HOLE EXPANSION PERFORMANCE AND THE RESISTANCE TO CRACK PROPAGATION IN HIGH STRENGTH DP AND QP STEELS

by
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ABSTRACT

The automotive industry is in constant search for steels that can reduce vehicle weight to help satisfy the stringent fuel economy and safety standards. Advanced high strength steels (AHSS) have become a viable option due to their excellent mechanical properties; however, these high strength steels exhibit edge cracking during stamping processes, failing at strains lower than those predicted by forming limit diagrams (FLD). Therefore, test methods such as hole expansion testing (HET) have been implemented to help understand edge failure in these AHSS. Characterizing the mechanical properties and microstructures of AHSS becomes crucial to understanding the hole expansion ratio (HER) determined through HET. In this study, we focus on the effects of mechanical properties and microstructure on HER performance. We also focus on a crack resistance parameter known as the crack tip opening angle (CTOA). The materials of interest are high strength dual phase (DP) and quenched and partitioned (QP) steels.

The results showed that the difference in hardness between martensite and ferrite proved to have the most effect on the HER for the DP steels. For the QP steels, there did not seem to be a mechanical property that was tested in this study that correlated with the HER. The microstructural aspects that contributed to the HER performance were the martensite morphology and retained austenite (RA) stability for the DP and QP steels, respectively. A martensite network along ferrite grain boundaries proved to be beneficial in the DP steels studied, while a steady, continuous amount of RA transformation throughout deformation was beneficial to HER performance in the QP steels. Finally, CTOA correlated well with HER and proved that a crack resistance parameter was important to begin understanding HER.
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CHAPTER 1
INTRODUCTION

Advanced high strength steels (AHSS) have gained attention in the automotive industry due to their excellent combination of strength and ductility and are a viable option for the increasing fuel economy and safety standards set out by the United States Department of Transportation [1-3]. Dual Phase (DP) and Quenched and Partitioned (QP) steels are 3rd generation AHSS that are of interest. DP steels have already helped satisfy the fuel economy and safety standards and have been implemented throughout the automotive industry. QP is a fairly new steel processing route that has been shown to have the potential to produce materials with excellent combinations of strength and ductility [4-5]. However, as higher strength levels are produced, the moderate fracture strains of these AHSS is often a limitation when the material undergoes extensive deformation during forming operations or when in service.

Typically, forming limit diagrams (FLD) are used to predict material performance during forming operations; however, FLD are not accurate with AHSS. AHSS exhibit failure at strains lower than those predicted on forming limit curves, where edge cracking occurs during the stamping process due to sheared edge effects [6-8]. Therefore, stretch-flangeability is important to the automotive industry and is typically measured by hole expansion testing (HET). The hole expansion ratio (HER) is determined from HET and has been used to clarify sheared edge formability. However, a better understanding of HER is needed as it depends on many factors such as edge condition and microstructure [7].

This study focuses on gaining a better understanding and correlating HER to mechanical properties and microstructural characteristics in high strength DP and QP steel grades. HET was also analyzed through various punch geometries (conical and flat bottom punch) and edge conditions (sheared and machined). A crack resistance test method known as crack tip opening angle (CTOA) testing was performed to analyze, better characterize, and aid in understanding HER. By taking all of this into account, this study hopes to provide a solution to the formability issues that the automotive industry is experiencing with AHSS.

The ensuing chapters begin with an introduction of the materials, a review of the microstructural effects on mechanical and fracture properties, HET, and finally introducing CTOA testing. We will then present the experimental procedures and results of this study that
will be tied into a discussion section along with the conclusions taken from this study. The final chapter will consist of future work and recommendations based on the findings of this study.
2.1 Overview

The automotive industry is in need of materials with higher strengths that can maintain levels of formability that satisfy component manufacturing and service requirements. DP and QP steels have shown excellent tensile properties, although there is still a need to understand the fracture behavior of these complex microstructures to help predict performance during manufacturing and in service. This chapter discusses processing routes, microstructures, and mechanical properties of DP and QP steel grades. The microstructural analysis specifically focuses on martensite and retained austenite (RA) for DP and QP steels, respectively. It also delves into the effects of testing conditions and microstructure during hole expansion testing (HET) and introduces a fracture toughness measurement for sheet steels known as the crack tip opening angle (CTOA).

2.2 DP Steels

2.2.1 DP Steel Processing

DP steels consist of a soft ferrite matrix and islands of hard martensite [9], and sometimes RA depending on how the material is processed [10]. This microstructure can be achieved through different processing routes such as hot rolling (Figure 2.1(a)) or intercritical annealing (IA) (Figure 2.1(b)). The hot rolling process involves the material being hot rolled to the austenite phase region followed by controlled cooling to the austenite + ferrite phase region where the ferrite phase fraction can be controlled. The final step in this process is quenching where most of the remaining austenite transforms to martensite.

The IA processing route involves taking a low or medium carbon steel to the intercritical temperature range (between $A_1$ and $A_3$), where austenite can form in the ferrite matrix followed by cooling directly to room temperature, transforming the austenite to martensite [11]. The IA temperature and the time held at this temperature both determine the austenite volume fraction [12]. At slower cooling rates, it is possible to form non-martensitic constituents (NMC) such as bainite. At higher cooling rates, the austenite will fully transform to martensite producing a typical DP microstructure.
Figure 2.1  (a) Hot rolling and (b) IA are two DP steel processing routes. They are both plotted as temperature versus time plots with the relevant phases and temperatures labeled. A and γ is austenite, F and α is ferrite, and M and α' is martensite. [13, 14]

In DP steels, it is generally understood that the ferrite phase contributes to the ductility of the steel while the martensite provides the strength [12]. In comparison to high strength low alloy (HSLA) steels with similar tensile strengths, DP steels tend to have a lower yield strength (YS) but a higher amount of ductility, high initial work hardening rate, and high ultimate tensile strength to yield strength ratio (UTS/YS) making them appealing candidates for the automotive industry [9, 12]. Figure 2.2 shows a comparison of a plain carbon steel, two HSLA steels and a DP steel.

Figure 2.2  Engineering stress-strain curves for HSLA steels (SAE 980X and SAE950X), DP steel (GM 980X), and plain carbon steel. SAE is the Society of Automotive Engineers designation for HSLA steels and the GM 980X is a dual-phase steel developed by GM. [15]
2.2.2 Effects of Martensite Volume Fraction on Mechanical and Fracture Properties

The strength of martensite is dependent upon its carbon content. The volume fraction and carbon content is dependent upon the IA temperature and holding time [16]. Davies performed experiments on Fe-Mn-C alloys with DP microstructures and proved that at a given IA temperature, the flow strength increased with increasing carbon content. He also showed that at a given carbon content, higher IA temperatures led to higher flow strengths [16], shown in Figure 2.3(a). Figure 2.3(b) shows the flow stress and tensile strength as a function of martensite volume fraction (MVF) for materials with carbon contents ranging from 0.06-0.29 wt. pct. [16]. Davies' reasoning was that the martensite content is a function of the carbon content and the IA temperature and therefore, focused on the MVF. Ferrite grain size was also analyzed and results showed that fine-grained materials led to higher YS and UTS compared to a coarse-grained material.

Lai et al. obtained similar results in their experiments with a DP steel containing 0.1 wt. pct. C and 3.5 wt. pct. Mn, where the MVF was varied through different IA holding times [17]. Figure 2.4 indicates that as MVF increases (maximum of 37 pct.), the flow strength and tensile strength increase while the uniform elongation (UE) decreases. Lai et al. also determined
from uniaxial tension tests that the fracture strain, the maximum amount of strain a material can withstand before fracture, significantly decreased with an increase in MVF. The DP steel with a MVF of 15 pct. had a fracture strain of 1.06, while the steel with a volume fraction of 37 pct. had a fracture strain of 0.33 [17]. The effect of MVF on the fracture strain is considerable which can become problematic during forming operations.

Bag et al. performed experiments on the influence of MVF on the tensile and impact properties of DP steels [18]. The as-received material was a hot rolled plate in the quench and tempered condition with 0.16 C content (wt.pct.). The samples were heat treated through an intermediate quench (IQ) or step quench (SQ) process. The IQ process involved austenitizing followed by quenching then IA (temperatures between 730 and 850°C) the sample followed by a final quench. The SQ process involved austenitizing followed by cooling to the IA temperature range (760-820°C) and then a final quench. The DP steels processed in this manner led to MVFs up to 78 pct. The SQ specimens consisted of a banded microstructure with a blocky morphology, while the IQ specimens consisted of finely dispersed globular and plate martensite with no evidence of microstructural banding. Both microstructures also contained approximately 2-3 pct. RA. Figure 2.5(a) shows YS, UTS, and Figure 2.5(b) shows total and uniform elongation as a function of MVF for both DP steel conditions. The IQ condition shows a decrease in strength at MVFs greater than 50 pct. and then increases at about 70 pct. It also shows that at MVFs greater than 50 pct., the total and uniform elongation both begin to decline. The SQ condition shows a constant UTS, but an increase in the YS with increasing MVF. Overall, the IQ condition (finely dispersed globular and plate martensite) shows a better combination of strength and ductility in comparison to the SQ condition (banded microstructure with blocky martensite). The finely dispersed globular and plate martensite (IQ) also had higher impact toughness values than SQ specimens. The results from Bag's study focused on the effects of microstructure morphology at a given MVF and therefore, show different results than those performed by Davies who just focused on MVF effects.
The true stress-true strain curves up to necking instability for four DP microstructures with a spheroidized microstructure that were IA with varying amounts of martensite. The MVFs are shown as percentages on the plot. Tensile tests were performed at an engineering strain rate of 0.001 s$^{-1}$. [17]

(a) YS and UTS and (b) UE and TE as function of MVF for SQ and IQ processed DP steels with 0.16 C content (wt. pct.). [18]

MVF has not only been shown to have an effect on mechanical properties, but also on fracture initiation and propagation. Lai et al. studied the effects of MVF on fracture of DP steels with a relatively equiaxed microstructure and martensite islands at ferrite grains boundaries. The martensite became interconnected as the MVF increased. At low volume fractions (approximately less than 15 pct.), there was evidence of fracture nucleating both in the martensitic phase and at the ferrite-martensite interface, with a majority at the ferrite-martensite interface [9]. As the MVF increased (typically greater than 28 pct.), large voids began to nucleate
within the martensite islands. There are also indications of interface fracture, though the voids in the martensite dominate [17]. As the MVF increases, the fracture initiates primarily within the martensite.

2.2.3 Martensite Morphology

The martensite morphology also has an effect on the mechanical properties. Kim and Thomas focused on various martensite morphologies with similar MVFs to compare the effects on mechanical properties and better understand the fracture mechanisms in DP steels. Through three different heat treatments (IQ, IA, and SQ), they created microstructures consisting of fine fibrous martensite, fine globular martensite along ferrite grain boundaries, and coarse martensite. The SQ process led to the highest UTS for a given MVF followed by the IA and IQ samples similar to the results from Bag et al [18], though the SQ morphology showed the lowest amount of ductility. Therefore, the microstructure with the best combination of ductility and strength was that of the fine fibrous ferrite and martensite, which is in agreement with the literature [18-20].

The martensite morphology has an effect on fracture of DP steels. Zhang et al. used two different processing routes (IQ and IA) to create a fine lath morphology (40 pct. MVF) and a blocky morphology (23 pct. MVF) in a commercial DP590 steel [21]. The two morphologies exhibited martensite cracking as well as micro-voids at the ferrite-martensite interface when deformed with the fine martensite more prone to cracking due to having twin substructure, which is typically associated with a higher carbon content. It was concluded that the fine martensite had martensite cracking and the blocky morphology was prone to the formation of micro-voids at the ferrite-martensite interface [21].

2.2.4 Microstructural Banding in DP Steels

During the hot rolling process, pearlite and ferrite are typically arranged in layers where upon heating, austenite forms in the prior pearlitic regions and upon rapid cooling becomes martensite. This can then lead to bands of martensite creating what is known as a "banded microstructure". This banding is typically due to manganese segregation during solidification and can be detrimental to properties such as ductility and impact toughness [22-24], though no significant effects on tensile and yield strength have been shown. Banded microstructures
typically lead to early crack initiation in the martensite due to hardness differences with continuous crack propagation that can lead to early failure in the material [25].

2.3 QP Steels

2.3.1 QP Steel Processing

Quench and partitioning is a heat treatment process introduced by Speer et al. [4]. QP steels are considered to be third-generation AHSS and are becoming a viable option in the automotive industry because of their excellent combination of strength and ductility. The QP process produces a final microstructure of martensite, RA, and possibly ferrite when partial austenization occurs [26]. Figure 2.6 illustrates the two-step QP process. The process begins with either a full austenization or IA step. The steel is then quenched to a temperature between the martensite start temperature ($M_s$) and the martensite finish temperature ($M_f$), transforming some of the austenite to martensite. The martensite at this point will be referred to as the martensite formed on the initial quench ($M_{QT}$). The next step is the partitioning heat treatment where a one-step or two-step process can be implemented. The one-step process uses the $M_{QT}$ temperature for partitioning while the two-step process uses a temperature higher than the $M_{QT}$ temperature. During the partitioning step, carbon from the martensite diffuses into the austenite allowing the austenite to become more thermally stable. Upon the final quench to room temperature, some of the austenite is retained while fresh martensite also forms ($M_{Fresh}$). Processing can also be tailored to include NMC such as bainite.

![Figure 2.6 Schematic of the QP thermal processing concept with expected microstructures at various steps. $C_\gamma$, $C_i$, and $C_m$ are the carbon concentration in austenite, initial carbon concentration and the carbon concentration in martensite, respectively. QT is the quenching temperature and PT is the partitioning temperature. [26]](image-url)
2.3.2 RA Stability in QP Steels

The transformation-induced plasticity (TRIP) effect is the transformation of RA to martensite upon straining. The stability of RA in steels that exhibit the TRIP effect is affected by chemical composition [27], grain size [28], morphology [29], and surrounding microstructure [30]. The TRIP effect increases the work hardening rate and delays necking, thereby improving UE. Therefore, a better understanding of RA stability can aid in designing a microstructure that will provide the desired combination of strength and formability.

RA stability can be analyzed from two different perspectives: 1) stability during heat treatment to produce the highest amount of RA in a starting microstructure and 2) stability during deformation. The stability during heat treatment is primarily dependent upon the chemical composition, which can be locally optimized by heat treatments such as TRIP or QP. The stability during deformation is dependent upon RA composition, grain size, morphology, and the microstructure surrounding the RA.

Austenite stabilizers such as Mn and C lower the eutectoid temperature which widens the temperature range that austenite is stable and also depresses the $M_s$ temperature. Silicon slows cementite precipitation which helps stabilize RA, by providing a source of carbon to enrich the austenite [31]. Therefore, increasing the amount of austenite stabilizers can lead to a greater amount of austenite at room temperature.

The chemical composition can also affect RA stability during straining. Jacques et al. performed experiments on steels with microstructures containing similar amounts of RA, and low and high silicon contents [27]. The microstructures had similar RA grain sizes, though the RA had different carbon contents and the volume fraction of the surrounding phases were different [27]. The austenite in the high silicon steels had a higher carbon content than the low silicon steels due to carbide precipitation and their results showed that increased carbon content reduced the austenite transformation rate during straining. The low silicon steels also followed this trend with the exception of one material and this was attributed to the presence of bainite and martensite surrounding the RA changing the transformation rate. Therefore, it was concluded that the high silicon steels had a slower RA transformation rate during straining because the higher carbon enriched RA required a higher amount of strain to transform to martensite [27].

The RA grain size also plays a role in stability where grain sizes larger than 1 micron have been shown to be less stable than those smaller than 1 micron [32, 33]. Brandt and Olson
performed experiments on low alloy steels and determined that both carbon enrichment and smaller grain sizes were important to RA stabilization [34]. They concluded that smaller RA grain sizes had less potential to transform and required a much higher driving force to transform to martensite. Therefore, grain size has an effect on whether RA will transform at lower strains than expected or on the other hand, transformation does not occur and the overall ductility of the material is not enhanced.

The RA morphology and its surrounding microstructure can also affect its stability. Literature has shown that blocky RA transforms at lower strains than film-like RA and can lead to lower total elongation (TE) and toughness values [29, 35]. Xiong et al. performed experiments comparing RA morphologies and concluded that film-like RA transformed at higher strains than blocky RA [29]. This was attributed to the surrounding microstructure where the film-like RA was surrounded by lath martensite that shielded the austenite during deformation delaying the transformation, while the blocky RA was surrounded by a softer proeutectoid ferrite [29]. They also attributed the stability to a higher hydrostatic pressure that was exerted on the film-like RA. A higher hydrostatic pressure created by the volume expansion during the transformation of RA to martensite exerts residual stress on the film-like RA which suppresses its transformation [29]. These mechanisms give us insight as to how the surrounding microstructure can affect RA stability.

Chiang et al. compared RA stability with tensile strain for equiaxed and lamellar TRIP steel microstructures of the same composition. The heat treatment process for the two microstructures involved IA, followed by cooling to the bainitic region and water quenching to room temperature. The starting microstructure for the equiaxed sample was ferrite and pearlite while the lamellar consisted of a fully martensitic microstructure. Due to differing starting microstructures, the lamellar microstructure involved an austenization step. The equiaxed microstructure had a higher RA volume fraction (14 vs 10 pct.) and larger austenite grains, though a lower RA carbon content than the lamellar microstructure. In the equiaxed microstructure, the RA was surrounded by ferrite while in the lamellar microstructure, the RA was surrounded by bainite. Figure 2.7 shows the percentage of RA transformed as a function of strain for the two microstructures. The lamellar microstructure has a slower rate of transformation than the equiaxed microstructure due to the effects of the surrounding microstructure, RA carbon content, and austenite grain size. The bainite reduces the stress and
strain that would otherwise be transferred to the austenite and allowed for a slower transformation [30].

![Figure 2.7](image)

Figure 2.7 Percent of RA transformed plotted versus true strain for TRIP steels with equiaxed and lamellar microstructures. [30]

### 2.3.3 Effects of RA on Mechanical and Fracture Properties

The strength of QP steels is attributed to the martensite and the ductility is derived from the TRIP behavior that occurs due to RA transformation [36]. The stability of RA is vital to having favorable mechanical properties, where both unstable and too stable RA can lead to no improvements in ductility. Chiang et al. showed that higher RA volume fractions led to a higher tensile strength, due to the transformation to martensite [30]. Chiang et al. also showed that the work hardening behavior can be characterized by the RA stability, where a lamellar microstructure sustains high work hardening rates at larger strains which is significant for formability. UTS x UE (ultimate tensile strength and uniform elongation product) is used as a measure of energy absorption during deformation and increasing RA volume fractions leads to increasing UTS x UE values.

De Moor et al. performed experiments on a C-Mn-Al-Si-P steel to examine how RA volume fractions varied with different QP processes as well as their effect on properties [37]. Results showed that the RA volume fraction increased with partitioning time, which also led to an increase in elongation. The literature also shows that fracture toughness is improved due to the TRIP effect, where crack initiation is delayed and crack propagation rate is reduced due to crack tip blunting [38]. Wu et al. showed that RA stability plays a role in determining toughness.
where martensite transformation at less stable RA at the crack tip can improve the toughness of QP steels by what has been attributed to crack tip shielding and blunting [38]. More stable RA due to a higher RA carbon content can actually reduce toughness because of a lower resistance to microcracking (due to hardness differences amongst phases) leading to faster crack initiation [38].

Since QP steels are relatively new, there have not been many studies examining the fracture mechanisms for this type of steel. Diego-Calderon et al. [39] analyzed the deformation response of individual phases in C-Mn-Si-Al QP steels. Specimens were fully austenitized followed by quenching to 244°C and partitioning temperatures of 300°C or 400°C for 100 or 500 seconds. Nanoindentation tests were performed on the RA in the samples and the data from the tests were plotted as load-displacement curves. The steel partitioned at 400°C for 100 s (QP4-100) had similar volume fractions of RA to the steel that was partitioned at the same temperature for 500 s (QP4-500). QP4-100 did have a smaller average grain size of RA in comparison to QP4-500. The microstructure of these steels consisted of tempered martensite, fresh martensite and RA. The nanoindentation experiments showed that the steel with a smaller average grain size and interlath lamellar morphology of RA (QP4-100) was able to accommodate higher loads before martensitic transformation than the steel with a higher average grain size and blocky RA morphology [39]. Their studies also included in-situ tensile testing to analyze the deformation behavior of each phase in the steel. SEM images showed that hard blocky un-tempered martensite with a great difference in strength caused localization of plastic deformation in the softer tempered martensite matrix [39]. Local rupture occurred at the tempered and fresh martensite boundary.

2.4 Hole Expansion Testing

The increased use of AHSS in the automotive industry has produced interest in edge cracking resistance during forming operations, and hole expansion testing (HET) has been used to characterize material performance. The hole expansion ratio (HER) is calculated as follows,

$$\text{HER} = \frac{D_f - D_o}{D_o} \times 100,$$

(2.1)

where $D_o$ is the initial hole diameter and $D_f$ is the final hole diameter once a through-thickness crack is observed [40]. A typical test configuration with a conical punch is shown in Figure 2.8.
The dies provide the clamping force to hold the specimen and prevent it from moving while the punch travels up through the hole. This type of setup causes the sheet edge to undergo bending and stretching compared to testing with a flat bottom punch where the material only exhibits stretching. Once a through-thickness crack is observed, the test is stopped. Crack detection is usually done by visual inspection or computer-aided imaging inspection software.

![Diagram of hole expansion testing setup](image)

Figure 2.8 Example of a hole expansion testing setup [41, 42]. $D_p$ is the diameter of the punch, $d$ is the hole diameter in the specimen, $F$ is the clamping force the dies are putting on the specimen to prevent the specimen from moving drawing inwards during testing, and $D$ is the final hole diameter after punching. $D_p$ is typically $10 + 0.02/-0.03$ mm.

### 2.4.1 Effects of Edge Condition on HER

The hole preparation method plays a significant role on the HER measured during HET. For example, shearing holes creates deformed edges which affect the local strain path during HET [43]. The shearing process creates a shear-affected zone (SAZ) where the material is hardened and the possibility of voids or cracks that act as stress concentrations. Sheared samples result in lower HERs than other hole preparation methods (i.e. electrical discharge machining (EDM), reaming, waterjet cutting, etc.) due to the presence of the SAZ [44]. The shearing process also creates burrs on the edges and removing this region increases the HER [45], perhaps by reducing potential crack nucleation sites.

The die/punch clearance in producing sheared holes affects the characteristics of the SAZ [46]. A larger die/punch clearance creates a larger burr relative to a smaller die/punch clearance. Most HET is done with the burrs in the up position, which means there are larger strains on the
outer edge, which results in lower HER values [47]. By creating a larger burr with a greater
die/punch clearance, even lower HER values can be expected. Therefore, limiting the burr size
with an appropriate die/punch clearance will help increase the HER.

Generally, sheared holes are the most studied hole preparation method during HET
because the production environment uses efficient and fast methods such as shearing to create
parts. Though, some studies have looked at other hole preparation methods such as drilling, wire
EDM, and machined surfaces. Karelova *et al.* looked into the effects of edge condition (punched,
drilled, and wire EDM) on HER [44]. Optical microscopy showed that the plastic deformation
near the hole was greatest with the punched surface, followed by the drilled and wire cut surface.
The wire cut edge condition only performed slightly better than the drilled edge condition. This
was attributed to the micro-cracks created by preparation methods where the drilled edge
consisted of circumferential cracks (Figure 2.9(a)) that would have the tendency to close during
testing and the wire cut surface consisted of axial cracks (Figure 2.9(b)) that would open up
during HET [44]. Therefore, the damage induced and imperfections created during hole
preparation methods have an influence on HER.

![Figure 2.9 Micrographs of (a) a circumferential crack in the drilled edge and (b) an axial

2.4.2 The Shearing Process

The shearing process is a rapid operation that is ideal for the high production rates in the
automobile industry. For HET, material is held between a die and a pad that applies a pressure to
hold the sheet, while a blade comes down and shears the material [48]. The clearance is defined
as the separation between the die and the shear blade and is calculated as follows:
\[ c = \frac{d_d - d_p}{2t} \times 100 \]  \hspace{1cm} (2.2)

where \(d_d\) is the diameter of the die, \(d_p\) is the diameter of the punch used for punching a hole in test piece which is usually 10 mm, and \(t\) is the thickness of the test piece (all in millimeters) [40].

The shearing process creates distinct regions on the shear face called the rollover, burnish, fracture, and burr regions. Rollover is the area that first comes into contact with the shear blade and is affected by die clearance, sheet tensile strength, and pad variables. Figure 2.10 from work done by Konieczny and Nakata, shows that increasing clearance increases the gap where the material flows and leads to a much larger rollover zone [42, 49]. The next region is known as the burnish and it also is affected by the die clearance and sheet tensile strength (shown in Figure 2.11). With increasing clearance or strength, the size of the burnish region decreases [42, 49]. The next region is known as the fracture region and Figure 2.12 shows that the size of this region is mainly dependent on the sheet tensile strength. The final region is the burr, and experiments have shown that burr height decreases with increasing sheet tensile strength [42, 49].

![Figure 2.10](image.png)

Figure 2.10  Data collected from the hole expansion experiments of Konieczny and Henderson, and Nakata et al. show the percentage of rollover varied with clearance and tensile strength. [42, 49]
The shearing process also creates a SAZ behind the fracture surface where material has work hardened. The depth of the SAZ can be determined from hardness measurements from the shear face into the base material [48]. Studies have shown that removing the SAZ in a HET sample geometry increases the limit strains in materials [50]. Limit strains are the curves that are
typically seen on forming limit diagrams, which signify the maximum strain until necking or fracture occurs in a material.

2.4.3 Punch Geometry Effects on HER

Punch geometry also has an effect on the HER during HET. A conical or flat bottom punch can be used to expand the hole during testing. The conical punch causes the sheet to undergo bending and stretching, while with the flat bottom punch the material only undergoes stretching. Pathak et al. compared the HER of complex phase (CP 590 and CP 800) and DP steels (DP 600 and DP 780) with flat bottom and conical punch testing configurations. Despite different steel grades, their results showed that testing with the conical punch resulted in a higher HER than the flat bottom punch [51]. Other studies have also shown this trend using various materials such as steel and aluminum sheets [52-55] and has been attributed to the bending component introduced by the conical punch that creates a material constraint that delays failure [56].

2.4.4 Mechanical Properties and its Effect on HER

Literature has shown that some mechanical properties correlate well with HER. Increasing tensile strength leads to an inverse correlation with HER up to a strength level (approximately 650 MPa), and thereafter, HER plateaus with increasing tensile strength [55, 57]. The normal anisotropy coefficient (the plastic strain ratio averaged over three directions (typically 0, 45, and 90 degrees) with respect to the rolling direction) and YS/UTS ratio also show a positive correlation with HER [55, 58]. Pathak et al. concluded that the reduction of area (ROA) also showed a positive correlation with HER in DP and complex phase steels with sheared and reamed edge conditions [51]. Jin et al. focused on IA QP, DP and TRIP steels of different thicknesses to better understand the relationship between mechanical properties and HER. It was concluded that materials with a low strain hardening exponents exhibited the least HER differences between different edge conditions because the smaller strain hardening exponents led to a smaller hardness increase at the hole edge due to edge preparation and therefore, creating similar HERs amongst various edge conditions.
2.5 Microstructure and HER in DP Steels

Microstructural characteristics such as martensite carbon content and morphology in DP steels have an effect on stretch-flangeability [59-62], with MVF not being so clear. Many studies have reported that the difference in hardness between ferrite and martensite in a DP microstructure is a dominant factor in stretch-flangeability [59-61], where the martensite carbon content can be used as an indication of the martensite hardness. The greater the difference in hardness between martensite and ferrite, the greater the strain localization on the ferrite by the harder martensite leading to lower HER [59-61].

The effect of MVF on HER is not as clear. Hasegawa et al. reported that increasing MVF resulted in a higher HER with single phase martensitic sheets having the highest HER, though the carbon contents between all the steel grades were slightly different. The MVFs were from a small range of 34 to 49 pct. and then a fully martensitic steel. Other studies have shown the opposite trend where increasing MVF decreases the HER, though the materials had different phase morphologies [60, 63]. Pushkareva et al. also performed a study on a wide range of DP steels with varying MVFs where many of the samples had similar carbon contents, though there did not seem to be any distinct trend with MVF and HER [64]. Therefore, it is reasonable to assume that other characteristics such as martensite morphology and hardness differences between microstructural constituents have a greater effect on HER than MVF.

Martensite morphology has also been shown to have an effect on HER. Kim et al. [62], examined the morphological differences between the DP steels studied by Hasegawa and concluded that the higher MVF steel consisted of a martensite network along ferrite grain boundaries, where the martensite takes on more of the plastic deformation than ferrite postponing failure. This morphology has been shown to prevent crack propagation at the martensite-ferrite interface improving stretch-flangeability [65]. Terrazas et al. characterized the martensite morphology in several DP steels by martensite dispersion throughout the material [60], concluding that HER increases with a more homogenous microstructure as fine and evenly dispersed martensite colonies provide more obstacles for crack propagation [60].

2.6 Microstructure and HER in QP Steels

There is limited published research on QP steels and HET, but there are studies on TRIP steels that can help explain the HER in QP steels [29, 66-68]. Senuma et al. stated that the TRIP
effect promotes micro-crack formation during the hole shearing process [66] but prevents strain localization and suppresses crack propagation during HET and improves HER [66]. Studies have shown that RA that transforms early during deformation is detrimental to HER [67, 68]. Therefore, it is reasonable to assume that ideally one would prefer for RA to transform in the later stages of any testing to increase the chances of delaying failure. It is also recommended that some RA remain after the TRIP effect to enhance ductility. Sugimoto et al. suggested that 2-4 pct. untransformed RA is good for HER in TRIP bainitic steels [67], and that RA that transforms at low strains can lead to strain induced martensite initiating voids that lower HER.

De Moor et al. performed HET with IA and fully annealed QP microstructures and compared HER performance to DP and TRIP steels. The IA QP steel performed better than the DP steel, which was attributed to the hardness differences between the constituents [68]. The TRIP steel performed better than the QP steel, which was attributed to the hardness difference as well as the RA morphology. Comparing lath and blocky RA, lath RA has been shown to increase the HER [29, 67, 68]. This is because lath RA tends to transform at higher strains than blocky RA. Though, a microstructure with all lath RA could be disadvantageous because depending on other factors such as surrounding microstructure and carbon content, the lath RA may never transform and the benefits of the TRIP effect will not be attained. Although, having a mixture of both blocky and lath RA should prove to be beneficial in that some of the blocky RA can transform early during deformation, and the lath RA can transform at later stages during testing. Therefore, it is clear that RA stability plays a vital role in HER in TRIP assisted steels such as QP steels.

### 2.7 Crack Tip Opening Angle

Fracture toughness in sheet steels is a difficult parameter to quantify because a plane strain stress state is not feasible. Elastic-plastic fracture mechanic conditions are present in thin sheets because the crack tip undergoes a significant amount of plasticity as opposed to thicker plates where the plastic zone size is small relative to the crack length and results in linear-elastic fracture mechanic conditions depending on the material yield strength. Therefore, different methods must be implemented to help understand the fracture behavior in sheet steels than in thick sections. One such method used in the aerospace industry to measure the resistance to crack propagation under mode I loading for mostly ductile materials is the CTOA fracture test [69, 70].
The angle between the top and bottom surface of the crack during stable crack growth results in a critical angle for fracture propagation [70] where a larger angle indicates better fracture resistance. CTOA testing measures the resistance to crack propagation, whereas other techniques such as HET and tensile tests involve both crack initiation and propagation.

The setup and procedure for this type of testing is critical to producing valid results [71, 72] in specific materials. The standard for CTOA testing of sheet materials, ASTM E2472, requires stringent sample dimension requirements [71]. Compact tension (CT) specimen geometries can be used for analysis and Figure 2.13 shows an example of a CT specimen with relevant dimensional parameters that will be discussed below. Specimens are fatigue pre-cracked, and there are two dimensions that must be satisfied to achieve low global crack-front constraint conditions which leads to reduced stress triaxiality at the crack tip, allowing the crack resistance to be characterized independent of in-plane dimensions [71]. The first requirement is that the crack length \(a_o\) to thickness \(B\) ratio is greater than or equal to four in order to achieve low global crack-front constraints. The second requirement is the ligament length \(b\) to thickness \(B\) ratio \((b/B, \text{also known as the ligament slenderness})\) is greater than or equal to four, which is implemented to reduce the amount of stress triaxiality near the crack tip. Figure 2.14 shows how the ligament slenderness can affect the stress triaxiality at the crack tip of the specimen. Anti-buckling guides are used to avoid out-of-plane bending of the sheet sample.

![Figure 2.13 Schematic illustration of a CT specimen that is typically used for fracture toughness testing. W is the specimen width, B is the specimen thickness, \(a_o\) is the initial crack length, and b is known as the ligament. [72]](image)
CTOA testing involves the use of a high-resolution camera to follow the crack tip for each successive stable tearing event. Since images are taken at the surface of the specimen, it is possible that crack tunneling occurs, which is where the crack front in the interior of the specimen is ahead of the crack front at the surface. Studies have shown that crack tunneling occurs at the early stages of crack propagation, but as a critical CTOA is observed, the tunneling effect is minimized such that it is negligible [70, 73]. Per ASTM E2472, the critical CTOA is determined between a calculated range of crack extension determined by the sample geometry, which avoids significant crack tunneling effects.

Another interesting factor to consider is that anisotropic properties (specimen orientation) may have an effect on the critical CTOA. Sutton *et al.* [74] focused on the orientation effects on the critical CTOA in an aluminum alloy sheet using results from Dawicke [70]. Per ASTM E1823, the L-T orientation is when the crack propagates perpendicular to the rolling direction and the T-L orientation is when the crack propagates along the rolling direction [75]. The focus of the study was to compare the L-T orientation results from Dawicke to the T-L orientation of the same aluminum alloy. Digital image correlation (DIC) and optical microscopy was performed on middle-crack tension (MT) specimens. MT specimens have a notch located in the center as opposed to an edge crack on CT specimens. Both CT and MT specimens have shown similar critical CTOA results despite geometry differences [69, 70]. Sutton’s analyses showed that the T-L orientation showed more crack tunneling and a critical CTOA of 4.7 degrees compared to 6.0 degrees for the L-T orientation [74]. Therefore, specimen orientation plays a
role in the values for critical CTOA in anisotropic materials and should be considered in experiments.

As previously mentioned, early research on CTOA was performed using aluminum sheets [69, 70]. However, CTOA methods have also been used to study AHSS. Xu et al. [76] performed CTOA experiments on DP 780 and DP 980 CT specimens in various specimen orientations. They used a high resolution camera, and specimen geometry met ISO 22889 requirements, but not the ASTM E2472 specified geometry. High initial CTOA values were seen which was similar to the studies done by Dawicke on aluminum sheets [69]. This indicates that the early stages of crack propagation may exhibit crack tunneling. The DP 980 in the L-T orientation had a critical CTOA of 7.0 degrees while the T-L orientation had a value of 4.7 degrees, suggesting anisotropic properties of this DP steel. Xu’s study concluded that the critical CTOA decreased with increasing YS, MVF and decreasing elongation [76].

CTOA tests show direct evidence of crack propagation. CTOA testing does involve some testing equipment such as a high-resolution camera setup that may not be readily available in some laboratory settings. However, this test method is standardized and a lot of information can be concluded from just one sample and one test which is beneficial for studies with minimal material. The test method is also very straightforward as it only involves fatigue and tension testing methods. There are other methods such as the essential work of fracture (EWF) that have been used as a fracture toughness measurement in sheet steels. Although this method is not standardized and requires many iterations of testing, research with this method is briefly described in the following section to highlight relationships made between crack initiation and propagation.

2.8 The Essential Work of Fracture

Alternative methods to measure the resistance to crack propagation in sheet steels have been explored. Cotterell and Reddel experimented with a method known as the EWF for ductile sheet steels using a double edge notched tensile (DENT) specimen [77]. The EWF consists of two components: 1) the essential work in the fracture zone in front of the crack tip and 2) the non-essential plastic work dissipated in an outer region. The EWF can be divided by the product of the DENT specimen’s ligament length and specimen thickness and plotted against the ligament length to determine the specific EWF from the y-intercept. This testing requires
samples with a DENT geometry and varying ligament lengths. In order to achieve a satisfactory linear regression, many samples are needed to collect data points and this can be time consuming. There is also no testing standard for EWF testing, so there are variables that can skew the results such as sample dimensions, ligament lengths, and test speed. However, Casellas et al. were able to correlate the HER with the specific EWF for various steel grades as shown in Figure 2.15 [78].

Casellas et al. concluded that tougher materials have higher HER and that EWF testing can be used to characterize fracture toughness and understand stretch-flangeability [78]. However, the materials that exhibit the TRIP effect have similar HERs but different EWF measurements. These results suggest that stretch-flangeability and a fracture toughness measurement can be correlated with further analysis on materials that exhibit the TRIP effect.

![HER vs \( w_e \)](image)

Figure 2.15 A plot of the HER versus the specific essential work of fracture. A linear regression can aid in determining the two components of the EWF fracture criteria. [78]
3.1 Overview

This chapter includes a description of the materials tested and the techniques performed to characterize them. The techniques include microscopy (both light optical and scanning electron microscopes), x-ray diffraction (XRD) and electron back scatter diffraction (EBSD). Mechanical properties were determined through tensile tests and an understanding of crack initiation and crack propagation resistance was determined through hole expansion and CTOA testing.

3.2 Materials

The materials for this study include DP 980, DP 1180, IA QP 980, and IA QP 1180 uncoated sheet steels each with a nominal thickness of 1.4 mm (0.055 in.). Table 3.1 shows the chemical compositions of these steel grades. The processing history of the steel sheets is unknown, though it is assumed that the sheets have been cold rolled. The steel sheets (350 x 900 x 1.4 mm) were provided by Baosteel after DP or QP processing. The four steel grades are designated DP 980, DP 1180, QP 980, and QP 1180, signifying the processing and minimum tensile strengths for each steel grade in MPa.

<table>
<thead>
<tr>
<th>Material</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP 980</td>
<td>0.09</td>
<td>0.52</td>
<td>2.07</td>
<td>0.03</td>
</tr>
<tr>
<td>DP 1180</td>
<td>0.13</td>
<td>0.24</td>
<td>2.24</td>
<td>0.04</td>
</tr>
<tr>
<td>QP 980</td>
<td>0.17</td>
<td>1.92</td>
<td>2.06</td>
<td>0.05</td>
</tr>
<tr>
<td>QP 1180</td>
<td>0.17</td>
<td>1.92</td>
<td>2.38</td>
<td>0.05</td>
</tr>
</tbody>
</table>

3.3 Metallography

Baseline microstructural characterization was performed using a light optical microscope (LOM) and a field-emission scanning electron microscope (FESEM). The sheet samples were mounted in Bakelite followed by standard metallographic preparation techniques [79]. Samples were ground through 1200 SiC grinding papers, polished to 0.5 μm with a diamond solution, and
then etched with 2 pct. Nital (mixture of nitric acid and ethanol) for approximately five seconds. Three orientations (planar, longitudinal, and transverse) with respect to the rolling direction from the as-received steel were mounted for metallographic examination. It is important to note that the grinding steps in specimen preparation could induce the TRIP effect in the QP steels; therefore, the microstructures observed by LOM and FESEM techniques may include transformed martensite.

In order to quantify the degree of banding, the methods laid out in ASTM E1268 were used [80]. Light optical images in the longitudinal orientation at 200x magnification were analyzed by placing 10 evenly spaced (30 microns apart) grid lines on the image through ImageJ. The standard requires a minimum total grid line length of 500 mm and the total line length parallel to the banded microstructure was just below 500 mm while the line length perpendicular to the banding satisfied the requirement. The number of particle interceptions and boundary intersections were determined for ferrite in the DP steels parallel and perpendicular to the banded microstructure for five different areas on a given sample. Ferrite was used for measurements because they showed elongated or globular morphologies making them easier to determine as particles. Equation 3.1 shows the anisotropy index (AI), where a non-banded microstructure has a value of one and as banding increases, the AI goes above one [80].

$$ AI = \frac{N_{\perp}}{N_{\parallel}} $$

where $N_{\perp}$ is the summation of the number of interceptions perpendicular to the deformation direction divided by the line length and the number of fields tested. $N_{\parallel}$ is the summation of the number of interceptions parallel to the deformation direction divided by the line length and the number of fields tested.

3.4 Volume Fraction Analysis

The MVF in the DP steels and M-A regions in the QP steels were calculated per ASTM E562 using secondary electron images of the sheet steels in the planar orientation [81]. Mounted samples were polished to a 0.5 μm finish and etched with 2 pct. Nital for five seconds. A total of 10 micrographs at 3000x (DP steels) and 1500x (QP steels) magnification for each steel was used for analysis. The micrographs were acquired using a JEOL 7000F FESEM at 20 kV accelerating voltage and 10 mm working distance. An overlay of three concentric circles with 50
proportionally spaced tick marks was randomly placed on the micrographs through the use of ImageJ software [82] with a total of 1500 counts per steel grade. The martensite carbon content was also determined using Equation 3.2 with the assumption that all bulk carbon is in the martensite:

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

(3.2)

where \(C_{\alpha'}\) is the martensite carbon content in wt. pct., MVF is the martensite volume fraction, and \(C_{\text{bulk}}\) is the carbon content of the steel in wt. pct. [60].

3.5 Retained Austenite Volume Fraction Through XRD

RA volume fractions for IA QP 980 and IA QP 1180 were calculated according to the SAE method [83] through XRD on a Phillips X’pert diffractometer equipped with an X’celerator detector. Nickel filtered copper radiation was used operating at 45 kV and 40 mA with a 1/8° divergence slit, a 1/2° anti-scatter slit on the source side and a 5.5° anti-scatter slit on the detector side. Four ferrite peaks ([110], [200], [211], [220]) and four austenite peaks ([111], [200], [220], [311]) were used in the calculation of RA to account for crystallographic texture that may have been introduced during cold rolling.

Samples were ground through 600 grit SiC grinding papers followed by being submerged in a solution of 2 parts hydrofluoric (HF) acid, 20 parts de-ionized water, and 20 parts hydrogen peroxide. The specimens were submerged for approximately 10 minutes and a minimum of 0.254 mm (0.010 in.) was removed to assure no grinding damage remained and surface RA values reflect bulk values as closely as possible. Samples were then soaked in a detergent (Alconox) to remove oxidation from the chemical removal process. The planar surface (with the x-ray beam parallel to the rolling direction) was measured using a two-theta range of 40 to 105° with a dwell time of 200 s and a step size of 0.05° for a total of 35 minutes per scan. Analysis of the XRD data was performed using the HighScore Plus software. In order to gain a better understanding of the RA transformation rates between the two steel grades, samples that were strained to 5 pct., 10 pct., and failure (uniform elongation region) were analyzed.

The RA carbon content was calculated on 12 samples that were taken throughout the sheet and tested via XRD using data from the {220} peak position of austenite and Equation 3.3:

\[
a_o = 3.555 + 0.044C_\gamma
\]  

(3.3)
where $a_o$ is the austenite lattice parameter in Ångstroms and $C_γ$ is the carbon content in austenite in wt. pct. [84].

3.6 RA Morphology Through EBSD

RA morphology in the IA QP steels was characterized through EBSD. 10 x 10 mm test samples were ground through 240 grit SiC grinding papers to create a wedge along the rolling direction (shown in Figure 3.1). The reason for a wedged sample is that it reduces the amount of time to mill through the surface. The wedged samples were then milled in a cross-section polisher for 8 hours to create a fresh surface that would attempt to minimize the amount of transformed RA. EBSD scans were performed on the longitudinal section of the sample using a FESEM equipped with an EDAX® DigiView Camera and EDAX® OIM-DC 5.2 collection software. 15 x 15 micron scan areas were analyzed at a 70° tilt, with a 14 mm working distance, 0.03 micron step size, and 20 keV accelerating voltage.

![Figure 3.1 Example of the wedge sample used for EBSD scans to characterize the RA morphology. Dimensions are in mm.](image)

3.7 Tensile Testing

Tensile tests were performed according to ASTM E8 procedures to quantify the yield strength (YS), tensile strength (UTS), uniform elongation (UE), total elongation (TE), strain hardening exponent (n) and the instantaneous n-value [85]. The reduction of area (ROA) was also calculated with measurements of the tensile samples taken pre- and post-tensile test. These measurements were done with both a flat-anvil micrometer and a point micrometer. Standard ASTM E8 sheet tensile samples were waterjet cut from the as-received sheet to minimize RA transformation in the IA QP steels, as may occur during mechanical cutting or machining. A total
of 12 samples were tested for each steel grade; four each with the loading axis parallel, perpendicular, and 45 degrees to the rolling direction.

The tensile samples consisted of a 57.2 mm (2.25 inch) minimum gauge section. A 50.8 mm (2 in.) gauge length was measured in the gauge section and strain was measured with a 50.8 mm (2 in.) extensometer within this section. The samples were tested at an engineering strain rate of $2.5 \times 10^{-3}$ s$^{-1}$. This strain rate is considered mid-range for static tension tests [86]. The samples were tested on an MTS Alliance RT/100 electro-mechanical test frame.

The YS was determined using the 0.2 pct. offset method [85]. The slope of the elastic region (Young's Modulus) was used in determining the YS, UE, and TE from the engineering stress-strain curve, as shown in Figure 3.2. The strain-hardening exponent, n, was calculated from a true strain of 0.02 up to the strain at UTS.

![Figure 3.2](image)

**Figure 3.2** An engineering stress-strain curve indicating how YS, UE, and TE were determined using the slope of the elastic region (dashed lines). YS was found using the 0.2 pct. offset method. The UE strain was determined at the UTS. The TE strain was determined at fracture.

### 3.7.1 Plastic Strain Ratio

Standard ASTM E8 tensile samples were waterjet cut to perform tension tests to calculate the plastic strain ratio ($r$). The plastic strain ratios are designated as $r_0$, $r_{45}$, and $r_{90}$ which signifies the tensile axis is parallel, 45 degrees, and perpendicular to the rolling direction during testing. In order to calculate r-values, the material must plastically deform which was determined to be at
4.5 and 10 pct. engineering strain for DP and IA QP steels, respectively, based on tensile results. In order to determine the plastic strain ratios, flat bottom and pointed micrometers were used to measure the strains in the width and thickness directions. Tensile tests were performed on an MTS Alliance RT/100 electromechanical test frame at an engineering strain rate of 2.5 x 10^{-3} s^{-1}. Nine samples were tested for each steel grade, three each with the loading axis parallel, 45 degrees, and perpendicular to the rolling direction.

Typically rolled sheets steels develop a level of planar anisotropy and this can have an effect on the plastic strain ratio. Due to this planar anisotropy, the plastic strain ratio is averaged over three directions with respect to the rolling direction to determine a normal anisotropy coefficient ($r_m$):

$$r_m = \frac{r_0 + 2r_{45} + r_{90}}{4}$$

### 3.8 Hole Expansion Testing

HET was performed according to ISO 16630 [40] and with an experimental setup similar to those typically reported in the literature [41, 42]. Testing was performed at ArcelorMittal and AK Steel research facilities. Test blanks were waterjet cut to 101.6 x 101.6 mm (4 x 4 in., conical punch testing) and 177.8 x 177.8 mm (7 x 7 in., flat bottom punch testing) geometry with a sheared or machined 10 mm hole with a die clearance of 11 pct. of the sheet thickness. Machine edges were created through a reaming process. All samples were tested in the burr-up position at a punch speed of 0.5 mm/s with a camera system monitoring crack formation; HER was determined when a through-thickness crack was detected. HER measurements at both facilities were done while the samples were still in the loaded condition. The nomenclature for the testing conditions is conical punch/machined edge (CM), conical punch/sheared edge (CS), flat punch/machined edge (FM), and flat punch/sheared edge (FS).

HET at ArcelorMittal consisted of testing with a conical punch on sheared hole expansion samples (CS). A powder was brushed on the hole edge to help the camera focus and provide contrast to better detect crack initiation. The HER was determined on an image that had a through-thickness crack that was 0.1 mm in width and was calculated with a program developed by ArcelorMittal.
At AK Steel, two types of edge conditions (sheared and machined) and punch geometries (conical and 25.4 mm flat bottom) were tested. When a through-thickness crack was observed on the camera, the test was stopped and the HER was calculated by taking diameter measurements in ImageJ analysis software [82].

Figure 3.3 is an example of the methodology used to clarify whether crack initiation occurred along the rolling direction during HET. A pair of lines (with an angle of 60° between them) were placed on the image through the center of the test sample in ImageJ. If the crack initiated on or within the two lines, the test sample was considered to have cracked along the rolling direction.

![Figure 3.3 DP 1180 in the CS test configuration with the rolling direction in the y-axis. This is the methodology used to determine if cracking initiated along the rolling direction.](image)

3.9 Crack Tip Opening Angle Testing

CTOA testing was performed according to ASTM E2472 [71] for one sample of each steel grade in the T-L and L-T orientations. Four samples in the L-T orientation were used as practice samples to improve the method and used as comparisons for reproducibility. As previously mentioned, the T-L orientation is when crack propagation occurs along the rolling direction and the L-T is when it propagates perpendicular to the rolling direction. The sample areas where crack propagation would be expected were prepared by lightly grinding through 1200 grit grinding paper and then polishing with a 6 µm diamond solution. This sample prep
helped to view the crack tip as well as the edges of the material where the crack propagated through. Figure 3.4 shows the sample geometry and a recommended anti-buckling plate design. Teflon sheet was placed between the anti-buckling plates and the steel specimen to help reduce the amount of friction during testing.

Figure 3.4 (a) The compact tension specimen geometry used for testing. The dimensions are in millimeters. (b) Anti-buckling plate design proposed in ASTM E2472, which is what was used for the CTOA tests. [71]

The camera system used was a Model K2 DistaMax™ long distance microscope with an Achromat™ 5x objective. This setup has a field of view (FOV) of approximately 0.96 x 0.96 mm. The platform on which the camera rests has a micrometer stage that can move in the x, y, and z directions. Imaging during testing was performed using the program TQSnap. Figure 3.5 shows the CTOA setup with camera and lighting. A second camera was used to continuously record the opposite side of the CT specimen. The purpose of this camera was to provide assurance that the test was constantly running, there were no discrepancies during testing with the sample, and that steady state crack propagation was occurring.

CTOA testing involves fatigue pre-cracking, and subsequently pulling the sample in tension to propagate the crack. Fatigue testing was performed under load control with the maximum fatigue pre-cracking forces shown in Table 3.2 for each steel grade.
The CTOA test setup which includes the test frame, test sample, both camera setups, and lighting fixtures.

The maximum fatigue force was determined by Equation 3.5 to minimize the plastic zone size ahead of the crack tip:

$$P_f = \frac{\xi E B W^{\frac{1}{2}}}{g_1(a_0/W)}$$  \hspace{1cm} (3.5)

where $\xi$ is $1.6 \times 10^{-4}$ m$^{1/2}$, $E$ is Young’s Modulus, $B$ is the sample thickness, $W$ is the sample width, and $g_1(a_0/W)$ is a geometrical factor that takes into account the original crack size and sample width [71].

<table>
<thead>
<tr>
<th>Material</th>
<th>Max Fatigue Force (kN)</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP 980</td>
<td>1.92</td>
</tr>
<tr>
<td>DP 1180</td>
<td>1.98</td>
</tr>
<tr>
<td>QP 980</td>
<td>1.98</td>
</tr>
<tr>
<td>QP 1180</td>
<td>1.84</td>
</tr>
</tbody>
</table>
Fatigue testing was run at an R ratio of 0.1, where the R ratio is defined as the minimum stress divided by the maximum stress. The ASTM standard suggests a minimum of 2 mm fatigue pre-crack extension, though initial testing showed that being relatively close to the minimum 2 mm led to rapid crack propagation upon the tension part of CTOA testing where the crack began to propagate out of the crack viewing region of the anti-buckling plates making it difficult to take images. This erratic crack growth made it difficult to make crack extension measurements with the micrometer stage since the crack was propagating diagonally in the images taken. The rapid crack propagation also made it difficult to keep up with the crack tip and crack extension as the micrometer stage was limited to 1 inch of travel distance. This deviated crack growth also made it difficult to analyze the images since the cracking was so erratic that the crack tip would sometimes deviate from the direction of the rest of the crack and CTOA measurements could not be made. Figure 3.6(a) shows an example of a sample that exhibited the deviated crack growth and buckling. Figure 3.6(b) shows an image of a crack where the crack tip began deviating from the rest of the crack path making it difficult to take measurements.

![Figure 3.6](image)

Figure 3.6 (a) A CTOA sample that exhibited both rapid crack propagation that began to move out of the crack-viewing region as well as buckling. (b) An example of rapid crack propagation where the crack tip began to deviate from the rest of the crack. The arrows are shown as a reference for the direction of crack propagation in comparison to where the crack tip ended up.

Therefore, in order to prevent rapid and erratic crack propagation, a much larger fatigue pre-crack of approximately 8 mm was used to assure that crack growth did not occur outside of the
crack viewing region. The anti-buckling plates (shown in Figure 3.7) consist of hot rolled mild steel, Teflon, and an additional piece of steel that was added to prevent the buckling that was shown during initial tests.

![Figure 3.7](image1)

(a) Anti-buckling plates used for CTOA testing. The white area is the Teflon sheet and the brown rectangular cutout is the extra piece of steel added to prevent the sample from buckling. (b) Image of the CT specimen and anti-buckling plates.

CTOA testing was run in displacement control at a displacement rate of 0.4 mm/min. This displacement rate was chosen because it satisfied the required stress intensification rate of 0.2 - 3 MPa*m^{1/2} s^{-1} within the elastic region for each material. Equation 3.6 shows the minimum and maximum amounts of crack extension where the critical CTOA is determined. The critical CTOA is the steady state angle measured on the surface of the sample at 1 mm behind the crack tip. According to Equation 3.6 and sample geometry, the critical CTOA should be taken between 8 and 37 mm of crack extension for all four materials. For this study, the critical CTOA was taken between 8 and 25 mm for all testing conditions because crack extensions higher than 25 mm put the crack close to the piece of steel that was added to prevent buckling.

\[
\Delta a_{\text{min}} = \frac{50}{(5 + B)}, \quad \Delta a_{\text{max}} = W - a_o - 4B
\]

(3.6)

where \(\Delta a_{\text{min}}\) and \(\Delta a_{\text{max}}\) are the minimum and maximum crack extensions to measure critical CTOA, B is the sample thickness, \(a_o\) is the original crack size, and W is the sample width [71].

In order to track and measure crack extension, the end of the fatigue pre-crack was placed at the right edge of the image frame where the micrometer would be set to zero in the direction
that the crack would traverse. By keeping the crack tip at the right edge of the frame, crack extension during testing could be measured with the micrometer measurements. Figure 3.8 shows an example of an image taken for CTOA analysis. The aperture on the camera was fully open and the lighting was positioned such that the reflection of the sample would help in locating the crack tip. The black lines in Figure 3.8 are drawn as markers for distances from the crack tip of 0.9, 0.93, and 0.96 mm (edge of the frame is the maximum FOV), where the CTOA was measured. The ASTM standard suggests taking measurements 1 mm behind the crack tip, but due to the FOV available, measurements were taken at just under 1 mm. During this interrupted tension test, the test was paused and an image was taken while the force, time, and crack extension were recorded for each individual event with a minimum of 6 events for determination of the critical CTOA. By taking CTOA measurements at various crack extensions, a plot of CTOA versus crack extension is created. From this plot, information such as areas where crack tunneling is occurring (initial high CTOA values) and the critical CTOA (CTOA values at the minimum and maximum amounts of crack extension determined from specimen geometry) can be determined.

Figure 3.8  Shown is QP 1180 in the T-L orientation after a cracking event. Three measurements were taken in ImageJ for each photo to determine the CTOA for a given cracking event.
4.1 Overview

The following chapter describes the results obtained from the methods laid out in the experimental procedure. The results include microstructural characterization, specifically MVF and morphology for the DP steels, RA volume fraction, morphology, and stability for the QP steels, and mechanical tests including tensile tests, HET, and CTOA testing.

4.2 Microstructural Characterization

4.2.1 DP Steel Microstructure

The microstructures of all steel grades examined were imaged via LOM and FESEM. Figure 4.1 shows light optical images of DP 980 and DP 1180 in the longitudinal orientation (normal direction (ND)-rolling direction (RD) plane) showing microstructural banding. Figure 4.2 shows secondary electron images of DP 980 and DP 1180 at the same magnification etched with 2 pct. Nital. The micrographs show the longitudinal orientation with the rolling direction horizontal. Figure 4.2 shows that both microstructures consist of ferrite (dark regions) and martensite (light regions) as expected for a typical DP steel. The DP steel micrographs did not show any apparent evidence of NMCs such as bainite and XRD scans did not reveal RA.

Table 4.1 shows the MVF and the standard deviation, the martensite carbon content, and the anisotropy index to quantify banding for the DP steels in this study. As expected by their designated strength levels, the DP 1180 micrograph has a higher MVF than the DP 980. The MVFs were measured to be 52 ± 6 % and 66 ± 5% for DP 980 and DP 1180, respectively. Results also showed that the DP 1180 had a higher martensite carbon content than the DP 980.

<table>
<thead>
<tr>
<th>Material</th>
<th>MVF (pct.)</th>
<th>Martensite Carbon Content (wt. pct.)</th>
<th>Anisotropy Index (AI)</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP 980</td>
<td>52 ± 6</td>
<td>0.173</td>
<td>2.30</td>
</tr>
<tr>
<td>DP 1180</td>
<td>66 ± 5</td>
<td>0.197</td>
<td>1.64</td>
</tr>
</tbody>
</table>
Figure 4.1  Light optical images of (a) DP 980 and (b) DP 1180 in the longitudinal orientation (ND-RD plane) with the rolling direction horizontal showing microstructural banding. The samples were etched with 2 pct. Nital and images were taken at 100x magnification.

Figure 4.3 shows the MVFs, UTS, and YS values determined for the DP steels in this study and are compared to the correlation Davies constructed between UTS and YS as a function of MVF. The materials in this study fall within the correlation Davies presented. The DP 980 also had a higher AI than the DP 1180 (2.30 versus 1.64), meaning that the degree of banding was higher.

The DP 980 microstructure consists of equiaxed ferrite with relatively coarse martensite situated along ferrite grain boundaries creating a martensite network. There is also evidence of martensite islands located within the ferrite grains. The DP 1180 microstructure also shows ferrite grains that appear to be equiaxed with a higher MVF than the DP 980. The DP 1180 also had martensite islands in the ferrite grains. The DP 1180 microstructure can be more aptly described as a martensite matrix with ferrite islands due to the high MVF.
Figure 4.2 Secondary electron images of (a) DP 980 and (b) DP 1180 in the longitudinal orientation (ND-RD plane) etched with 2 pct. Nital. The rolling direction is horizontal. Images were taken at 1500x magnification.

Figure 4.3 UTS and flow stress as a function of MVF fraction for a Fe-Mn-C alloy with a DP microstructure and carbon contents ranging from 0.06-0.29 wt. pct. tensile tested at a crosshead rate of 0.02 mm/s. The red X’s on the plot are the DP 980 and DP 1180 in this study. [16]

4.2.2 QP Steels Microstructure

Figure 4.4 shows light optical images of QP 980 and QP 1180 in the longitudinal orientation. The QP 980 (Figure 4.4(a)) did not show any microstructural banding, while the
QP 1180 (Figure 4.4(b)) exhibited banding on some areas of the sheet and not on other sections. Secondary electron images of QP 980 (Figure 4.5(a)) show a microstructure consisting of intercritical ferrite (dark areas), martensite-austenite (M-A) regions showing substructure which is presumed to include both blocky and lath RA, and lath-like NMC assumed to be bainite (labeled in Figure 4.5(a)). The QP 1180 (Figure 4.5(b)) shows a smaller amount of intercritical ferrite, but a greater amount of M-A constituent. There is also evidence of NMC though it is less than the amount seen in the QP 980.

Figure 4.4 Light optical images of (a) QP 980 and (b) QP 1180 in the longitudinal orientation (ND-RD plane) with the rolling direction horizontal. The QP 1180 shows microstructural banding though does not seem to be consistent throughout the sheet. Samples were etched with 2 pct. Nital and images were taken at 100x magnification.

RA volume fractions were measured through XRD, and a preliminary study was performed on the QP steels to better understand RA stability with respect to preparation techniques. Figure 4.6 shows XRD patterns for both QP steels that underwent two different preparation techniques (GD and HF). GD signifies that the sample was ground through 600 grit SiC papers, while HF signifies that the samples were ground through 600 grit SiC papers followed by being chemically thinned with an HF solution prior to be scanned. In Figure 4.6, it is evident that the \{220\} and \{311\} austenite peaks are no longer identified or have greatly reduced in intensity in the GD samples with respect to the HF samples for both QP steels, indicating that mechanical deformation caused transformation of RA and subsequently a reduction in the
amount of RA measured. The GD samples resulted in 5.7 and 7.0 pct. RA for the QP 980 and QP 1180, respectively, which proved to be lower than the RA calculated for the HF samples.

Figure 4.5  SEM micrographs of (a) QP 980 and (b) QP 1180 etched with 2 pct. Nital in the longitudinal orientation (ND-RD plane) with the rolling direction horizontal. Images were taken at 1500x magnification.

Figure 4.6  XRD patterns of QP 980 and QP 1180 steels that underwent different preparation techniques, where GD means the sample was ground and HF means the sample was ground and chemically polished.
Table 4.2 shows the RA volume fractions calculated from XRD, M-A, ferrite, and NMC volume fractions with standard deviations calculated through the point counting method, and the calculated RA carbon content using XRD data. Table 4.2 shows that the QP 1180 had a higher M-A volume fraction and lower ferrite and NMC volume fractions than the QP 980. The calculated RA volume fractions from XRD using samples that were prepared by HF resulted in higher volume fractions than the GD samples. The HF sample preparation method proved to be more representative of the bulk RA volume fraction since most of the mechanical deformation was removed by the chemical removal process.

<table>
<thead>
<tr>
<th>Material</th>
<th>RA Volume Fraction (pct.)</th>
<th>M-A Volume Fraction (pct.)</th>
<th>RA Carbon Content (wt. pct.)</th>
<th>Ferrite Volume Fraction (pct.)</th>
<th>NMC (bainite) (pct.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>QP 980</td>
<td>16.0</td>
<td>51.1 ± 4.0</td>
<td>1.14</td>
<td>44.8 ± 3.0</td>
<td>4.1 ± 1.8</td>
</tr>
<tr>
<td>QP 1180</td>
<td>12.6</td>
<td>73.3 ± 5.9</td>
<td>1.15</td>
<td>24.9 ± 5.4</td>
<td>1.8 ± 0.9</td>
</tr>
</tbody>
</table>

Since the materials were presumed to be cold-rolled, it is possible that the materials have crystallographic texture which may result in a variation in RA throughout the QP steel sheets. Therefore, a study was performed to better understand RA variation and is shown in Figure 4.7, where RA measurements were taken along the length of the provided sheets. At first glance, Figure 4.7 shows that there is variation in RA throughout the sheet, with the QP 980 ranging from 13-17 pct. and the QP 1180 ranging from 11-15 pct. RA. However, it is not clear if the RA volume fraction variation shown in Figure 4.7 is actual variation throughout the sheet or error associated with the method used to calculate the volume fraction. The SAE method for calculating RA volume fraction through XRD includes many variables such as sample preparation, number of peaks analyzed, radiation source, and sample tilting and rotation which can all lead to different RA volume fractions and a high uncertainty. Therefore, it is possible that testing parameters may stray away from the true bulk RA volume fractions. However, for this study all samples were tested on the same machine with the same testing configurations, therefore standardizing the results. The average RA volume fraction of each steel shown in Figure 4.7 was used as the bulk volume fraction shown in Table 4.2.
RA volume fractions measured for (a) QP 980 and (b) QP 1180 to better understand RA variation throughout the sheet.

RA stability in the QP steels was another important factor considered. RA stability was analyzed by measuring the RA volume fraction on QP samples that were strained to 5 pct., 10 pct., and after failure (measurements were taken from the uniform deformation region of the sample). Figure 4.8 shows the RA volume fraction as a function of strain for the two QP steel grades. The RA volume fraction of the strained samples were an average of 3 samples. The results suggest that the RA in the QP 1180 transforms at larger strains than the QP 980 as presented in Figure 4.8. Figure 4.8 also shows that the QP 1180 had a higher amount of RA after fracture. The differences in RA stability between the QP steels called for an understanding of the RA morphology.

Figure 4.9 shows EBSD scans of QP 980 and QP 1180 with a total confidence index (CI) ≥ 0.5. The scans were then filtered to only show areas ≥ 0.1 CI. The measured RA volume fractions for the scanned areas shown in Figure 4.9 are 2 pct. and 4 pct. for QP 980 and QP 1180, respectively. The RA volume fractions are lower than those calculated by XRD because lath RA is difficult to index in EBSD due to resolution limitations and EBSD only takes into account the area scanned. Therefore, the RA volume fraction measured through EBSD is only that of the blocky RA indexed as well a small sample area (15 x 15 microns) and is not a measure of bulk RA volume fraction.
Figure 4.8 Plot of RA volume fraction in QP 980 and QP 1180 as a function of strain, showing the transformation rate of RA in each material. The error bars are taken from the standard deviation of the RA variation throughout each sheet.

Figure 4.9 shows that both QP steels show evidence of blocky RA. In order to get a sufficient scan (C.I. > 0.5) through EBSD, the sample must be in a finely polished condition which is difficult to achieve with the chemical thinning process used for the XRD samples. Therefore, the samples were prepared in a cross-section polisher which provided a fresh surface through milling of the material. The amount of deformation the milling process causes has not been measured, though it is assumed that some amount of deformation has occurred, but is much less than typical polishing techniques. The darker areas in Figure 4.9 can be either an indication of transformed RA that occurred during sample preparation or fresh martensite which is highly substructured and does not produce a good pattern to be indexed. Appendix B has more EBSD scans that were performed.
Figure 4.9  Superimposed image quality and phase maps of (a) QP 980 and (b) QP 1180 obtained through EBSD. The red represents BCC-Fe (ferrite or martensite) and the green represents FCC-Fe (austenite). The darker areas can be transformed RA or fresh martensite.

4.3 Mechanical Testing

4.3.1 Tensile Testing

Figure 4.10(a) shows the engineering stress vs. engineering strain tensile curves from four samples of each of the four steel grades with the loading axis parallel to the rolling direction and Figure 4.10(b) shows tensile curves for each steel grade with the loading axis parallel, perpendicular and 45 degrees to the rolling direction. The curves in Figure 4.10(b) are an average of the four samples tested for each direction. All steel grades met their minimum tensile strength designation and as the UTS increases, the elongation decreases for each steel type. All steel grades show continuous yielding and the QP steels show greater elongation relative to the DP steels which may be attributed to the TRIP effect. Table 4.3 presents average tensile properties and the standard deviations determined for each steel grade loaded along the rolling direction. The elongation of the DP 1180 is somewhat lower than the DP 980; the n value, and YS:UTS ratios are similar. The ROA in the DP 980 is slightly higher than the DP 1180. The QP 1180 has a higher YS, UTS, and YS/UTS ratio, but lower elongation, ROA, and n-value relative to the QP 980.
Figure 4.10  (a) Engineering stress vs. engineering strain curves of tensile samples oriented along the rolling direction for each steel grade. Samples were tested at an engineering strain rate of $2.5 \times 10^{-3}$ s$^{-1}$. Strain was measured on a 50.8 mm (2 in.) gage length using a 50.8 mm (2 in.) extensometer. (b) Engineering stress vs. engineering strain tensile curves of each steel grade parallel, perpendicular and 45 degrees to the rolling direction.

Table 4.3 – Mechanical Properties of the Experimental Steels

<table>
<thead>
<tr>
<th>Material</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>UE (pct.)</th>
<th>TE (pct.)</th>
<th>YS/UTS</th>
<th>n</th>
<th>ROA (pct.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP 980</td>
<td>716 ± 33</td>
<td>1043 ± 15</td>
<td>7.5 ± 0.4</td>
<td>13.8 ± 0.3</td>
<td>0.69 ± 0.02</td>
<td>0.11</td>
<td>36.5 ± 0.8</td>
</tr>
<tr>
<td>DP 1180</td>
<td>827 ± 5</td>
<td>1256 ± 3</td>
<td>5.9 ± 0.2</td>
<td>9.9 ± 0.7</td>
<td>0.66 ± 0.01</td>
<td>0.10</td>
<td>32.5 ± 2.3</td>
</tr>
<tr>
<td>QP 980</td>
<td>698 ± 15</td>
<td>1057 ± 5</td>
<td>15.7 ± 0.3</td>
<td>20.7 ± 0.2</td>
<td>0.66 ± 0.02</td>
<td>0.20</td>
<td>41.9 ± 1.9</td>
</tr>
<tr>
<td>QP 1180</td>
<td>990 ± 27</td>
<td>1188 ± 8</td>
<td>11.7 ± 0.2</td>
<td>16.2 ± 0.4</td>
<td>0.83 ± 0.02</td>
<td>0.11</td>
<td>36.0 ± 1.1</td>
</tr>
</tbody>
</table>

Figure 4.11 shows the instantaneous n-values versus true strain for the experimental steels. The DP steels both show a rapid decrease in n as the true strain increases, suggesting very high initial strain hardening. At low strains, the QP steels show a decrease in n as the true strain increases. However Figure 4.11 shows that at higher strains (greater than 0.015) there is a gradual increase in n as the strain increases which is associated with the TRIP effect. Figure 4.11 also suggests that the QP 980 shows a higher amount of RA transformation at lower strains (also seen in Figure 4.8) as it has a higher increase in n as true strain increases relative to the QP 1180. Figure 4.11 also suggests that approximately half of the RA transformation in the QP 980 occurs
at true strains below approximately 0.04 (comparing with Figure 4.8) and then slowly declines with increasing true strain. On the other hand, the instantaneous n value of the QP 1180 continues to gradually increase to a true strain of approximately 0.10 indicating that the TRIP effect is still occurring up to that point. The noise seen in the QP steels data is artifacts of the machine specifically slip occurring in the grips used and measurements representative of the TRIP effect.

Figure 4.11 Instantaneous n as a function of true strain for all experimental steels. Average slopes were taken from a true strain of 0.015 to failure excluding negative values. The noise seen in the QP data is due to slip in the grips.

Figure 4.12(a)-(c) shows the plastic strain ratios for all three directions. Figure 4.12(d) shows the normal anisotropy coefficient for the four steel grades examined in this study. The DP 1180 and DP 980 in this study have similar normal anisotropy coefficients, however, the plastic strain ratios for each orientation are different. The DP 980 has a higher plastic strain ratio in the \( R_0 \) and \( R_{90} \) orientations while the DP 1180 has a higher plastic strain ratio in the \( R_{45} \) orientation. The QP 980 shows a higher normal anisotropy coefficient than the QP 1180 as well as higher plastic strain ratios in the three directions.
Figure 4.12 Plastic strain ratios (a) parallel, (b) 45 degrees, and (c) perpendicular to the rolling direction for each steel grade. (d) The normal anisotropy coefficient, $R_m$, determined for each steel grade. The DP values were calculated at an engineering strain of 4.5 pct. and the QP steels were calculated at 10 pct. The error bars are from the standard deviation.

4.3.2 Hole Expansion Ratio Testing

HERs were calculated for all steel grades, and Figure 4.13 shows the HET results for samples tested with a conical punch and a sheared hole at ArcelorMittal. The samples were examined under a stereographic microscope at 50x magnification prior to HET to make sure there were not any signs of severe cracking that occurred during the shearing process that could have skewed the results. It is expected that micro-cracking would occur from the shearing
process, though this was the methodology used to determine whether severe cracking occurred on the surface. HET showed that DP 980 had a higher HER than the DP 1180. This trend was expected because higher tensile strengths typically come at the expense of ductility. The QP steels show very interesting hole expansion results, as the QP 980 and QP 1180 have similar HERs. These results do not follow the general trend that higher strength steels typically have a lower HER.

![HER measurements for all steel grades.](image)

Figure 4.13 HER measurements for all steel grades. Measurements are from an average of 3 tests with a sheared edge using a conical punch. Errors bars were calculated from the standard deviation of the 3 tests.

Before comparing HER results using different punch geometries, Sriram et al. compared HER results using a conical punch (what was considered a hole extrusion test) and flat bottom punch (what was considered a hole expansion test) with sheared edge conditions [55]. They concluded that there is a correlation between the two test configurations and that either test method could be implemented [55]. The results also showed that the hole extrusion test resulted in higher HERs than the hole expansion test. Figure 4.14 shows their data as well as the data in the study for comparison.

Figure 4.15 shows the HER results of the DP steels as a function of edge condition and punch geometry. The overall trend seen in Figure 4.15 is that the DP 980 has a higher HER than the DP 1180 independent of edge condition or punch geometry. The ratios of HER shown in Figure 4.15 were calculated between the two materials for a given test condition. Between the
two punch geometry conditions, the results show that samples tested with the conical punch have a higher HER than the flat bottom punch for the same edge conditions. With respect to edge condition, the CM samples showed a higher HER than the CS samples for a given material. However, the HER of the FM samples were nearly identical to the FS samples despite having different edge conditions.

Figure 4.14 A comparison between hole extrusion limits and hole expansion limits. The correlation proves that either test method can be implemented. [55]

Figure 4.15 HERs for DP 980 and DP 1180 as a function of punch geometry and edge conditions performed at AK Steel Research and Innovation Center. Error bars were calculated from the standard deviation.
Figure 4.16 shows the HER for the QP steels as a function of edge condition and punch geometry. The QP980 and QP 1180 have similar HERs (within error) for all testing conditions with the exception of the FM condition. The effect of punch geometry does not seem to be evident in the QP steels with the exception of the QP 1180 CM and FM conditions. When considering edge condition, however, the machined QP samples show a higher HER than the sheared samples, similar to the DP steels tested here, meaning the QP steels are also sensitive to edge conditions.

Since testing occurred at two different facilities and the methods for calculating HER were done differently, the DP 980 results from both facilities were compared for samples with sheared holes using a conical punch. Figure 4.17 shows that the results were consistent for 3 different testing machines and two different analysis methods. The AK Steel Hille Wallace press (AK Hill Wall) is a machine where hole expansion testing can be performed. Testing on this machine involves a technician watching for a through-thickness crack and stopping the test once a crack occurs. Three measurements are then made with calipers and an average diameter is used to calculate the HER. It was determined that the crack initiated along the rolling direction for all samples tested at both facilities. Therefore, since the results were consistent between the two facilities it is valid to make comparisons between the two sets of results.

![HERs for QP 980 and QP 1180 as a function of punch geometry and edge conditions performed at AK Steel Research and Innovation Center. Error bars shown are from the standard deviation.](image)
4.3.3 CTOA Testing

CTOA testing was performed on all steel grades. All samples were validated to have the stress intensification rates suggested by the standard. Figure 4.18 shows plots of CTOA versus crack extension for DP steels in two different orientations. In both orientations, the DP 980 had a higher critical CTOA than the DP 1180. The L-T orientation also had a higher critical CTOA for both steel grades in comparison to the T-L orientation. Specimen orientation did have an effect on the DP steels with the DP 1180 being less sensitive than the DP 980.

Figure 4.19 plots CTOA vs crack extension for QP steels in two different orientations. QP 980 has a higher critical CTOA for both orientations than the QP 1180. Samples in the L-T orientation had a higher critical CTOA than the T-L orientation. Most samples showed high CTOAs at early crack extensions with the exception of the QP 1180. These high CTOAs are attributed to crack tunneling and crack blunting. The QP steels were also sensitive to specimen orientation with the QP 1180 being less sensitive.

Table 4.4 shows the critical CTOA value for each steel grade in each orientation with its standard deviation. The standard deviation comes from the CTOA values measured between the minimum and maximum amounts of crack extension where the critical CTOA is determined.
Figure 4.20 shows the reproducibility with one user performing the measurements with two different specimens of the same steel grade. Figure 4.20 shows that for the DP 1180 in the L-T orientation, the critical CTOAs fall within error of each other proving the reproducibility of this test method.

Figure 4.18 Plot of CTOA vs crack extension for (a) and (b) DP 980 and DP 1180 in the L-T orientation and (c) and (d) of DP 980- and DP 1180 in the T-L orientation.
Figure 4.19  Plot of CTOA vs crack extension for (a) and (b) QP 980 and QP 1180 in the L-T orientation and (c) and (d) of QP 980- and QP 1180 in the T-L orientation.

Table 4.4 – Critical CTOA values

<table>
<thead>
<tr>
<th>Material</th>
<th>L-T (°)</th>
<th>T-L (°)</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP 980</td>
<td>6.59 ± 0.80</td>
<td>4.94 ± 0.55</td>
</tr>
<tr>
<td>DP 1180</td>
<td>4.98 ± 0.67</td>
<td>4.62 ± 0.76</td>
</tr>
<tr>
<td>QP 980</td>
<td>7.92 ± 0.46</td>
<td>6.50 ± 0.50</td>
</tr>
<tr>
<td>QP 1180</td>
<td>6.62 ± 0.65</td>
<td>6.10 ± 0.66</td>
</tr>
</tbody>
</table>
Figure 4.20  Plot of CTOA versus crack extension for two different DP 1180 samples in the L-T orientation. These results show the reproducibility and consistency of the test method with one operator and two different samples of the same steel grade in the same orientations.
5.1 Overview

The results in the previous section will be further analyzed to better correlate HER and mechanical properties, microstructure, and testing conditions for AHSS. The microstructural analysis focuses on martensite and RA effects on HER for the DP and QP steels, respectively. CTOA testing is also compared with HER to clarify if a fracture toughness measurement can help understand edge fracture during HET.

5.2 DP and QP Steel Processing

The processing routes of the four steel grades in this study are unknown. Based upon microstructural analysis, an attempt has been made to determine the likely processing routes of each steel grade.

5.2.1 DP Steels

The DP steels are both presumed to have had a ferrite-pearlite starting microstructure and were cold rolled to final thickness prior to heat treatment. The steels were then intercritically annealed followed by direct quenching. There were no apparent signs of NMCs such as bainite or RA in the DP steels. The $A_1$ and $A_3$ temperatures were calculated according to Equations 5.1 and 5.2 based on the chemical composition and resulted in temperatures of 725 and 881°C for the DP 980 and 716 and 849°C for the DP 1180 [87]. Thermo-Calc analysis using the chemical compositions and measured MVFs of each DP steel showed that the IA temperatures were approximately 773°C and 767°C for DP 980 and DP 1180, respectively.

$$A_1 = 723 - 10.7Mn - 16.9Ni + 29.1Si + 16.9Cr + 290as + 6.38W$$ (5.1)

$$A_3 = 910 - 203\sqrt{C} - 15.2Ni + 44.7Si + 104V + 31.5Mo + 13.1W$$ (5.2)

5.2.2 QP Steels

The QP steels are also presumed to be cold rolled prior to heat treatment due to the fact that the QP steels have RA and cold rolling post heat treatment would likely reduce RA volume fraction. The QP steels show microstructures with similar features (featureless intercritical
ferrite, lath NMCs, and M-A regions with substructure) to those previously observed for a TRIP steel that underwent QP heat treatment [88]. The QP steels have gone through an IA treatment because intercritical ferrite is shown in both microstructures. The $A_1$ and $A_3$ temperatures were calculated using Equations 5.1 and 5.2 based on chemical compositions and resulted in temperatures of 757 and 913°C for QP 980 and 753 and 913°C for QP 1180. Thermo-Calc analysis using the chemical composition and the ferrite volume fraction resulted in approximate IA temperatures of 802°C and 823°C for QP 980 and QP 1180, respectively. The QP steels were then quenched to a temperature between the $M_s$ and $M_f$ with calculated $M_s$ temperatures of 391°C and 382°C for QP 980 and QP 1180, respectively, using Equation 5.3 [89]. It is assumed that the QP 1180 was quenched to a lower temperature to increase the amount of $M_{QT}$ due to the fact that the QP 1180 showed a higher M-A volume fraction (73 vs. 51 pct.). The QP steels show evidence of NMCs which possibly formed during the partitioning stage [88]. It is also assumed that both microstructures have fresh and carbon depleted martensite that was formed after the partitioning stage.

$$M_s = 539 - 423C - 30.4Mn - 7.5Si + 30Al$$ (5.3)

5.3 Mechanical Properties and HER

Analysis of mechanical properties and their effect on HER has been performed. The data used for HER in this section is from the testing performed at ArcelorMittal.

5.3.1 DP Steels

Mechanical properties, including UTS, $R_m$, YS/UTS ratio, ROA, and phase hardness differences, have all been suggested to correlate with HER [55, 58-61, 90]. Therefore, the focus of this section will be on those mechanical properties and their effect on HER. Sadagopan et al. showed that HER decreases as the UTS increases up to approximately 650 MPa, but at higher UTS values the HER remains constant [55]. Figure 5.1 shows the DP steels in this study along with data from Sadagopan et al. for HER vs UTS [55]. The DP steels in our study follow the trend that as the strength increases, the HER decreases, however, the DP steels fall in the range where Sadagopan et al. predicted that HER tends to remain constant [55]. The other DP steels shown in Figure 5.1 also do not seem to follow the trend line, similar to the steels in this study. Therefore, it can be assumed that DP steels do not correlate well with the trend line determined
by Sadagopan et al. especially at higher strengths, suggesting microstructure plays a significant role in HER performance.

Sadagopan et al. [55] also compared HER with the normal anisotropy coefficient and noticed that increasing HER correlated with higher normal anisotropy coefficients. The DP steels in this study had very similar normal anisotropy coefficients, and fall in the same range as the DP steels in previous studies shown in Figure 5.2. The DP 980 and DP 1180 sheets examined in this study have similar normal anisotropy coefficients, but have significantly different HER performance.

![Figure 5.1 HERs for various steel grades tested with a conical punch and sheared edge condition plotted against UTS. Error bars are from the standard deviation. The DP and QP steels are plotted from this study: [55]](image)

Studies have shown a positive correlation between HER and YS/UTS ratio [58] in DP sheets tested with a conical punch and a sheared edge condition. Figure 5.3 shows a collection of various steel grades, including the DP steels in this study, plotted as HER versus YS/UTS ratio. The DP steels in this study show a positive correlation not inconsistent with Figure 5.3. Figure 5.4 shows HER versus ROA for the steels in this study. For the DP steels, a higher HER comes with a higher ROA. This trend is not clear with the QP steels since they have very similar HERs, with the exception of the FM condition.
Figure 5.2  HERs for various steel grades tested with a conical punch and sheared edge condition plotted against the normal anisotropy coefficient. DP 980:+, DP 1180:▼, QP 980:●, and QP 1180:□. Error bars are from the standard deviation. [55]

Figure 5.3  HERs for a variety of steel grades tested with a conical punch and a sheared edge condition plotted as a function of YS/UTS ratio. There is a positive correlation seen with HER and YS/UTS ratio. The steels in this study are labeled on the plot. [58]
Pathak et al. showed that ROA had a positive correlation with HER for DP steels with both sheared and reamed holes tested with a conical punch. This trend also holds true for the DP steels in this study. Finally, the difference in hardness between martensite and ferrite has been shown to be an important factor in stretch-flangeability [51, 60, 61]. For this study, the martensite carbon content was used as an indication of the hardness where a higher martensite carbon content means a higher martensite hardness. Figure 5.5 plots HER as a function of martensite carbon content for a variety of DP steels including those in this study. All testing configurations are plotted for the DP steels in this study, while the rest of the data is only testing performed with a conical punch and a sheared edge. It is clear that the steels in this study show that the lower martensite carbon content correlates with a higher HER despite testing configurations (DP 980 always has a higher HER than DP 1180). Figure 5.5 shows that there is a decreasing trend of HER as martensite carbon content increases.
Figure 5.5    HERs for a variety of DP steel grades plotted as a function of martensite carbon content. All testing configurations for the DP steels in this study are plotted, while the rest of the data comes from testing performed with a conical punch and a sheared edge condition. Error bars are from the standard deviation. [59, 60]

5.3.2 QP Steels

The QP 980 and QP 1180 sheets tested here had similar HERs for all testing conditions except for the FM condition. Jin et al. showed that YS/UTS ratio has a positive correlation with HER for a variety of QP, DP, and TRIP steels (shown in Figure 5.3) [58]. The QP 1180 in this study does have a higher YS/UTS ratio which may factor into the comparable HER seen with the QP 980. The QP 980 has a higher normal anisotropy coefficient and ROA and still has a HER similar to QP 1180. Therefore, trends previously suggested in the literature of HER with mechanical properties do not seem to apply for these QP steels [55, 58]. The hardness differences between phases/constituents was not measured for the QP steels and therefore, cannot be ruled out as a possible characteristic that may correlate with HER. However, other factors such as a microstructural aspect like RA stability, could be playing a larger role in the hole expansion performance of the QP steels in this study. RA stability will be discussed in section 5.5.2.
5.4 Effects of Punch Geometry and Edge Condition on HER

The effects of two punch geometries (conical and flat bottom) and two edge conditions (sheared and machined) were analyzed for their effect on HER. The data for the different testing configurations of HET is from testing performed at AK Steel's Research and Innovation Center.

5.4.1 DP Steels

The DP 980 had a higher HER than the DP 1180 independent of punch geometry and edge condition, which was expected based on typical results reported in the literature that indicate HER decreases with increasing strength or decreasing ductility [55]. Although the HER trends are not surprising, there are some interesting points that came out of the different testing conditions.

For the DP steels, the conical punch resulted in a higher HER than the flat bottom punch for a given edge condition which is in agreement with the literature [51]. However, with respect to varying edge conditions, the DP steels did not seem to follow all of the trends previously reported in the literature where a machined edge performs better than a sheared edge [51, 61]. The CS and CM testing configurations show that machined edges outperform the sheared edge condition which is in agreement with the literature [51, 61]. However, the FM and FS testing conditions, show similar HERs (within error of one another) despite the different edge conditions and contradicts the literature. The FM condition does not induce effects on the material that are seen in the other configurations such as there is no strain hardening effects at the sample edge due to a machined edge condition, and the flat bottom punch does not create the strain gradient that is induced on the sample by the conical punch (bending component).

Pathak et al. compared lower strength DP steels (DP 600 and DP 780) in a FM and FS configuration and showed similar results [51]. They concluded that the SAZ did not have a great influence on the HER of samples tested with the flat bottom punch because they noticed that cracking initiated away from the edge of the sample. In this study, there was no evidence of crack initiation away from the edge. However, due to resolution and frame rate capabilities, it is possible that the exact location of crack initiation was not seen. From the test data, it can be concluded that the DP steels in this study that were tested with a flat bottom punch are not sensitive to edge effects.
Another point of interest is that the relative ratios of HER for DP 980 and DP 1180 shown in Figure 4.15 are independent of test conditions (i.e., punch geometry and edge condition) with the exception of the FM samples. This suggests that HET with the CS, CM, and FS test conditions result in similar relative HER values for DP steels with similar mechanical properties (normal anisotropy coefficient and YS/UTS ratio).

5.4.2 QP Steels

The QP steel HERs are interesting in that they performed within error of one another with the exception of the FM condition. The machined holes showed a higher HER than the sheared holes due to there not being a SAZ which is in agreement with the literature [51, 61]. The sensitivity to edge conditions during HET is well established, though the effects of punch geometry do not seem be as clear for these materials. Typically, a conical punch produces a higher HER than a flat bottom punch due to the bending component introduced by the conical punch. However, the QP steels tested in the CS and FS conditions have similar HERs within experimental error, which leads to the interpretation that the transformation of RA at the sheared edge plays a much larger role than the effects of punch geometry for the sheared edge conditions.

The QP 980 also shows no sensitivity to punch geometry for all testing conditions, while the QP 1180 only shows punch geometry sensitivity for the CM and FM conditions. This warrants further investigation to better understand punch geometry effects and the amount of RA transformed at the edge due to shearing in QP steels or more specifically to steels that exhibit the TRIP effect.

The relative ratios of HER shown in Figure 4.16 once again are independent of test conditions with the exception of the FM samples. Figure 5.6 shows the ratios of HER calculated for the 980 and 1180 steels. As the ratios of HER comparing DP 980/1180 (Figure 4.15) and QP 980/1180 (Figure 4.16) show that HER is independent of testing conditions with the exception of the FM samples, Figure 5.6 shows that the FM condition is also different when comparing two different steel grades with the rest of the testing conditions being similar. There is limited research for HET with a flat bottom punch and varying edge conditions and these results suggest that further investigation is needed.
5.5 Microstructure and HER

Microstructural analysis was performed to understand its effects on HER. Specifically, the DP steels focused on the martensite morphology and volume fraction while the QP steels focused on the RA.

5.5.1 DP Steels and HER

For the DP steels, the microstructural focus was on the MVF and morphology. Literature has shown opposing trends for MVF and HER with MVFs up to 50 pct. [57, 59-61]. Figure 5.7 shows HERs for various DP steels tested with a conical punch and sheared edge condition plotted against the MVF. Looking at the data from Hasegawa et al., one would assume that the HER increases with increasing MVF with 100 pct. MVF having the highest HER. However, when we only focus on the DP steels in this study and those studied by Terrazas et al., the HER decreases with increasing MVF. The DP steels studied by Tsipouridis et al., show that increasing MVF has no effect on the HER. Figure 5.7 shows that there is a gap of missing data between 70-100 pct. MVF that may help in clarifying the trend between HER and MVF if one exists.
It is also believed that the martensite morphology plays a role in the HER. Based on the results of this study and those from other studies [57, 59-61], it is possible that there is a peak in HER as DP steels reach 50 pct. MVF after which higher MVFs no longer enhance HER. It is believed that at higher MVFs, HER is controlled by other microstructural characteristics such as the martensite morphology since both the materials in this study and of Terrazas et al. showed that MVFs over 50 pct. led to a decrease in HER.

An interconnected martensite morphology along ferrite grain boundaries and fine dispersion of martensite has been shown to have an effect on HER [60]. The DP 980 microstructure showed equiaxed ferrite with a martensite network (and a lower martensite carbon content) along ferrite grain boundaries, while the DP 1180 microstructure showed what is better described as a martensite matrix (and a higher martensite carbon content) with ferrite islands. A martensite network along ferrite grain boundaries has been shown to prevent crack propagation at the martensite-ferrite interface as well as take on more stress than the ferrite reducing strain localization which aids in delaying failure [59], which may be one of the reasons for the much higher HER seen in the DP 980 with respect to the DP 1180. Therefore, a martensite network along ferrite grain boundaries (with a carbon content low enough to attain minimum designated strength levels) seems ideal for DP microstructural design to prevent crack propagation as this
would reduce the hardness differences between ferrite and martensite and this morphology effectively “shields” ferrite during deformation. However, retaining this morphology may be difficult as the MVF increases.

Martensite morphology is dependent upon the MVF in that as MVF continues to increase, it becomes difficult to retain martensite close to ferrite grain boundaries as opposed to becoming a majority of the microstructure such as the DP 1180 in this study. Higher strength steels may then need to focus on alternative routes such as solid solution strengthening of the ferrite, finding the right combination of MVF and carbon content to increase strength levels, or look through tempering martensite to reduce the strength difference between ferrite and martensite. Therefore, MVF and morphology appear to work synergistically to an undetermined MVF, where thereafter martensite morphology seems to then have a greater influence.

5.5.2 QP Steels and HER

The microstructural focus of the QP steels is the RA stability. XRD analysis of the two QP steels showed that the RA stability was apparently different during tensile deformation with the QP 980 RA transforming larger volume fractions of austenite at lower strains (Figure 4.8). During HET, the better HER in respect to RA stability would be one that has a steady, continuous transformation of RA to martensite throughout the test as the TRIP effect has been shown to prevent strain localization and suppress crack propagation [66]. Large RA volume fraction transformations occurring at low strains have been shown to decrease the HER [67]. Therefore, RA stability plays a vital role in stretch-flangeability and this is also shown in the HER data collected. RA stability is dependent on many factors such as austenite chemical composition, crystallographic texture, RA grain size, morphology, and surrounding microstructure [27-30]. This qualitative analysis has been made focusing on the RA carbon content, morphology, and surrounding microstructure to characterize the RA stability between the two QP steels.

The RA carbon contents were calculated as 1.14 and 1.15 wt. pct. for the QP 980 and QP 1180, respectively. Carbon enriched austenite leads to greater stability hence being less prone to transformation upon deformation. The RA carbon contents are very similar and therefore can be ruled out as a reason for contrasting stabilities. The QP process aims to stabilize RA, but chemical composition can also play a role. The QP 1180 had a higher bulk Mn content than the
QP 980 which reduces the $M_s$ temperature (approximately 391°C for QP 980 and 381°C for QP 1180) and increasing the range for RA stability.

The RA morphology was analyzed through EBSD. EBSD techniques were useful in showing that both QP steels contained blocky RA, but was not able to resolve lath RA. The scans also showed the possibility that some of the blocky RA could have transformed during sample preparation. It is presumed that both lath and blocky RA reside in the M-A regions. The M-A volume fractions were 51 and 73 pct. for the QP 980 and QP 1180, respectively. Therefore, since the QP 1180 showed a greater amount of M-A region, it is possible that the probability of having a higher amount of lath RA is greater than in the QP 980. It is also possible that the carbon contents between lath and blocky RA are different as studies have shown that lath RA tends to have a higher carbon content than blocky RA [92, 93], though they were not analyzed for this project. The volume fraction calculated by EBSD also only takes into account the area scanned and is therefore not representative of the bulk material.

The final area analyzed was the surrounding microstructure. This was done by analyzing both SEM micrographs and EBSD scans. As previously mentioned, the QP 1180 micrograph shows a higher fraction of M-A region with a smaller amount of intercritical ferrite than the QP 980. With this in mind, it is assumed that a majority of the RA in the QP 1180 is surrounded by martensite in the M-A regions while with the QP 980, the RA has a greater chance of being surrounded by ferrite. Although the EBSD scans shown in Chapter 4 are of certain size and scan area, they show that a majority of the blocky RA in the QP 1180 is surrounded by martensite where in the QP 980 most of the RA is surrounded by ferrite. Therefore, the RA surrounded by martensite would have a better chance of resisting transformation since it is essentially shielded by a stronger phase. The RA morphology and surrounding microstructure seem to be the two driving forces for RA stability in this study and are believed to be the contributing factors for comparable HERs in the QP steels.

5.6 Critical CTOA

The CTOA results showed that the sample orientation had an effect on the critical CTOA values. The L-T orientation proved to have the better cracking resistance in comparison to the T-L orientation for all the steel grades tested which is in agreement with the literature [76]. It is interesting that both 1180 steel grades seem to have a lower sensitivity to the orientation effects
than the 980 grades. For the DP steels, it is possible that the degree of banding plays an effect on the lower sensitivity seen by the DP 1180 since the DP 1180 had a lower AI than the DP 980 (1.64 vs 2.30). The higher AI means that the degree of banding is higher and could have affected the critical CTOA of the DP 980 more than the DP 1180. However, the degree of banding does not explain the lower sensitivity to the orientation effects exhibited by the QP 1180 since the QP 980 did not exhibit microstructural banding and the QP 1180 showed microstructural variation where banding occurred in some areas and not in others. It is typical that during initial crack extension, the CTOA values are rather high and this is attributed to crack tunneling. However, the QP 1180 does not seem to show this like the other steel grades did.

CTOA has been shown to adequately characterize stable crack propagation. A previous study used crack propagation resistance to compare with edge cracking resistance for various AHSS including DP 1000, QP 1180, complex phase steels, TRIP bainitic ferrite steels, and press hardened steels [78]. During HET, for a crack to initiate on the surface of the sample it is preceded by crack initiation and propagation within the thickness of the material. Therefore, an understanding of crack propagation would help in characterizing edge cracking resistance in different materials. Figure 5.8 plots CTOA for the L-T orientation and various mechanical properties taken along the rolling direction: (a) TE, (b) ROA, (c) UTS*TE, and (d) 0.5*(UTS+YS)*TE. These comparisons show minimal data; therefore, a true correlation cannot be made but an understanding of mechanical properties and crack propagation is gained. Figure 5.9 shows the critical CTOA determined in the T-L orientation plotted against these mechanical properties taken transverse to the rolling direction.

Higher critical CTOA values mean the higher the fracture resistance and better toughness a material has. Therefore, the mechanical properties that seem to correlate well with CTOA (TE, ROA, UTS*TE, 0.5*(UTS+YS)*TE) make sense. TE and ROA are indicative of ductility, and toughness can be characterized as a good combination of strength and ductility. Since all the materials tested are considered to be of relatively high strength, higher ductility will then lead to a better toughness. UTS*TE is indicative of energy absorption during deformation which is essentially the definition of toughness and therefore correlates well with CTOA. Another toughness parameter (0.5*(YS+UTS)*TE) commonly used correlates well with CTOA. Other mechanical properties such as the YS and YS/UTS ratio did not seem to correlate as well.
Figure 5.8  Critical CTOA values of each steel grade in the L-T orientation plotted against mechanical properties taken along the rolling direction: (a) TE, (b) ROA, (c) UTS*TE, (d) 0.5*(YS+UTS)*TE.
Figure 5.9  Critical CTOA values of each steel grade in the T-L orientation plotted against mechanical properties taken transverse to the rolling direction: (a) TE, (b) ROA, (c) UTS*TE, and (d) 0.5*(YS+UTS)*TE.

Figure 5.10(a-d) shows HER plotted against the critical CTOA for each steel grade under each testing condition. The error bars shown in Figure 5.10(a-d) are the standard deviation for HER and critical CTOA. The DP steels correlate well in that increasing HER comes with increasing critical CTOA values for all testing configurations. However, in the QP steels it is
difficult to differentiate a trend in the CS, CM, and FS testing configurations as the HERs are similar between the two steel grades. The FM testing condition proved to show the best trend because it is believed that it keeps the two different methods under similar testing conditions. The flat bottom punch creates an in-plane testing condition, which is the testing condition during CTOA since the sample is in tension and the buckling plates are preventing any out of plane deformation. The sample preparation is also similar in that the machined edge does not create a SAZ and during CTOA testing the sample is fatigue pre-cracked so the starting condition is almost defect free in that it is away from the EDM edge.

To better understand the relationship between HER and critical CTOA we will look into HET and the manner in which crack initiation occurred. Higher HERs come from those materials that have a higher critical CTOA when one thinks about how cracking occurs during HET. During HET, the test is stopped when a through-thickness crack is initiated on the surface of the sample. In order for crack initiation to occur on the surface, microscopic crack propagation through the material thickness has to occur. Therefore, the higher the crack propagation resistance is, the longer it will take for a crack to initiate on the surface and hence a relationship can be made between HER and CTOA.

A relationship has also been made with the orientation effects during CTOA testing and crack initiation during HET. It was determined that a majority of the samples tested for hole expansion performance showed crack initiation along the rolling direction. Also all the samples tested in the T-L orientation, meaning the crack propagated along the rolling direction, had a lower critical CTOA than the L-T orientation. This means cracking should occur along the rolling direction which is what was seen in a majority of the hole expansion samples. This then is another way to connect HER testing results and critical CTOA values.
Figure 5.10  HER of all steel grades in the (a) CS, (b) CM, (c) FS, and (d) FM testing conditions plotted against critical CTOA in the L-T orientation. Error bars are from the standard deviation.
Figure 5.11 HER of all steel grades in the (a) CS, (b) CM, (c) FS, and (d) FM testing conditions plotted against critical CTOA in the T-L orientation. Error bars are from the standard deviation.
CHAPTER 6
CONCLUSIONS

Advanced high strength steels (AHSS) have shown impressive mechanical properties and are a viable option for the automotive industry as they deal with increasing fuel economy and safety standards, however, these higher strength steels exhibit edge cracking at strains lower than those predicted by forming limit curves during the stamping process. The automotive industry uses hole expansion testing (HET) to help explain the edge fracture seen in AHSS. The purpose of this thesis was to gain a better understanding of the hole expansion ratio (HER) with respect to mechanical properties and microstructures of AHSS at strength levels near 1 GPa and above, and the use of a fracture toughness measurement to help understand HER and the resistance to crack propagation. The materials of interest were dual phase (DP) 980, DP 1180, intercritically annealed (IA) quenched and partitioned (QP) 980, and IA QP 1180.

The following conclusions were made:

- Despite various testing conditions, the DP steels in this study showed that the hardness difference between phases and the yield strength/ultimate tensile strength ratio correlate well with HER. However, the QP steels did not show a trend between mechanical properties and HER since the HERs were similar for the two steel grades. Therefore, there does not seem to be a shared mechanical property that was tested that governs HER in DP and QP steels, which leads us to believe that microstructure plays a larger role in HER.

- In this study, DP 980 always had a higher HER than the DP 1180 and microstructurally we focused on the martensite morphology. The DP 980 showed a martensite network along ferrite grain boundaries which has been shown to prevent crack propagation and shields ferrite during deformation [59]. Therefore, this morphology in combination with reducing the hardness difference between phases should be at the forefront when it comes to DP microstructural design.

- In this study, QP 980 and QP 1180 had similar HERs and we focused on the retained austenite (RA) stability to better explain this. HER performance showed that a steady, continuous transformation of RA during deformation is desired. This type of RA stability was exhibited by the QP 1180 which showed comparable HER performance to the
QP 980. The RA stability seen in the QP steels was attributed to the combination of RA morphology and the surrounding microstructure.

- HET with various punch geometries and edge conditions proved that the relative HER was similar for all testing conditions with the exception of the flat bottom punch/machined edge condition for the DP and QP steels studied.

- Critical crack tip opening angle (CTOA) values correlated well with HER in that higher critical CTOA values led to a higher HER. The specimen orientation effects during CTOA testing also correlated with the direction of crack initiation seen during HET. CTOA is then a test method that can help understand the fundamentals of crack resistance and use this understanding to better understand crack initiation during HET.
Opportunities for possible projects in the future that were not achieved in this study.

- It would be interesting to perform the same characterization and mechanical testing techniques performed in this research project, but with a wider variety of high strength DP and QP steels with a specific distribution of microstructures such as a fully austenitized QP steel or DP steels with MVFs greater than 70 pct. This would help fill in some of the gaps such as MVFs between 70 and 100 pct. as well as looking into the varying microstructures and what affect they would have.

- It is important to revisit the methods used to measure RA volume fractions. The XRD method proved to have many variables that could affect the measured volume fraction. Therefore, it would be of interest to perform many XRD tests with different parameters such as radiation source, sample tilting and rotation. This could then help achieve a much more accurate bulk RA volume fraction.

- With respect to RA, it would be interesting to explore the RA volume fraction after the shearing process, characterization of the SAZ, and how that would affect the measured HER.

- Further analysis of DP steels should involve trying to design a microstructure that can retain the martensite network morphology along ferrite grain boundaries with slight variations of MVF, carbon content, solid solution strengthening, or heat treatments and still reach strength levels of the DP 1180.

- Further analysis of QP steels should involve looking into the effects of carbon depleted and fresh martensite and their effects on RA stability and the HER. It would also be useful to quantify hardness among the different constituents.

- Finally, CTOA testing with more materials but also implementing DIC measurement techniques to remove human error. There is also a need for techniques to improve crack tip visibility as oppose to just polishing the material. It would also be interesting to measure RA at different distances ahead of the crack tip and create a transformation graph. Temperature measurements would also be interesting around the crack propagation area.
REFERENCES


This appendix contains SEM micrographs that were taken at different magnifications. Figure A.1 shows representative micrographs that were used to calculate the MVF in the DP steels. It is important to note that both steels showed evidence of microstructural variation where some micrographs have heavy martensitic areas and others did not. A mixture of the two examples were taken into account when determining the MVF.

Figure A.1 Secondary electron images of (a-b) DP 980 and (c-d) DP 1180 in the planar orientation etched with 2 pct. Nital. The rolling direction is horizontal. Images were taken at 3000x magnification.
Figure A.2 shows representative micrographs that were used to calculate the M-A volume fraction in the QP steels. As you can see the QP 1180 does show some areas that have a minimal amount of ferrite (Figure A.2(d)) and other areas with a good amount of ferrite (Figure A.2(c)). An average amount of images of both these representative areas were used in calculating the M-A volume fraction.

Figure A.2 Secondary electron images of (a-b) QP 980 and (c-d) QP 1180 in the planar orientation etched with 2 pct. Nital. The rolling direction is horizontal. Images were taken at 1500x magnification.
This appendix contains EBSD scans of the QP steels that had a minimum CI of 0.5 and have not been filtered to show a 0.1 minimum CI. Figure B.1 shows EBSD scans of the QP 980 at different areas within the sheet and at different magnifications. The darker areas are indicative of transformed RA that could have occurred during cross-section polishing as the EBSD scans can still index traces of RA in those areas. Figure B.2 shows EBSD scans of the QP 1180 and there is not as much dark areas as the QP 980. This can be related to the higher RA stability that the QP 1180 showed and therefore, preparation techniques did not affect it as much.

Figure B.1 Superimposed image quality and phase maps of QP 980 obtained through EBSD with a scan size of 5x5 microns. The red represents BCC-FE (ferrite or martensite) and the green represents FCC-FE (austenite). The darker areas are transformed RA. (a) This figure is taken at 3000x magnification with a calculated RA volume fraction of 6 pct. (b) This figure is taken at 5000x magnification with a calculated RA volume fraction of 4 pct.
Figure B.2 Superimposed image quality and phase maps of QP 1180 obtained through EBSD. The red represents BCC-FE (ferrite or martensite) and the green represents FCC-FE (austenite). The darker areas are transformed RA. (a) This figure is taken at 1500x magnification and a scan size of 10x10 microns with a calculated RA volume fraction of 5 pct. (b) This figure is taken at 5000x magnification and a scan size of 5x5 microns with a calculated RA volume fraction of 4 pct.
APPENDIX C
SHEARED EDGE CHARACTERIZATION

A side study was performed to look at the different regions on the face of the sheared edge and to understand the work hardening effects of each material due to the shearing process. The regions of the face of the sheared edge were characterized using an optical stereoscope. Sheared hole test blanks that were not used for HET were cut in half and images were taken in three different areas of the sheared face at a magnification of 50x so that all regions on the sheared face were visible. Each image resulted in three different measurements of the rollover, burnish, and fracture plus burr regions using ImageJ analysis software. A total of nine measurements for each region were measured and averaged with the standard deviation as well.

Figure C.1 shows an example of an image taken on the optical stereoscope to calculate the percent of sheet thickness for each region. The region labeled as (1) is the rollover, (2) is the burnish, and (3) is the fracture plus burr regions. All samples had the same clearance of approximately 11 pct. of the sheet thickness. Figure C.2 shows the percentage of each region of the sheared face.

Figure C.1 Sample stereoscope image of the sheared face of QP 980. The regions were measured in the ImageJ image analysis software. Region 1 is the rollover, region 2 is the burnish, and region 3 is the fracture and burr regions.
In order to quantify the work hardening effects at the sheared edge, Vickers micro-hardness measurements were performed. Cross sectioned samples in the longitudinal and transverse orientations were tested with both 500gf and 50gf for a dwell time of 10 seconds. The spacing between each indent was 200 microns and 75 microns for the 500 gf and 50 gf loads, respectively. The 50 gf load was used to provide hardness measurements as close to the shear face as possible to give a better indication of the maximum work hardening of the material.

The shearing process creates a SAZ where the material is work hardened and a strength gradient is created which can lead to a reduction in ductility and cracking at low strains. Figure C.3 shows sample micro-hardness profiles along the rolling direction of each steel. From these profiles, an estimate of the relative depth of the SAZ can be calculated as summarized in Table C.1.

<table>
<thead>
<tr>
<th>Material</th>
<th>Depth of SAZ - RD (microns)</th>
<th>Depth of SAZ - TR (microns)</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP 980</td>
<td>410</td>
<td>380</td>
</tr>
<tr>
<td>DP 1180</td>
<td>275</td>
<td>360</td>
</tr>
<tr>
<td>QP 980</td>
<td>400</td>
<td>400</td>
</tr>
<tr>
<td>QP 1180</td>
<td>310</td>
<td>390</td>
</tr>
</tbody>
</table>
Figure C.3  Microhardness plots in the rolling direction. The plot shows both the 500 and 50 gf loading conditions. Measurements were taken from the sheared edge until the base material hardness was measured. Rockwell C hardness data was converted to Vickers hardness data to assess when the base material hardness was measured.

The relative depth of the SAZ was calculated using the 500 gf data and the intersection of a liner fit of the data points that were close to the base hardness of the material with a curve fit of the data points that were harder than the base material. Both 980 steels show a higher depth of the SAZ than the 1180 steels.
Figure D.1 and Figure D.2 shows the crack initiation sites for all steels grades. As mentioned in the procedure section, if the crack initiated within 60 degrees (dictated by the red lines drawn on the image) then the crack was determined to have initiated along the rolling direction. All samples showed crack initiation along the rolling direction.

Figure D.1 Hole expansion samples showing crack initiation sites in (a) DP 980 and (b) DP 1180 in all testing configurations. The number of tests with crack initiation along the rolling direction were DP 980: 10/18 and DP 1180: 15/18.
Figure D.2  Hole expansion samples showing crack initiation sites in (a) QP 980 and (b) QP 1180 in all testing configurations. The number of tests with crack initiation along the rolling direction were QP 980: 15/18 and QP 1180: 11/18.
This appendix section contains fractography performed on the CTOA samples to clarify that crack tunneling effects were minimized during testing. In order to determine the amount of crack tunneling each sample exhibited, Figure E.1 was used as a reference where $B$ is the sample thickness, $w$ is the width and $c$ is the initial crack length once fatigue testing is completed [94]. Crack tunneling is shown in Figure E.1 as the triangular region between the fatigue pre-cracking and shear mode stages. Once this region transitions to shear mode, crack tunneling is minimized. Therefore, measurements of the crack tunneling area were done from the fatigue surface to the tip of the triangular region.

Figure E.1  The typical fracture surface of a CT specimen after CTOA testing. $B$ is the sample thickness, $w$ is the width, and $c$ is the initial crack length. The triangular area shown is indicative of crack tunneling. [94]

Figure E.2 shows fractography of DP 980 and DP 1180 in the T-L orientation. The fracture surfaces show that crack tunneling is minimized at 3 mm (DP 980) and 2 mm (DP 1180) of crack extension.
Figure E.2 Fracture surfaces of (a) DP 980 and (b) DP 1180 CTOA samples in the T-L orientation.

Figure E.3 shows fractography of QP 980 and QP 1180 in the T-L orientation. The fracture surfaces show that crack tunneling is minimized at approximately 2.5 mm (QP 980) and 3 mm (QP 1180) of crack extension.
Figure E.3 Fracture surfaces of (a) QP 980 and (b) QP 1180 CTOA samples in the T-L orientation.

Figure E.4 shows the fracture surface of QP 1180 in the L-T orientation. The fracture surface in this orientation shows that crack tunneling is minimized at approximately 2 mm of crack extension. Therefore, crack tunneling was minimized in all samples in two different orientations. The critical CTOA was taken from a minimum crack extension of 8 mm which is away from the effects of crack tunneling seen in the fracture surfaces. Therefore, the crack tunneling effects are in agreement with the CTOA standard in that crack tunneling is avoided where the critical CTOA is measured.
Figure E.4  Fracture surface of QP 1180 CTOA sample in the L-T orientation.
CTOA values are not used in the industry as a parameter to characterize a materials toughness. As mentioned throughout the thesis, it is not feasible to calculate a \( K_{IC} \) value for sheets steels which is a parameter of toughness that most people are accustomed to. However, CTOA can be related to a stress intensity factor under mode I loading \( (K_I) \). Equation F.1 shows how CTOA can be related to crack tip opening displacement (CTOD).

\[
\Psi_c = 2\tan^{-1}\left( \frac{\delta_c}{2d} \right) \tag{F.1}
\]

Where \( \Psi_c \) is the critical CTOA value, \( \delta_c \) is CTOD and \( d \) is specified distance from the crack tip where measurements were taken which would be just under 1 mm for this study.

Once a CTOD value is calculated, it can be related to the stress intensity factor under mode I loading using the following equation:

\[
\delta_c = \left( \frac{K_I^2}{m\sigma_y E} \right) \tag{F.2}
\]

where \( m \) is a factor used to account for the level of plane stress or plain strain, \( \sigma_y \) is the yield stress of the material, and \( E \) is the young modulus. Table F.1 shows the calculated stress intensity factors for the materials in this study using the material properties and critical CTOA determined in the L-T orientation.

<table>
<thead>
<tr>
<th>Material</th>
<th>( K_I ) (MPa*mm(^{1/2}))</th>
</tr>
</thead>
<tbody>
<tr>
<td>DP 980</td>
<td>170.5</td>
</tr>
<tr>
<td>DP 1180</td>
<td>159.9</td>
</tr>
<tr>
<td>QP 980</td>
<td>188.2</td>
</tr>
<tr>
<td>QP 1180</td>
<td>189.5</td>
</tr>
</tbody>
</table>

Table F.1 – Stress Intensity Factors under Mode I Loading