b), DP980 C (c), DP980 D (d), DP1180 (e), and TBF980 (f). Etched with 2 pct. nital. (Color Image see PDF Copy).
4.2.1 Volume Fraction Analysis

A volume fraction analysis was performed on the multiphase steels as specified in the Experimental Procedures chapter by manual count following ASTM standard E562-11 [32]. Table 4.1 shows the results of volume fraction analysis for all steels with the exception of 22MnB5 which has a tempered martensite microstructure. The volume fraction of hard phases, which include martensite and bainite is shown, with the remainder being ferrite. The DP980 steels contain a similar volume fraction content of martensite. The TBF980 retained austenite content was determined via x-ray diffraction. This retained austenite volume fraction was calculated at $3 \pm 1$ and is lower than expected based on the strength of the material. Figure 4.4 shows an SEM micrograph for TBF980 where it can be observed that it is a fine microstructure, which complicates discerning the different phases present. Therefore, it is difficult to quantify the volume fractions in this steel.

<table>
<thead>
<tr>
<th>Steel</th>
<th>Martensite</th>
<th>Ferrite</th>
<th>Martensite C-Content</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP980 A</td>
<td>45 ± 1</td>
<td>Bal.</td>
<td>0.320</td>
</tr>
<tr>
<td>DP980 B</td>
<td>39 ± 1</td>
<td>Bal.</td>
<td>0.297</td>
</tr>
<tr>
<td>DP980 C</td>
<td>38 ± 1</td>
<td>Bal.</td>
<td>0.266</td>
</tr>
<tr>
<td>DP980 D</td>
<td>37 ± 1</td>
<td>Bal.</td>
<td>0.378</td>
</tr>
<tr>
<td>DP1180</td>
<td>50 ± 1</td>
<td>Bal.</td>
<td>0.334</td>
</tr>
</tbody>
</table>
4.2.2 X-ray Diffraction

XRD scans were performed on the TBF980 steel to determine retained austenite content in the as-received material and, after being tested for hole expansion, a section from the edge to determine the amount of retained austenite transformed during deformation. A small area on the perimeter of the expanded hole was sectioned off for analysis, which is where most of the deformation occurred. Figure 4.5 shows the patterns for both conditions. It can be observed that in the post-testing pattern there is an absence of most of the retained austenite peaks, which confirms transformation of retained austenite to martensite. The initial retained austenite content was determined as 3 ± 1 wt. pct., but it is not possible to quantify with accuracy the remaining retained austenite upon transformation since the detection accuracy limit is 1 ± 1 wt. pct. in XRD.
4.2.3 Quantitative Stereology

An analysis of microstructural characteristics was performed on the DP steels in order to better understand the effect of microstructure morphology on sheared edge formability. Some of the measurements made were mean free distance (MFD) between martensite colonies, contiguity, size of martensite colonies (radius in µm), and number of colonies per unit area (mm$^2$ in this case). Table 4.2 lists the results of these measurements. Several assumptions were made in order to quantify these results, as described in the Experimental Procedure chapter. These assumptions allow a means to quantify some of the morphological aspects of the microstructures.

Table 4.2 – Summary of Quantitative Stereology Results

<table>
<thead>
<tr>
<th>Steel</th>
<th>MFD between Martensite Colonies (µm)</th>
<th>Contiguity</th>
<th>Martensite Colony Size (µm)</th>
<th># of Martensite Colonies/ mm$^2$</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP980 A</td>
<td>7.03</td>
<td>0.171</td>
<td>4.31</td>
<td>33,500</td>
</tr>
<tr>
<td>DP980 B</td>
<td>6.06</td>
<td>0.240</td>
<td>2.91</td>
<td>39,600</td>
</tr>
<tr>
<td>DP980 C</td>
<td>5.37</td>
<td>0.272</td>
<td>2.47</td>
<td>36,700</td>
</tr>
<tr>
<td>DP980 D</td>
<td>7.52</td>
<td>0.274</td>
<td>3.31</td>
<td>26,800</td>
</tr>
<tr>
<td>DP1180</td>
<td>4.97</td>
<td>0.265</td>
<td>3.73</td>
<td>29,000</td>
</tr>
</tbody>
</table>

Figure 4.5 XRD patterns of the TBF steel. Both as-received and after testing patterns are shown. (*Color Image see PDF Copy*).
4.3 Sheared Edge Characterization

Samples for hole expansion testing were blanked and punched for each steel. The punching clearance was between 12.5-15% of the respective steel thickness. Observations and measurements were made using optical microscopy on the shear face after punching of the hole. These measurements were made to determine the height of each region on the shear face. Table 4.3 summarizes the region sizes on the shear face for all steels as a percentage of the respective thickness, along with an uncertainty. Figure 4.6 compares in a bar chart the data presented in Table 4.3.

Table 4.3 – Summary of Shear Face Region Analysis as Percentage of Thickness

<table>
<thead>
<tr>
<th>Steel</th>
<th>Rollover (%)</th>
<th>Burnish (%)</th>
<th>Fracture + Burr (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP980 A</td>
<td>4 ± 0.2</td>
<td>20 ± 0.6</td>
<td>76 ± 0.5</td>
</tr>
<tr>
<td>DP980 B</td>
<td>4 ± 0.4</td>
<td>29 ± 0.6</td>
<td>67 ± 0.9</td>
</tr>
<tr>
<td>DP980 C</td>
<td>4 ± 0.2</td>
<td>24 ± 0.8</td>
<td>72 ± 0.7</td>
</tr>
<tr>
<td>DP980 D</td>
<td>4 ± 0.4</td>
<td>12 ± 0.2</td>
<td>84 ± 0.5</td>
</tr>
<tr>
<td>TBF980 (EG)</td>
<td>5 ± 0.5</td>
<td>29 ± 0.9</td>
<td>66 ± 1.4</td>
</tr>
<tr>
<td>DP1180</td>
<td>5 ± 0.4</td>
<td>25 ± 0.8</td>
<td>70 ± 0.9</td>
</tr>
<tr>
<td>22MnB5</td>
<td>3 ± 0.1</td>
<td>29 ± 0.3</td>
<td>68 ± 0.2</td>
</tr>
</tbody>
</table>

Figure 4.6 Rollover, burnish, and fracture + burr region sizes as a function of sheet thickness.
4.3.1 Shear Affected Zone (SAZ)

The work-hardening effect of the shearing process can be observed by the difference in micro hardness from the edge to the base material. Figures 4.7a – 4.7g are plots of micro hardness as a function of distance from the shear face for the DP980 A, DP980 B, DP980 C, DP980 D, TBF980, DP1180, and 22MnB5 steels, respectively. It can be observed that the depth of the SAZ is different for all steels. This depth difference is due to effect of work hardening of the different microstructures, which can then be related to sheared edge formability. The 22MnB5 steel exhibits a very small SAZ, where the base material hardness is reached within 200 µm. Even though there was a shear face produced during the shearing process, a significant work-hardening effect is not observed in this steel. Table 4.4 summarizes the average depth of the SAZ, magnitude of SAZ (ΔHV), and relative depth of the SAZ, which is the ratio of the depth of the SAZ to the thickness of the sheet.

It can be observed that DP980 A, DP980 D, and DP1180 have a very similar relative depth of SAZ but their corresponding hardness magnitudes are different. The same is observed with DP980 B and TBF980. The 22MnB5 and DP980 C steels show the smallest relative depth of SAZ.

![Figure 4.7](image_url) Micro hardness profiles for (a) DP980 A, (b) DP980 B, (c) DP980 C, (d) DP980 D, (e) TBF980, (f) DP1180, and (g) 22MnB5 measured in the fracture and burr region of the sheared edge.
Figure 4.7 Continued.
Table 4.4 – Depth, Magnitude, and Relative Depth of the Shear Affected Zone

<table>
<thead>
<tr>
<th>Steel</th>
<th>Depth of SAZ (µm)</th>
<th>Magnitude of SAZ (ΔHV)</th>
<th>Relative Depth of SAZ</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP980 A</td>
<td>377 ± 61</td>
<td>94 ± 19</td>
<td>0.26</td>
</tr>
<tr>
<td>DP980 B</td>
<td>379 ± 61</td>
<td>71 ± 18</td>
<td>0.38</td>
</tr>
<tr>
<td>DP980 C</td>
<td>296 ± 77</td>
<td>46 ± 18</td>
<td>0.20</td>
</tr>
<tr>
<td>DP980 D</td>
<td>378 ± 61</td>
<td>75 ± 11</td>
<td>0.27</td>
</tr>
<tr>
<td>TBF980</td>
<td>349 ± 51</td>
<td>90 ± 13</td>
<td>0.35</td>
</tr>
<tr>
<td>DP1180</td>
<td>297 ± 76</td>
<td>54 ± 14</td>
<td>0.27</td>
</tr>
<tr>
<td>22MnB5</td>
<td>128 ± 51</td>
<td>21 ± 7</td>
<td>0.12</td>
</tr>
</tbody>
</table>

4.4  Tensile Properties

Tensile tests were performed on all steels in the longitudinal orientation. Table 4.5 summarizes the tensile properties for all steels. Figure 4.8 shows the representative engineering stress versus strain curves of a single test specimen for all steels. The 22MnB5 steel showed an interesting tensile behavior where the yield strength (YS) is also the ultimate tensile strength (UTS). There is a possible yield point elongation where due to absence of work-hardening instability occurs. It also exhibited the least amount of elongation. For all DP steels, a continuous yielding behavior is observed with high initial work-hardening rates. The TBF steel exhibits a more defined yield point, with a lower initial work-hardening rate. The TBF steel exhibits the greatest amount of elongation.

Table 4.5 – Summary of Tensile Properties for Experimental Steels

<table>
<thead>
<tr>
<th>Steel</th>
<th>Total Elongation (pct)</th>
<th>Uniform Elongation (pct)</th>
<th>UTS (MPa)</th>
<th>YS (MPa)</th>
<th>n-value 4-6</th>
<th>K (MPa)</th>
<th>Reduction in Area (pct)</th>
<th>True Fracture Strain (εf)</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP980 A</td>
<td>16.8 ± 0.2</td>
<td>10.2 ± 0.1</td>
<td>1012 ± 3</td>
<td>691 ± 2</td>
<td>0.115 ± 0.001</td>
<td>1466 ± 3</td>
<td>45 ± 0.5</td>
<td>0.60</td>
</tr>
<tr>
<td>DP980 B</td>
<td>13.7 ± 0.5</td>
<td>8.6 ± 0.0</td>
<td>987 ± 6</td>
<td>703 ± 12</td>
<td>0.095 ± 0.001</td>
<td>1363 ± 10</td>
<td>57 ± 0.3</td>
<td>0.85</td>
</tr>
<tr>
<td>DP980 C</td>
<td>15.5 ± 0.8</td>
<td>9.8 ± 0.4</td>
<td>965 ± 6</td>
<td>655 ± 4</td>
<td>0.110 ± 0</td>
<td>1383 ± 8</td>
<td>53 ± 0.8</td>
<td>0.76</td>
</tr>
<tr>
<td>DP980 D</td>
<td>15.4 ± 0.7</td>
<td>10.0 ± 0.1</td>
<td>1005 ± 8</td>
<td>633 ± 3</td>
<td>0.111 ± 0</td>
<td>1444 ± 13</td>
<td>30 ± 0.2</td>
<td>0.36</td>
</tr>
<tr>
<td>TBF980</td>
<td>17.6 ± 0.0</td>
<td>11.9 ± 0.1</td>
<td>987 ± 3</td>
<td>793 ± 8</td>
<td>0.118 ± 0.002</td>
<td>1424 ± 3</td>
<td>57 ± 0.8</td>
<td>0.84</td>
</tr>
<tr>
<td>DP1180</td>
<td>11.0 ± 0.4</td>
<td>6.9 ± 0.1</td>
<td>1175 ± 7</td>
<td>846 ± 10</td>
<td>0.077 ± 0.001</td>
<td>1548 ± 14</td>
<td>46 ± 0.6</td>
<td>0.61</td>
</tr>
<tr>
<td>22MnB5*</td>
<td>6.6 ± 1.5</td>
<td>0.6 ± 0.2</td>
<td>976 ± 25</td>
<td>976 ± 25</td>
<td>-</td>
<td>-</td>
<td>38 ± 0.5</td>
<td>0.48</td>
</tr>
</tbody>
</table>

*22MnB5 Tensile properties after heat treatment.
In addition to 0.2% offset yield strength, ultimate tensile strength, total and uniform elongations, other tensile properties were determined as well. The strain hardening exponent, $n$, was determined from true stress-true strain curves in the true strain range of 0.04-0.06 true strain. The strength coefficient, $K$, was determined from the logarithmic true stress-logarithmic true strain from data between YS and UTS. Reduction in area measurements were made on fractured tensile specimens with a point micrometer. The true fracture strain was calculated from these reduction in area measurements.

It can be observed from Table 4.5 that total elongations for the DP steels vary from 11.0 pct to 16.8 pct, and the TBF and 22MnB5 are at 17.6 pct and 6.6 pct, respectively. The TBF has a higher total elongation, and the 22MnB5 has a lower total elongation than the range for the DP steels. The same is true for uniform elongation. The uniform elongations for the DP steels vary from 6.9 pct to 10.2 pct and the TBF and 22MnB5 are at 11.9 pct and 0.6 pct, respectively. All steels meet the expected UTS with the exception of DP980 C, which is slightly under at 965 ± 6 MPa. The true fracture strain measurements vary from 0.36 to 0.85, which is a large range. DP980 B, DP980 C, and TBF980 are at the top of this range, with DP980 A and DP1180 in the middle, and DP980 D and 22MnB5 at the bottom.

![Figure 4.8](Color Image see PDF Copy)
4.5 Hole Expansion

Table 4.6 summarizes the hole expansion ratio results for all steels. Figure 4.9 compares in a bar chart the data presented in Table 4.6. It can be observed that TBF980, DP980 B, and DP980 C steels exhibited the highest HERs, followed closely by 22MnB5 steel. DP980 A is still on the higher end of the range, followed by DP1180. The DP980 D steel exhibited the lowest HER of all steels at 14.5 ± 0.9 %.

Table 4.6 – Summary of HER for Experimental Steels

<table>
<thead>
<tr>
<th>Sample</th>
<th>DP980 A</th>
<th>DP980 B</th>
<th>DP980 C</th>
<th>DP980 D</th>
<th>TBF980</th>
<th>DP1180</th>
<th>22MnB5</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>45.4</td>
<td>51.9</td>
<td>54.3</td>
<td>14.8</td>
<td>44.6</td>
<td>22.6</td>
<td>48.0</td>
</tr>
<tr>
<td>2</td>
<td>29.2</td>
<td>53.0</td>
<td>55.3</td>
<td>15.1</td>
<td>57.2</td>
<td>28.7</td>
<td>45.0</td>
</tr>
<tr>
<td>3</td>
<td>36.9</td>
<td>54.4</td>
<td>46.0</td>
<td>16.1</td>
<td>52.7</td>
<td>19.4</td>
<td>48.0</td>
</tr>
<tr>
<td>4</td>
<td>33.5</td>
<td>48.4</td>
<td>49.1</td>
<td>13.8</td>
<td>57.6</td>
<td>26.9</td>
<td>45.8</td>
</tr>
<tr>
<td>5</td>
<td>46.2</td>
<td>53.5</td>
<td>36.4</td>
<td>14.6</td>
<td>50.5</td>
<td>26.3</td>
<td>40.5</td>
</tr>
<tr>
<td>6</td>
<td>34.9</td>
<td>57.0</td>
<td>50.2</td>
<td>14.1</td>
<td>46.4</td>
<td>28.6</td>
<td>52.7</td>
</tr>
<tr>
<td>7</td>
<td>37.1</td>
<td>48.7</td>
<td>52.6</td>
<td>13.4</td>
<td>57.4</td>
<td>24.0</td>
<td>41.7</td>
</tr>
<tr>
<td>8</td>
<td>42.9</td>
<td>49.8</td>
<td>55.3</td>
<td>15.7</td>
<td>48.8</td>
<td>36.4</td>
<td>45.9</td>
</tr>
<tr>
<td>9</td>
<td>41.7</td>
<td>44.6</td>
<td>41.1</td>
<td>13.2</td>
<td>47.9</td>
<td>20.2</td>
<td>43.4</td>
</tr>
<tr>
<td>10</td>
<td>37.2</td>
<td>47.8</td>
<td>49.3</td>
<td>14.2</td>
<td>49.8</td>
<td>22.8</td>
<td>45.5</td>
</tr>
<tr>
<td>Average</td>
<td>38.5</td>
<td>50.9</td>
<td>49.0</td>
<td>14.5</td>
<td>51.3</td>
<td>25.6</td>
<td>45.7</td>
</tr>
<tr>
<td>Std. Dev.</td>
<td>5.2</td>
<td>3.5</td>
<td>5.9</td>
<td>0.9</td>
<td>4.5</td>
<td>4.7</td>
<td>3.3</td>
</tr>
</tbody>
</table>

Figure 4.9 Average hole expansion ratios and standard deviations for all steels.
Several observations were made during hole expansion testing. Cracks developed and propagated without following a preferred orientation on all steels. All cracks developed on the top surface, which is where the fracture + burr region is located. There were multiple cracks developed through the test prior to one crack, or in some cases several of them, propagating through the entire sheet thickness. Figures 4.10a and 4.10b show representative images of hole expansion samples as viewed from the top with multiple cracks for TBF980 and DP980 A, respectively. Figures 4.11a and 4.11b show representative images of hole expansion samples as viewed from the top with partial cracks for DP980 A and DP980 C, respectively.

Figure 4.10 Representative images of hole expansion samples as viewed from the top with multiple cracks. Images shown are (a) TBF980 and (b) DP980 A.

Figure 4.11 Representative images of hole expansion samples as viewed from the top with partial cracks. Images shown are (a) DP980 A and (b) DP980 C.
SEM micrographs were taken of the cracks propagating through the microstructure of tested hole expansion samples. Figure 4.12a shows a crack propagating through martensite for DP1180. Figure 4.12b shows a crack propagating through the ferrite, and possibly along ferrite-martensite boundaries, for DP980 D. It is interesting to see the crack propagating through the ferrite. Evidence in literature has shown crack propagation through ferrite/martensite boundaries or through martensite, but not primarily through ferrite. DP980 D exhibited the lowest HER, which may indicate that strain partitioning directly into ferrite results in lower sheared edge formability.

![Representative images of cracks propagating through the microstructure of hole expansion samples. Images shown are for (a) DP1180 and (b) DP980 D.](image)

4.6 **Angular Stretch Bend**

The angular stretch bend test provides another formability measurement. The edges of the samples used for ASB were milled, and therefore there is no SAZ present. A separate study was performed in order to investigate the effect of SAZ on ASB. The results of this small comparison study are given in Appendix B. Height at failure and reduction in area are the measurements made and/or determined for this test in order to quantify formability. The reduction in area results were determined by using a point micrometer to take measurements of thickness near the fracture surface. Although it is considered a near plane-strain test, the width at fracture was measured as well in order to provide accurate reduction in area measurements. These
measurements provide insight regarding the amount of plastic deformation the material experienced before fracture with different punch radii.

Figure 4.13a shows the relationship between R/t ratio and height at failure. It can be observed that as R/t ratio increases, the height at failure increases as well. This trend is observed for all experimental steels. Figure 4.13b shows the relationship between R/t ratio and reduction in area. It can be observed that there is no apparent trend or correlation between R/t ratio and reduction in area.

![Figure 4.13](image_url)

Figure 4.13 Relationship between (a) R/t ratio and height at failure, and (b) R/t ratio and reduction in area for all experimental steels.
CHAPTER 5: DISCUSSION

Sheared edge formability has been determined to be influenced by factors such as microstructural characteristics, shearing conditions, and tensile properties. In this chapter, the hole expansion ratio, a measurement of sheared edge formability, will be compared and correlated to measurements of the factors previously discussed. The results presented in the previous chapter will be used in these comparisons and correlations, where the focus will primarily be on DP steels.

5.1 Influence of Microstructure on Sheared-Edge Formability

Sheared edge formability is influenced by the microstructure. Figure 5.1a shows the correlation between HER and martensite volume fraction for DP steels. The HER is independent of martensite volume fraction for these steels. Figure 5.1b shows the correlation between HER and martensite carbon content. There is a correlation where HER decreases with increasing martensite carbon content, which results in higher hardness difference between martensite and ferrite. This is consistent with results shown by Taylor et al. [40].

![Hole expansion ratio as a function of (a) martensite volume fraction and (b) carbon content of martensite.](image)

Figure 5.1 Hole expansion ratio as a function of (a) martensite volume fraction and (b) carbon content of martensite.
Quantitative stereology results provide a more in-depth microstructural analysis. These measurements were compared to HER. Figure 5.2a – 5.2c show HER as a function of martensite contiguity, mean free distance between martensite colonies, and martensite colony size, respectively. There is no definitive correlation between HER and any of these measurements alone. However, each one of these measurements represents a single aspect or characteristic of the overall microstructure, and it may be more appropriate to look at their combined effects.

Figure 5.2 Hole expansion ratio as a function of (a) contiguity of martensite, (b) mean free distance between martensite colonies, and (c) colony size of martensite for DP steels.
Figure 5.3 shows the correlation between HER and the product of contiguity, mean free distance, and colony size. The product of these measurements is a combination of individual microstructural features that provides a combined measurement or parameter that represents martensite dispersion throughout the microstructure. It takes into account the individual martensite colony size, along with the spacing between these colonies. Figure 5.3 shows that there is a slight trend where HER decreases as the product of these measurements increases.

![Figure 5.3](image)

Figure 5.3 Hole expansion ratio as a function of the product of colony size, contiguity, and mean free distance of martensite for DP steels.

Another measurement that provides information regarding dispersion of martensite is the number of colonies per unit area. Figure 5.4 shows the correlation between HER and number of martensite colonies per unit area. There is a strong correlation where HER increases with increasing number of colonies per unit area. In other words, as the martensite colonies are more dispersed throughout the microstructure, the better performance in sheared edge formability. Data from Appendix A are included for DP980 and DP1180 from the preliminary study. The data follow the trend observed for the experimental steels.

While the martensite volume fraction is very similar for the DP980 steels and the DP1180 martensite volume fraction is slightly higher, the number of colonies per unit area represents the coarseness of the dispersion of martensite colonies throughout the microstructure. At similar
martensite volume fractions, with larger colony sizes (i.e. less colonies per area) there will be higher strain incompatibility at the martensite-ferrite interface, and damage that initiates at larger colonies results in larger flaws. While martensite colonies are the stronger phase, they still have some ductility but they also are higher stress concentration areas. If there are any defects inside the martensite colonies, the surrounding high stresses help in creating strain localization at that defect. Smaller martensite colony size more dispersed throughout the microstructure provides more obstacles for the propagating crack. Looking at mean free distance between martensite colonies, the larger the distance between martensite colonies means that there are large blocks of ferrite in between. This means that there is a larger possible distance for dislocation pile-ups that will ultimately give rise to potential stress concentrations and/or micro-voids creating a path of less resistance for crack propagation. Figure 5.5 provides a schematic of dislocation pile-up theory [41]. These concepts apply to shear edge formability, especially when characterized by hole expansion where the failure criteria consists of a through-thickness crack. The crack propagating through the microstructure will be affected by martensite colony size and mean free distance between martensite colonies.

Correlation coefficients for relationships between HER and quantitative stereology measurements are presented in Table 5.1. These correlation coefficients provide the accuracy of linearity and were calculated using Data Analysis in the Microsoft Excel software. They support the trends discussed in the figures. The strongest relationships are highlighted. There is a strong positive linear relationship between HER and number of colonies per unit area with a correlation coefficient of 0.98. HER has negative linear relationships with carbon content in martensite and the product of colony size, contiguity and mean free distance between martensite colonies, with correlation coefficients of -0.81 and -0.89, respectively.
Figure 5.4  Hole expansion ratio as a function of number of martensite colonies per unit area.

Figure 5.5  Schematic of a dislocation pile-up at an obstacle [41].

Table 5.1 – Correlation Coefficients between HER and Quantitative Stereology Measurements

<table>
<thead>
<tr>
<th></th>
<th>HER</th>
<th>Contiguity</th>
<th>MFD</th>
<th>GS</th>
<th>MVF</th>
<th>#/mm²</th>
<th>%C in Martensite</th>
<th>GS<em>MFD</em>Cont</th>
</tr>
</thead>
<tbody>
<tr>
<td>HER</td>
<td>1.00</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Contiguity</td>
<td>-0.28</td>
<td>1.00</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>MFD</td>
<td>-0.39</td>
<td>-0.37</td>
<td>1.00</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>GS</td>
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<td>-0.70</td>
<td>0.35</td>
<td>1.00</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>MVF</td>
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<td>-0.31</td>
<td>-0.34</td>
<td>0.75</td>
<td>1.00</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>#/mm²</td>
<td>0.98</td>
<td>-0.25</td>
<td>-0.30</td>
<td>-0.46</td>
<td>-0.37</td>
<td>1.00</td>
<td></td>
<td></td>
</tr>
<tr>
<td>%C in Martensite</td>
<td>-0.81</td>
<td>-0.31</td>
<td>0.63</td>
<td>0.82</td>
<td>0.45</td>
<td>-0.79</td>
<td>1.00</td>
<td></td>
</tr>
<tr>
<td>GS<em>MFD</em>Cont</td>
<td>-0.89</td>
<td>0.02</td>
<td>0.75</td>
<td>0.47</td>
<td>0.07</td>
<td>-0.82</td>
<td>0.88</td>
<td>1.00</td>
</tr>
</tbody>
</table>
5.1.1 Influence of Microstructure on Tensile Properties

Correlations between the corresponding tensile properties and microstructure were also investigated. It is of interest to see if any similarities arise in tensile behavior and sheared edge formability caused by the microstructure.

Figure 5.6 shows the correlation between UTS and martensite volume fraction. There is a slight trend where UTS increases with increasing martensite volume fraction. This is not surprising since it is consistent with results found in the literature where increasing martensite content results in an increase in flow stress [42]. Figure 5.7a shows the correlation between true fracture strain, \( \varepsilon_f \), and the product of colony size, contiguity, and mean free distance. The trend indicates that as the product of these measurements increases the true fracture strain decreases. Martensite colony size, mean free distance between martensite colonies, and number of colonies per unit area have a significant effect on strain localization in tensile tests just as they have on hole expansion. Figure 5.7b shows the correlation between true fracture strain, \( \varepsilon_f \), and the number of martensite colonies per unit area, where the trend indicates that as the number of colonies per unit area increases the true fracture strain increases. These same trends are observed for HER.

![Figure 5.6 Ultimate tensile strength as a function of martensite volume fraction.](image)
5.2 Influence of Sheared Edge Conditions on Sheared Edge Formability

Upon shearing, a shear face is produced on the edge of the material. This shear face consists of the rollover, burnish, fracture and burr regions as previously discussed. The size of each region was determined. The influence of the size of these regions on sheared edge formability is discussed in this section.

Figure 5.8 shows the correlation between HER and the fraction of the fracture region on the shear face of all steels. The overall trend is that as the fracture region increases, the HER decreases. This trend could be due to the possible increased area for defects and/or microcracks to exist. Upon sheared edge deformation, i.e. hole expansion, these defects and/or microcracks will limit sheared edge formability. Premature cracks will initiate at these locations. A smaller fracture region will have a lower number of defects and/or microcracks present.

Evidence in literature has shown that fracture initiates in the fracture + burr region of the shear face, which is the top surface in the hole expansion test and where there are larger amounts of strain present. The size of the fracture + burr region increases with martensite volume fraction [1], but considering that volume fraction contents are similar for DP980 steels, this might indicate that the size of the regions on the shear face are also affected by the size of martensite colonies and the mean free distance between them. A larger fracture region does not
imply that a crack will initiate right at the end of the fracture region where the burr is located, but rather at a point where a microcrack or defect present within the fracture region will be subjected to higher strains.

No correlations were observed between HER and relative depth of SAZ or magnitude of SAZ micro hardness. Appendix C contains figures showing the independence of HER to these parameters. The effects of martensite characteristics seem to be a bigger factor than the amount of work-hardening taking place at the sheared edge.

![Figure 5.8](image)

**Figure 5.8** Hole expansion ratio as a function of the fraction of the fracture region on the shear face of all steels.

### 5.3 Correlation Between Sheared Edge Formability and Tensile Properties

Understanding the influence of tensile properties on sheared edge formability has been studied extensively in literature. There is interest in predicting sheared edge formability from a simple tensile test but it has only had limited success. For this reason, the role of microstructure needed to be investigated further. However, relationships can be established that provide some insight on sheared edge formability performance based on tensile properties. A correlation between select tensile properties and HER values are presented in this section.
Figure 5.9 shows the correlation between HER and UTS for all steels. The trend observed follows the literature, where overall the HER decreases with increasing UTS. DP980 D is somewhat of an exception and is not within the observed trend. While there is an overall trend, it is not definitive. Other factors likely influence the HER of these steels with similar UTS values.

Figure 5.9 Hole expansion ratio as a function of ultimate tensile strength for all steels.

Figure 5.10 shows the correlation between HER and post-uniform elongation for all steels. It can be observed that the data separates into two sets. One set consists of DP980 D and DP1180, while the other consists of the rest of the steels. Both sets follow the same trend of a decreased HER with increased post-uniform elongation. This trend is consistent with results in the literature [15]. DP980 D and DP1180 are outliers from the trend. Other factors likely influence the HER of these steels with similar UTS values and similar post-uniform elongation values.

Figure 5.11 shows the correlation between HER and the YS/UTS ratio for all steels. The YS/UTS ratio is a rough measure of work hardening. There is a slight trend where the HER increases with increasing YS/UTS ratio. In Appendix A, it can be observed for the preliminary set of steels with varying tensile strengths and microstructures that the correlation is the opposite. In contrast, Jin et al. [27] showed a correlation consistent with the results in Figure 5.11. This correlation seems to be inconsistent, which leads to the conclusion that work hardening behavior
alone is not sufficient to predict sheared edge formability. 22MnB5 steel is an outlier based on its unique tensile behavior where its YS is also the UTS.

![Figure 5.10](image1.png)

**Figure 5.10** Hole expansion ratio as a function of post-uniform elongation for all steels.

![Figure 5.11](image2.png)

**Figure 5.11** Hole expansion ratio as a function of YS/UTS ratio for all steels.

Figure 5.12 shows the correlation between HER and true fracture strain (TFS) for all steels. With the exception of 22MnB5, there is a consistent trend where the HER increases with increasing TFS. This correlation is consistent with the work by Link *et al.* [43], where the same trend is observed. Since the true fracture strain is determined from a direct reduction in area measurement of tensile specimens in the necked region, it is a measurement of local formability,
similar to the hole expansion ratio. It is interesting to see that local formability measurements from different tests correlate so well, at least for the DP steels.

![Figure 5.12](image.png)

Figure 5.12 Hole expansion ratio as a function of true fracture strain for all steels.

Even though the TFS correlates well with HER, the 22MnB5 is an outlier. The 22MnB5 steel consists of a single phase microstructure of tempered martensite. Taylor et al. showed that hardness ratio between phases present in the microstructure affect hole expansion and true fracture strain [40]. Having a single phase present means that there will hardly be any hardness differences, and therefore better hole expansion performance should be expected. Following the same concept from the microstructural aspect, sheared edge formability must be influenced by more than one factor from the tensile properties perspective. Based on this hypothesis, products of different tensile properties were examined and correlated to HER.

Figure 5.13 shows the correlation between HER and the product of TFS and terminal n-value for all steels, with the exception of 22MnB5 which had no strain hardening. Terminal n-value was calculated within the last 0.02 true strain values. The TFS*n parameter was chosen as a representation of the strain hardening taking place up to the onset of local formability, which is similar to what happens during the hole expansion test. The HER increases as TFS*n increases in a similar way to the Figure 5.12.

Figure 5.14 shows the correlation between HER and the product of TFS and post-uniform elongation for all steels. This parameter was chosen as a representation of local formability
leading to fracture, with the aim of taking into consideration the plastic deformation occurring beyond the onset of instability. There is a trend where the HER increases as the TFS*P-U El increases, with the exception of 22MnB5 which has a very different tensile behavior as compared to the other steels in this study. This trend is again similar to the one observed in Figure 5.12.

Figure 5.13 Hole expansion ratio as a function of the product of true fracture strain and terminal n-value for all steels, with the exception of 22MnB5.

Figure 5.15 shows the correlation between HER and the product of TFS and YS/UTS ratio for all steels. It must be noted that for 22MnB5 the YS is the UTS, providing a YS/UTS ratio of 1. It is also worth noting that it is the one parameter where 22MnB5 fits the overall trend. This parameter was chosen based on the YS/UTS ratio having an inconsistent correlation to HER. The product of TFS and YS/UTS ratio correlates well to HER, where HER increases as TFS*(YS/UTS) increases.

While 22MnB5 did not fit the trend observed in Figure 5.11, it does fit the one in Figure 5.15. Including TFS into YS/UTS corrected 22MnB5 being an outlier. YS/UTS ratio individually does not account for the complex microstructural interactions during local and global formability. It is necessary to further explore YS/UTS and its effects of microstructure, or viceversa.
Figure 5.14 Hole expansion ratio as a function of the product of true fracture strain and post-uniform elongation for all steels.

Figure 5.15 Hole expansion ratio as a function of the product of true fracture strain and YS/UTS ratio for all steels.

Figure 5.16 shows the correlation between HER and the TFS/UTS ratio for all steels. This parameter was chosen in order to express a relationship between the local formability and strength of the material, especially after observing in literature that there were inconsistencies for
DP steels when HER correlated to UTS. It was previously shown that HER correlates to some extent for both the UTS and TFS individually, and there is a good correlation with the combined parameter where the HER increases with increasing TFS/UTS ratio. With the exception of 22MnB5, which is consistently an exception for some of these correlations, there is an obvious linearity to this relationship.

Correlation coefficients for relationships between HER and tensile properties are presented in Table 5.2. The strongest relationships are highlighted. There is a positive linear relationship between HER and TFS with a correlation coefficient of 0.78. There are even stronger relationships between HER and the combination of TFS and other tensile properties such as the following: 0.89 for TFS*(YS/UTS), 0.92 for TFS*n, 0.88 for TFS*P-U El, and 0.85 for TFS/UTS. They support the trends observed in the figures. Appendix C provides some of the correlations that did not result in any significant trends. The figures in Appendix C contain results from both the microstructural and tensile comparisons to HER. Some of the correlations shown there follow work presented in literature that did not work for the present study.

Figure 5.16 Hole expansion ratio as a function of the product of TFS/UTS ratio for all steels.
Table 5.2 – Correlation Coefficients between HER and Tensile Properties

<table>
<thead>
<tr>
<th></th>
<th>HER</th>
<th>UTS</th>
<th>YS/UTS</th>
<th>True εf</th>
<th>P-U El</th>
<th>TFS*(YS/UTS)</th>
<th>TFS*n</th>
<th>TFS*P-U El</th>
<th>TFS/UTS</th>
</tr>
</thead>
<tbody>
<tr>
<td>HER</td>
<td>1.00</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>UTS</td>
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<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>YS/UTS</td>
<td>0.45</td>
<td>-0.19</td>
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<td></td>
<td></td>
<td></td>
<td></td>
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<td></td>
</tr>
<tr>
<td>True εf</td>
<td>0.78</td>
<td>-0.16</td>
<td>-0.04</td>
<td>1.00</td>
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<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>P-U El</td>
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<td>-0.74</td>
<td>0.20</td>
<td>-0.09</td>
<td>1.00</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>TFS*(YS/UTS)</td>
<td>0.89</td>
<td>-0.23</td>
<td>0.39</td>
<td>0.90</td>
<td>0.00</td>
<td>1.00</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>TFS*n</td>
<td>0.92</td>
<td>-0.53</td>
<td>0.80</td>
<td>0.88</td>
<td>0.34</td>
<td>0.91</td>
<td>1.00</td>
<td></td>
<td></td>
</tr>
<tr>
<td>TFS*P-U El</td>
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<td>0.03</td>
<td>0.90</td>
<td>0.36</td>
<td>0.84</td>
<td>0.94</td>
<td>1.00</td>
<td></td>
</tr>
<tr>
<td>TFS/UTS</td>
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<td>-0.35</td>
<td>0.00</td>
<td>0.98</td>
<td>0.06</td>
<td>0.90</td>
<td>0.92</td>
<td>0.94</td>
<td>1.00</td>
</tr>
</tbody>
</table>

5.4 Correlation of Angular Stretch Bend Results to HER and Tensile Properties

Figures 5.17a – 5.17b show the correlation between HER and height at failure and reduction in area, respectively, from the ASB results. It is observed in Figure 5.17a that HER generally increases with increasing height at failure. 22MnB5 is the exception to the observed trend, which does not stay in accordance with HER results. The results observed from height at failure correlate more with total elongation rather than with post-uniform elongation or HER. In Figure 5.17b there is an even stronger correlation where HER increases with increasing reduction in area. These correlations between HER and ASB results support the previous correlations that involve local formability measurements from tensile properties where reduction in area is comparable to true fracture strain. Figure 5.17a shows a correlation where height at failure is comparable to elongation, indicating that it involves overall formability rather than local.

Figure 5.18 shows the correlation between height at failure and total tensile elongation for all steels. It can be observed that height at failure correlates well with total tensile elongation where height at failure increases with increasing total elongation. This trend is to be expected since the stretch bend test is in a way a tensile test where the sample is stretch along with super imposed bending. The TBF980 outperforms the observed trend in Figures 5.17a and 5.18, which could be caused by the nature of its microstructure where the retained austenite transforms into martensite upon deformation.
Figure 5.17  Relationship between (a) HER and height at failure, and between (b) HER and reduction in area for all steels.

Figure 5.18  Relationship between height at failure and total tensile elongation for all steels.
5.5 Discussion Summary

Upon analysis of results there were correlations developed between microstructure, tensile properties, angular stretch bend results and hole expansion results. HER did not correlate directly to individual measurements for contiguity of martensite, size of martensite colonies, and mean free distance between martensite colonies. However, HER had a direct correlation to a combined measurement or parameter that consisted of the product of the mentioned three measurements. HER also correlated directly to carbon content in martensite and number of martensite colonies per unit area. When compared to tensile properties, it was found that HER has a direct correlation to TFS. HER exhibited some correlations to combined parameters of TFS and other tensile properties that represent in different ways measurements of local formability and work-hardening. The ASB results were comparable to tensile results in terms of ductility and local formability, and they supported the tensile results when correlated to HER.
CHAPTER 6: SUMMARY AND CONCLUSIONS

Sheared edge formability of advanced high strength steels (AHSS) is influenced by shearing conditions, microstructure, and tensile properties. The work presented in this thesis was initiated with the purpose of investigating some of the correlations between sheared edge formability and various material factors. Seven commercially produced steels with tensile strengths within 1000 ± 100 MPa were evaluated: five dual-phase (DP) steels with different chemical compositions and varying microstructural features, one trip-aided bainitic ferrite (TBF) steel, and one press-hardened steel tempered to a tensile strength within the desired range. These steels were evaluated by microstructural characterization, sheared edge characterization, tensile testing, angular stretch bend testing, and hole expansion testing.

Microstructural characterization provided information of the as-received material such as phases present and volume fractions. To get a better understanding of the influence of microstructure on sheared edge formability, a more focused approach was used. Quantitative stereology measurements were made for the DP steels which provided information regarding martensite size and distribution. Crack propagation was investigated for tested hole expansion samples, and it was determined that crack propagation occurred through the martensite for the higher HER steels and both through ferrite and along the ferrite-martensite boundaries for the DP980 D steel, which had the lowest HER.

Sheared edge characterization provided information of the influence of shearing conditions on sheared edge formability. The work-hardened zone behind the shear face, known as the shear affected zone (SAZ), was evaluated. The magnitude and depth of the SAZ were determined for all steels, but did not correlate to hole expansion ratio (HER). The shear face was also characterized by measuring the rollover, burnish, fracture and burr regions.

Tensile testing provided information regarding the influence of tensile properties on sheared edge formability. Having steels with a similar ultimate tensile strength (UTS) but with slight differences in other properties such as post-uniform elongation, n-value, YS/UTS ratio, and true fracture strain provided insight on which tensile properties are significant factors to sheared edge formability.
Hole expansion testing provided the results necessary to evaluate sheared edge formability. These results were used in order to compare and correlate to results determined from microstructural characterization, sheared edge characterization, and tensile testing. Hole expansion testing was performed on all steels with a conical punch in the burr-up position. These tests were performed until failure, which is determined by a through-thickness crack. Significant differences in HER were observed for the steels in this study. Multiple cracks were observed in samples throughout testing without following any preferential orientations. Partial cracks were observed as well, all of them initiating in the top surface of the sample (where the fracture and burr regions are located). Angular stretch bend testing provided another formability measurement, which consists of a combination of bending and stretching of steel samples. Some of the results were compared and correlated to HER and tensile properties.

Several conclusions were determined from this project and are listed below.

1. Sheared edge formability is influenced by the martensite in DP steels. Quantitative stereology measurements provided results that showed martensite size and distribution affect HER. The overall trend determined is that HER increases with more even martensite dispersion throughout the microstructure. This correlation involves a combined measurement of martensite colony size, contiguity, and mean free distance between martensite colonies, as well as a single measure for number of colonies per unit area. HER also correlates to carbon content of martensite. HER decreases with increasing carbon content of martensite.

2. Shear face analysis provided results that showed the size of the fracture and burr region correlates with HER. It was determined that HER decreases with increasing size of fracture and burr region. The HER and size of the fracture region are possibly correlated because a larger fracture region will have more defects and/or microcracks as a result of the shearing process, which facilitates premature cracking in sheared edge formability.

3. Sheared edge formability is directly correlated to true fracture strain (TFS). HER increases with increasing true fracture strain. This trend was also supported by different combinations of indexes with TFS and other tensile properties that represent strain hardening and local formability in different ways.
ASB results provided correlations involving local formability where reduction in area is comparable to true fracture strain, as well as a correlation where height at failure is comparable to elongation, indicating that it involves overall formability rather than local.
CHAPTER 7: FUTURE WORK

In order to support and expand the research started in this thesis, it would be beneficial to perform additional work in the following areas:

- Perform microstructural characterization using more advanced techniques. Instead of determining microstructural feature and measurements with light optical micrographs, it would be significantly more accurate to use scanning electron micrographs and investigate the effect of variations of individual martensite crystallographic units, such as packets and blocks, instead of colonies.

- Follow the approach used in this thesis, but use different materials. Instead of focusing on DP steels of similar strength, it would be interesting to work with other microstructures, such as quenched and partitioned (Q&P) steels or transformation-induced plasticity (TRIP) steels. It is suggested to maintain the focus on a single type of steel and within a similar strength range, to observe the sheared edge formability based on specific microstructures associated with a given class of steel.

- Assess delayed fracture response of some of these DP steels. During angular stretch bend testing, a couple DP steels showed short-time delayed fracture. Upon test completion and no punch motion, fracture occurred between 1-3 seconds after completion of the test.

- Follow the approach used in this thesis, but use milled edge conditions. Instead of performing hole expansion tests with samples containing a shear affected zone (SAZ), it would be interesting to work with the same microstructures without having a work-hardened zone at the edge. It is suggested to use the same steels used in this thesis in order to observe their difference, if any, in HER.
REFERENCES CITED


APPENDIX A: PRELIMINARY HOLE EXPANSION STUDY

This appendix contains results of a preliminary study and includes hole expansion, tensile, and angular stretch bend testing of steels of various microstructures and strength levels. An initial microstructural and sheared edge characterization was performed, as well as a heat treatment for a press-hardened steel. There was no in-depth microstructural analysis. Table A.1 shows the steel thicknesses (in mm) and chemical compositions (in wt pct) of the sheet steels used in this preliminary study. The purpose of the preliminary study was to become familiar with the testing methods and to examine a wide range of AHSS with a wide range of microstructures.

Table A.1 – Chemistry of the Steels Used for the Preliminary Study (wt. pct.)

<table>
<thead>
<tr>
<th>Steel</th>
<th>Thickness (mm)</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>Ti</th>
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<td>-</td>
<td>-</td>
<td>-</td>
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<td>TRIP700</td>
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<td>1.63</td>
<td>0.066</td>
<td>0.017</td>
<td>0.026</td>
<td>-</td>
<td>0.005</td>
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<td>0.36</td>
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<tr>
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<td>-</td>
<td>-</td>
<td>-</td>
</tr>
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<td>0.243</td>
<td>0.01</td>
<td>0.19</td>
<td>0.01</td>
<td>0.04</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Steel</th>
<th>Nb</th>
<th>V</th>
<th>Al</th>
<th>N</th>
<th>S</th>
<th>P</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP600</td>
<td>-</td>
<td>-</td>
<td>0.032</td>
<td>-</td>
<td>0.0038</td>
<td>0.009</td>
<td>-</td>
</tr>
<tr>
<td>TRIP700</td>
<td>-</td>
<td>-</td>
<td>1.39</td>
<td>0.0018</td>
<td>0.0005</td>
<td>0.021</td>
<td>0.022</td>
</tr>
<tr>
<td>DP980</td>
<td>0.002</td>
<td>0.001</td>
<td>0.043</td>
<td>0.005</td>
<td>0.006</td>
<td>0.012</td>
<td>0.03</td>
</tr>
<tr>
<td>DP1180</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.0048</td>
<td>0.001</td>
<td>0.009</td>
<td>-</td>
</tr>
<tr>
<td>CR15B21</td>
<td>0.01</td>
<td>0.01</td>
<td>0.043</td>
<td>0.004</td>
<td>0.01</td>
<td>0.013</td>
<td>0.02</td>
</tr>
</tbody>
</table>

The preliminary study has all test specimens oriented in the transverse direction. All specimens were sheared to the dimensions stated by ASTM standard E8 [39], with the exception of the tensile specimens which were cut via EDM wire. The CR15B21 steel was heat treated using the salt pot furnaces at CSM. The samples were held at a temperature of 880°C for 30 s. followed by a water quench. A martensitic microstructure was desired in order to evaluate its sheared edge formability behavior. Figure A.1 shows the microstructures of the steels listed in Table A.1, including initial and final microstructures for CR15B21. The DP steels have a ferritic matrix with dispersed martensite islands with the volume fractions of each phase varying accordingly to grade. The TRIP steel contains polygonal ferrite and also likely has bainitic ferrite.
and retained austenite islands. The CR15B21 steel has a ferritic-pearlitic microstructure initially, but after heat treatment it has a martensitic microstructure.

Figure A.1  LOM micrographs of (a) DP600, (b) TRIP700, (c) DP980, (d) DP1180, (e) CR15B21 pre-heat treatment, and (f) CR15B21 post-heat treatment. Etched with 2% nital.

Figures A.2a – A.2e are plots of microhardness as a function of distance from the edge for the DP600, TRIP700, DP980, DP1180 and CR15B21 steels, respectively. The DP1180 steel exhibits a small SAZ, where the base material hardness is reached within 300 µm. The CR15B21, prior to heat treatment, exhibits no SAZ at all. Even though there was a shear face produced during the shearing process, a significant work-hardening effect is not observed in these two steels.
Figure A.2  Micro hardness profiles for (a) DP600, (b) TRIP700, (c) DP980, (d) DP1180, (e) CR15B21 steels measured in the fracture region of the sheared edge.
Table A.2 contains the relative depth of the SAZ, which is the ratio of the depth of the SAZ to the thickness of the material. The lower strength steels have a higher relative depth, 0.45 for DP600 and 0.41 for TRIP700. As material strength increases, the relative depth of the SAZ decreases as can be observed for DP980 with 0.35 and DP1180 with 0.19.

Table A.2 – Size and Relative Depth of the Shear Affected Zone

<table>
<thead>
<tr>
<th>Steel</th>
<th>Depth SAZ (µm)</th>
<th>Relative Depth of SAZ</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP600</td>
<td>626 ± 83</td>
<td>0.45</td>
</tr>
<tr>
<td>TRIP700</td>
<td>623 ± 82</td>
<td>0.41</td>
</tr>
<tr>
<td>DP980</td>
<td>455 ± 83</td>
<td>0.35</td>
</tr>
<tr>
<td>DP1180</td>
<td>267 ± 75</td>
<td>0.19</td>
</tr>
</tbody>
</table>

Table A.3 reports the tensile properties for all steels. All tensile properties were obtained using specimens with tensile axis transverse to the rolling direction, and the strength expectations for each grade were met in most cases. Only the TRIP700 is slightly below the expected UTS. Figure A.3 shows representative tensile curves for each of the steels. The DP steels and the boron steel have continuous yielding behavior, with a high initial work-hardening rate. The TRIP700 has a more defined yield point, with a lower initial work hardening rate but a higher work hardening rate at larger strains, as expected for a TRIP alloy. Data in Table A.3 were used to compare with hole expansion results, in an attempt to correlate sheared edge formability with tensile properties. It must be noted that the tensile properties for CR15B21 are representative of the post-heat treatment condition.

Table A.3 – Tensile Properties for Preliminary Steels

<table>
<thead>
<tr>
<th>Steel</th>
<th>Total Elongation (pct)</th>
<th>Uniform Elongation (pct)</th>
<th>UTS (MPa)</th>
<th>YS (MPa)</th>
<th>n</th>
<th>K (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP600</td>
<td>23.6 ± 0.3</td>
<td>15.3 ± 0.1</td>
<td>678 ± 4</td>
<td>301 ± 10</td>
<td>0.187 ± 0.002</td>
<td>1146 ± 8</td>
</tr>
<tr>
<td>TRIP700</td>
<td>38.2 ± 0.7</td>
<td>31.4 ± 0.3</td>
<td>696 ± 0</td>
<td>376 ± 7</td>
<td>0.199 ± 0.002</td>
<td>1124 ± 4</td>
</tr>
<tr>
<td>DP980</td>
<td>9.6 ± 0.2</td>
<td>6.1 ± 0.1</td>
<td>1009 ± 1</td>
<td>552 ± 4</td>
<td>0.180 ± 0.002</td>
<td>1870 ± 15</td>
</tr>
<tr>
<td>DP1180</td>
<td>9.5 ± 0.1</td>
<td>5.9 ± 0.1</td>
<td>1242 ± 2</td>
<td>620 ± 9</td>
<td>0.212 ± 0.006</td>
<td>2605 ± 55</td>
</tr>
<tr>
<td>CR15B21*</td>
<td>6.0 ± 0.3</td>
<td>4.1 ± 0.3</td>
<td>1890 ± 14</td>
<td>1365 ± 27</td>
<td>0.204 ± 0.007</td>
<td>3978 ± 110</td>
</tr>
</tbody>
</table>

Figure A.3  Representative engineering stress versus engineering strain curves for preliminary steels.

Table A.4 shows the Δ HV values between the indent closest to the surface, and the base material for the steels that exhibited presence of a SAZ. Table A.4 also reports the UTS-YS, YS/UTS and n-values, which are possible measures of the work-hardening during a tensile test. As the Δ HV values decrease, the UTS-YS values increase, with the exception of TRIP700. As the material strength increases, the relative depth of the SAZ decreases, but the work-hardening capacity during a tensile test appears to increase. There seems to be an inverse relationship between work-hardening capacity during a tensile test and the amount of work-hardening induced in a shearing process as measured by the micro hardness change. Comparing the Δ HV values to the YS/UTS ratio shows that as Δ HV decreases, the YS/UTS ratio increases, with DP1180 being the exception.

Table A.4 – Comparison of Various Work-Hardening Measurements

<table>
<thead>
<tr>
<th>Steel</th>
<th>Δ HV*</th>
<th>UTS-YS</th>
<th>YS/UTS</th>
<th>n</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP600</td>
<td>124</td>
<td>377</td>
<td>0.44</td>
<td>0.187</td>
</tr>
<tr>
<td>TRIP700</td>
<td>107</td>
<td>320</td>
<td>0.54</td>
<td>0.199</td>
</tr>
<tr>
<td>DP980</td>
<td>39</td>
<td>457</td>
<td>0.55</td>
<td>0.180</td>
</tr>
<tr>
<td>DP1180</td>
<td>22</td>
<td>622</td>
<td>0.50</td>
<td>0.212</td>
</tr>
</tbody>
</table>

*Difference in hardness between sheared edge and base material.
Table A.5 lists the HER results for the steels in the preliminary study. Figure A.4 shows the correlation between HER and UTS. As expected, the HER value decreases with increasing strength. The DP980 steel has a lower than expected HER based on the trend from the other steels examined. All the cracks on DP980 were parallel to the rolling direction. The two different HER values for CR15B22 correspond to the heat treatments being performed a month before testing and same day before testing, the latter being represented with the CR15B22* designation.

Table A.5 – HER of the Steels Used for the Preliminary Study

<table>
<thead>
<tr>
<th>Sample</th>
<th>DP600</th>
<th>TRIP700</th>
<th>DP980</th>
<th>DP1180</th>
<th>CR15B21</th>
<th>CR15B22</th>
<th>CR15B22*</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>61.5</td>
<td>41.8</td>
<td>24.4</td>
<td>34.8</td>
<td>1.4</td>
<td>0.3</td>
<td>4.4</td>
</tr>
<tr>
<td>2</td>
<td>46.1</td>
<td>44.9</td>
<td>25.2</td>
<td>30.7</td>
<td>0.5</td>
<td>0.7</td>
<td>7.5</td>
</tr>
<tr>
<td>3</td>
<td>59.3</td>
<td>45.2</td>
<td>26.5</td>
<td>44.8</td>
<td>0.3</td>
<td>0.3</td>
<td>11.0</td>
</tr>
<tr>
<td>4</td>
<td>59.2</td>
<td>47.1</td>
<td>24.5</td>
<td>34.8</td>
<td>0.4</td>
<td>0.5</td>
<td>16.0</td>
</tr>
<tr>
<td>5</td>
<td>58.9</td>
<td>40.8</td>
<td>23.7</td>
<td>27.1</td>
<td>0.5</td>
<td>0.5</td>
<td>7.6</td>
</tr>
<tr>
<td>6</td>
<td>45.6</td>
<td>45.8</td>
<td>23.7</td>
<td>35.4</td>
<td>0.2</td>
<td></td>
<td></td>
</tr>
<tr>
<td>7</td>
<td>54.1</td>
<td>42.9</td>
<td>25.7</td>
<td>42.0</td>
<td>0.3</td>
<td></td>
<td></td>
</tr>
<tr>
<td>8</td>
<td>53.7</td>
<td>46.1</td>
<td>26.4</td>
<td>36.8</td>
<td>0.1</td>
<td></td>
<td></td>
</tr>
<tr>
<td>9</td>
<td>58.9</td>
<td>45.0</td>
<td>24.0</td>
<td>29.3</td>
<td>0.6</td>
<td></td>
<td></td>
</tr>
<tr>
<td>10</td>
<td>55.3</td>
<td>46.0</td>
<td>27.4</td>
<td>38.7</td>
<td>0.2</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Average</td>
<td>55.3</td>
<td>44.6</td>
<td>25.1</td>
<td>35.4</td>
<td>0.5</td>
<td>0.5</td>
<td>9.3</td>
</tr>
<tr>
<td>Std. Dev.</td>
<td>5.3</td>
<td>1.9</td>
<td>1.2</td>
<td>5.2</td>
<td>0.3</td>
<td>0.2</td>
<td>3.9</td>
</tr>
</tbody>
</table>

Figure A.4 Relationship between HER and UTS for the preliminary set of steels.
Other correlations between HER and tensile properties that were examined and which appeared to provide insight include YS/UTS ratio and post-uniform elongation. Figure A.5a shows the correlation between HER and YS/UTS, where it can be observed that as YS/UTS ratio increases HER decreases. YS/UTS ratio is a rough measure of work hardening of the steel. Larger values of YS/UTS are for steels with less work hardening capability. The trend seen in Figure A.5a can be interpreted as an increase in HER with higher work hardening capacity. It is interesting to see an opposite trend to what Jin et al [27] showed, making YS/UTS an inconsistent correlation. Figure A.5b shows the correlation between HER and post-uniform elongation, where it can be observed that as post-uniform elongation increases so does HER. Other tensile properties such as n-value and total elongation did not correlate well to HER for the steels in this preliminary investigation.

![Figure A.5 Relationship between (a) HER and YS/UTS and between (b) HER and post-uniform elongation for the preliminary set of steels.](image)

Figure A.6 shows the relationship between R/t ratio and height at failure for DP and TRIP steels for the angular stretch bend tests. It can be observed that as R/t ratio increases, the height at failure increases as well. This relationship was found for all the steels with the exception of DP980, but it must be noted that fracture occurred at the drawbead and not at the punch nose due to a higher bending severity at that location for the DP980 at 2.5 mm and 5.0 mm radii punches, and for DP1180 at 5.0 mm radius punch. Figure A.7 shows the relationship between R/t ratio and
reduction in area. It can be observed that as R/t ratio increases, the reduction in area decreases. It must be noted that there were no reduction in area measurements taken on the samples that fractured at the drawbead. It can be observed from Figures A.6 and A.7 that the strength of the material does not correlate to either height at failure nor reduction in area. Looking at the tensile data for these steels, height at failure can be correlated to post-uniform elongation. Both DP600 and TRIP700 provide an interesting combination of results here, with height at failure increasing but reduction in area decreasing with increasing R/t ratio.

Figure A.6  Relationship between R/t ratio and height at failure for DP and TRIP steels. Open symbols represent samples that fractured at the drawbead.

Figure A.7  Relationship between R/t ratio and reduction in area for DP and TRIP steels.
This combination of results could be attributed to the work hardening behavior of these materials which allows for significant amount of plastic deformation before fracture occurs. Comparing these results to HER, it can be observed that height at failure is consistent with HER values for the respective steel grade. This correlates back to post-uniform elongation as well. In order to correlate HER results to reduction in area measurements, additional testing is required for ASB samples of higher strength grades with a smaller diameter drawbeads to avoid fracture in that location and have fracture on the punch nose.
APPENDIX B: INFLUENCE OF SHEAR AFFECTED ZONE ON
ANGULAR STRETCH BEND TEST

This appendix contains results of a study performed on DP600 steel on angular stretch bend test with two different edge conditions. These edge conditions are sheared and milled. The edges for both conditions were characterized by creating a micro hardness profile. ASB tests were performed with 1.0, 2.5, and 5.0 mm radii punches. Three samples were tested per radius. Load versus displacement curves from the ASB test were used to determine height at failure and maximum load values. Reduction in area measurements were determined by using a point micrometer and measuring the thickness near the fracture surface.

Figure B.1 shows the difference in micro hardness profiles for different edge conditions in DP600. Figure B.1a shows the milled edge condition where essentially no SAZ exists. There is a slight work-hardened effect near the edge, but the magnitude of change in micro hardness between the edge and base material is only ~20 HV. Figure B.1b shows the sheared edge condition where a SAZ is present. The magnitude of change in micro hardness between the edge and base material is approximately 100 HV.

Figure B.1  Average micro hardness profiles for DP600 (a) milled edge and (b) sheared edge conditions.
Figure B.2 shows representative load vs. displacement curves for milled and sheared edge conditions tested with the 1.0 mm radius punch. It can be observed that there is a difference between sheared and milled edge curves in both maximum load and height at failure. Table B.1 summarizes height at failure and maximum load values for all conditions. It can be observed that there is a difference in all punch radii between milled and sheared edge conditions, where sheared specimens resulted in a lower height at failure and higher maximum load value. This is observed in Figure B.2 as well.

Figure B.2 Representative load vs. displacement curves of a single test specimen for milled and sheared edge conditions. Specimens were tested with the 1.0 mm radius.

Table B.1 – Summary of ASB Results for DP600

<table>
<thead>
<tr>
<th>Punch Radius (mm)</th>
<th>Edge Condition</th>
<th>Height at Failure (mm)</th>
<th>Maximum Load (kN)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0</td>
<td>Milled</td>
<td>13.7 ± 0.05</td>
<td>12.80 ± 0.01</td>
</tr>
<tr>
<td>2.5</td>
<td>Milled</td>
<td>16.4 ± 0.05</td>
<td>15.90 ± 0.01</td>
</tr>
<tr>
<td>5.0</td>
<td>Milled</td>
<td>19.2 ± 0.16</td>
<td>19.14 ± 0.05</td>
</tr>
<tr>
<td>1.0</td>
<td>Sheared</td>
<td>13.6 ± 0.04</td>
<td>13.33 ± 0.00</td>
</tr>
<tr>
<td>2.5</td>
<td>Sheared</td>
<td>16.2 ± 0.07</td>
<td>16.79 ± 0.21</td>
</tr>
<tr>
<td>5.0</td>
<td>Sheared</td>
<td>19.0 ± 0.15</td>
<td>20.22 ± 0.27</td>
</tr>
</tbody>
</table>
There is also a difference in where failure initiates between milled and sheared edge conditions. Figure B.3a shows the top view of a representative milled edge specimen tested with a punch radius of 5.0 mm. Figure B.3b shows a side view of the edge of the same specimen. Figure B.4a shows a top view of a representative sheared edge specimen tested with a punch radius of 1.0 mm. Figure B.4b shows a side view of the edge of the same specimen. It can be observed that in the milled edge specimen the crack initiated at the center and propagated out to the edges, while in the sheared edge specimen the crack initiated at the edge and propagated towards the center. This difference can be attributed to the SAZ and the shear face where the fracture region and burr have a significant amount of damage caused by the shearing process and potentially create micro cracks.
APPENDIX C: INCONCLUSIVE CORRELATIONS

This appendix contains correlations investigated between hole expansion ratios, microstructural measurements, and tensile properties that did not provide any useful trends nor results for the steels described in the Experimental Procedure chapter. Some of these correlations are shown in literature in studies of varying strengths and microstructures that did not work for this focused study with materials of a single strength and of a similar microstructure.

Figure C.1 HER as a function of (a) product of grain size and mean free distance, (b) mean free distance to martensite volume fraction ratio, (c) martensite volume fraction to contiguity ratio, and (d) product of true fracture strain and contiguity.
Figure C.2 HER as a function of (a) n-value, (b) yield strength, (c) relative depth of SAZ, and (d) total elongation.