TENSILE DEFORMATION OF THIRD GENERATION ADVANCED HIGH STRENGTH SHEET STEELS UNDER HIGH STRAIN RATES

by
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ABSTRACT

New third generation advanced high strength steels (AHSS) are being developed for structural components of vehicles to reduce vehicle weight and improve crash performance. The goal of this work was to gain an understanding of the role of retained austenite in the high strain rate tensile deformation behavior of third generation AHSS and to understand the effect of adiabatic heating on the austenite to martensite transformation. Four steels were selected for the study: an experimental medium Mn transformation induced plasticity (TRIP) steel (TRIP7Mn), an experimental quenched and partitioned (Q&P) steel (QP3Mn), and two commercially produced steels, DP980 and QP980.

The effect of strain rate on the deformation behavior of QP3Mn and TRIP7Mn, was examined through tensile tests in the range of 0.0001 – 200 s$^{-1}$ and retained austenite X-ray diffraction measurements from interrupted tensile tests at various strain rates. The effects of adiabatic heating on flow stress and retained austenite stability were further investigated with quasi-static tensile tests at 22 – 85 °C for QP3Mn and 22 – 115 °C for TRIP7Mn. The combined effects of temperature and strain rate on flow stress and the austenite to martensite transformation were also examined with tensile tests at -10, 22, and 70 °C at 0.0005, 0.1, and 10 s$^{-1}$. Room temperature tensile tests in the range of 0.0001 – 100 s$^{-1}$ were performed on the DP980 and QP980 for as-received and pre-strained conditions.

The QP3Mn has positive YS strain rate sensitivity (meaning the overall trend in flow stress with increasing strain rate) possibly related to the contribution of austenite to yielding, and it has low UTS strain rate sensitivity due in part to the mostly martensitic microstructure at UTS. The TRIP7Mn upper and lower YS have positive strain rate sensitivity, but the TRIP7Mn has significant negative UTS strain rate sensitivity up to 1 s$^{-1}$ with slight positive strain rate sensitivity above 10 s$^{-1}$. Elongation follows a similar trend as UTS. The UTS and elongation trends are due to a combination of factors: the disappearance of serrated plastic flow with increasing strain rate, a softening effect from adiabatic heating, and stabilization of austenite due to adiabatic heating. The austenite in the QP980 may lead to a slightly higher YS strain rate sensitivity compared to the DP980, but it does not affect trends in UTS or ductility with respect to strain rate. The DP980 and QP980 have slightly negative UTS strain rate sensitivity below 1 s$^{-1}$ due to a softening effect from adiabatic heating. The austenite to martensite transformation is not affected by strain rate for the QP980 and QP3Mn; whereas, the TRIP7Mn has a lower transformation rate above 10% strain at high strain rates (0.1 – 100 s$^{-1}$) due to adiabatic heating stabilizing the austenite.
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CHAPTER 1
INTRODUCTION

In the past few decades, advanced high strength steels (AHSS) such as transformation induced plasticity (TRIP), dual phase (DP), and press hardening steels (PHS) have been incorporated into vehicles to reduce the body weight and improve strength of body parts [1–3]. More stringent fuel economy standards and crash tests continue to drive the implementation and development of AHSS. The U.S. fuel economy standard established by the Environmental Protection Agency (EPA) and the National Highway Traffic Safety Administration (NHTSA) for passenger cars and light trucks mandates a fleet wide average of 54.5 mpg in 2025 [4]. Concurrently, the Insurance Institute for Highway Safety (IIHS) continues to devise new crash tests to improve safety, such as the small overlap front test implemented in 2012 [5]. The new small overlap front test raises challenging design issues that may be resolved in part with new AHSS [1].

Third generation AHSS that are currently under development will play a critical role in vehicle weight reduction and improved crash performance for future vehicle designs. Retained austenite is a key microstructural constituent that contributes to the high strength and ductility combinations in third generation AHSS [6]. A fundamental understanding of the contribution of retained austenite to the overall high strain rate deformation behavior will be beneficial for forming components, selecting alloys with optimal crashworthiness, and designing future AHSS grades for vehicles. Under forming operations and crash conditions, steel components are subjected to deformation at a wide range of strain rates, some of which can be studied with tensile tests in the strain rate range of $0.01 \to 1000 \text{ s}^{-1}$ [7]. The mechanical property information from high strain rate tensile tests is potentially useful for developing forming models and automotive crash models to predict the performance of sheet steel components in the vehicle body.

1.1 Research Objectives

The goal of this work is to gain an understanding of the role of retained austenite in the high strain rate deformation behavior of new third generation AHSS. Adiabatic heating at strain rates above quasi-static rates could have a stabilizing effect on the retained austenite, which could in turn reduce ductility and strength. Previously, the effect of strain rate on the strain-induced austenite to martensite transformation during room temperature tensile
tests in the range of $0.0001 - 1000 \text{ s}^{-1}$ was studied in austenitic stainless steels and first generation AHSS [8–16]. In austenitic stainless steels, the transformation rate was reduced with increasing strain rate, which was attributed to the stabilizing effect of adiabatic heating [8–13]. However, there was very little or no significant difference in the transformation rate with increasing strain rate for first generation TRIP steels [14–16]. The current work focuses on understanding the effect of adiabatic heating on retained austenite in third generation AHSS during tensile deformation in the strain rate range of $0.0001 - 200 \text{ s}^{-1}$.

There are many different austenite containing microstructures that produce tensile mechanical properties within the third generation AHSS regime [6]. A medium Mn TRIP steel and a quenched and partitioned (Q&P) steel were selected for this study. The material selection was narrowed down to two experimental grades that have very different quasi-static mechanical properties with just two as-processed microstructural constituents in each steel. This material selection allows for a comparison between the high strain rate deformation behavior of austenite-ferrite (medium Mn TRIP) and austenite-martensite (Q&P) microstructures. Two commercially produced steels, DP980 and QP980, that have tensile mechanical properties between first and third generation AHSS mechanical properties were examined with a smaller experimental matrix to help interpret the differences in mechanical behavior at high strain rates between first and third generation AHSS. The DP980 and QP980 were also used in a pre-strain study that is relevant to the performance of formed components during a vehicle crash. There are two specific research objectives with the DP980 and QP980: 1) to determine whether the austenite in the QP980 has an effect on high strain rate tensile deformation behavior compared to the equivalent strength DP980 and 2) to determine whether prior tensile deformation history affects subsequent high strain rate tensile mechanical behavior in high strength steel.

The overall research objectives are:

1. To understand the effect of increasing strain rate on the stability of retained austenite and flow stress.

2. To study the effect of altering austenite stability through heating or cooling on high strain rate deformation behavior.

3. To understand the contribution of specimen heating to austenite stability and flow stress through experiments and modeling.
1.2 Thesis Overview

The literature review in Chapter 2 provides an overview of AHSS including discussion of the various generations and the different classifications of microstructures and properties. The processing and properties associated with Q&P and medium Mn TRIP steels are discussed in detail. Background is provided on the strain-induced austenite to martensite transformation at quasi-static strain rates, and the effect of strain rate on the transformation is also discussed. Finally, strain rate sensitivity of steel, specimen self-heating during high strain rate tensile tests, and the high strain rate deformation response of various AHSS are covered.

The experimental design approach and experimental methods are presented in Chapter 3. Background on the steels is provided. The experimental methods that are discussed include Q&P heat treatments, metallography preparation and characterization, X-ray diffraction (XRD), \textit{in situ} temperature measurement techniques, low strain rate and high strain rate tensile testing, and data analysis approaches for high strain rate tensile data. Updates to the high strain rate tensile test system are also presented.

Chapter 4 contains the results that are organized into separate sections for QP3Mn, TRIP7Mn and DP980 and QP980 combined. For the QP3Mn and TRIP7Mn, the tensile test results at various strain rates include mechanical properties, specimen temperature measurements, and retained austenite measurements as a function of strain. Mechanical properties and retained austenite measurements are also reported for 1) various elevated temperatures at a quasi-static strain rate and 2) a low and a high temperature condition at various strain rates. The DP980 and QP980 results include mechanical properties at various strain rates for the as-produced specimens and also for three pre-strain conditions. Specimen temperature and retained austenite measurements were also performed for the QP980.

Chapter 5 contains a discussion of the results. The effect of strain rate on the flow stress is compared for the four steels in this study. The role of adiabatic heating in the high strain rate tensile deformation behavior of the QP3Mn and TRIP7Mn is examined through strain rate change tests and tests that were interrupted to allow for cooling. The change in the specimen temperature at UTS due to adiabatic heating is predicted for various strain rates using a model in an effort to predict the specimen temperature at strain rates that are too high for \textit{in situ} temperature measurements. The effect of strain rate on the austenite to martensite transformation is discussed for the QP980, QP3Mn, and TRIP7Mn. Finally, a modified Olson-Cohen model for the austenite to martensite transformation kinetics that accounts for adiabatic heating at high strain rates is presented. Retained austenite mea-
urements from the QP3Mn and TRIP7Mn quasi-static tensile tests at elevated temperature were used to determine the effect of temperature on the Olson-Cohen model parameters.

In Chapter 6, a summary of the major results is provided and conclusions are drawn regarding the role of retained austenite in the high strain rate deformation response of third generation AHSS. The future work recommendations in Chapter 7 include advanced characterization options for microstructures deformed at various strain rates, possible routes for expanding modeling efforts, and possible options for further examining mechanical behavior at high strain rates.
CHAPTER 2
LITERATURE REVIEW

This chapter begins with a brief overview of the various types of advanced high strength steel (AHSS). The alloying, processing, microstructure, and tensile mechanical properties of Q&P and medium Mn TRIP steels are discussed in detail. The theory behind the strain-induced austenite to martensite transformation and a model for the transformation are presented. Factors affecting the retained austenite stability in AHSS are also reviewed. Finally, the high strain rate tensile deformation section provides a discussion of the effects of strain rate on deformation mechanisms, adiabatic heating, first generation AHSS tensile flow properties, and the strain-induced austenite to martensite transformation in metastable austenitic stainless steels and first generation TRIP steels. The quantity, size, and adjacent microconstituents of austenite in third generation AHSS can differ from first generation TRIP steels and could lead to differences in high strain rate tensile behavior.

2.1 Advanced High Strength Sheet Steels

A major driving force in AHSS development has been the need to make vehicles lighter without compromising passenger safety and still minimizing cost. A strength and ductility map of a variety of traditional automotive sheet steels and AHSS is shown in Figure 2.1 [3, 17]. Advanced high strength steels have higher strength-ductility combinations than low strength mild and interstitial free (IF) steels and high strength carbon-manganese (C-Mn), bake hardenable (BH), high-strength IF, and high-strength low-alloy (HSLA) steels [3]. Complex phase (CP), dual phase (DP), TRIP, and martensitic (MART) steels are classified as first generation AHSS, and austenitic stainless steels, twinning-induced plasticity (TWIP) steels, and lightweight steels with induced plasticity (L-IP) are considered to be second generation AHSS [3, 18]. DP, TRIP, and TWIP steels will be discussed in more detail here, since they are the types of AHSS that have been the focus of much of the recent high strain rate literature. First generation AHSS, especially DP steels, are being incorporated into vehicle designs [19]. These steels have multi-constituent microstructures, and consequently, they exhibit more complex deformation behavior than traditional ferritic steels. Most DP steels consist of a ferrite matrix with varying amounts of martensite, and they have characteristic high ductility and high work hardening rates [3]. The TRIP steel microstructure consists of bainite, a minimum of 5 vol.% of retained austenite, and possibly martensite in a ferrite
matrix [3]. TRIP steels also have a high work hardening rate from the combination of a soft ferrite matrix and islands of harder phases, but the work hardening rate is further increased at higher strain levels compared to DP steels due to the progressive strain-induced austenite to martensite transformation [3]. Additions of Si, Al, and/or P to TRIP steel suppress carbide precipitation during bainite formation, which enriches the carbon content in austenite to stabilize it at room temperature [20]. TWIP steels are alloyed with 17 – 24 wt.% Mn to produce a fully austenitic microstructure at room temperature, and a significant portion of the plastic deformation occurs by deformation twinning, which contributes to very high strength-ductility combinations [3]. However, automotive applications of second generation AHSS are limited due to the high cost associated with alloying and processing issues [3].

![Total elongation versus tensile strength map of low strength, high strength, and advanced high strength steel grades. Figure from the WorldAutoSteel group [3].](image)

The property gap between first and second generation AHSS presents an opportunity for the development of new third generation AHSS. Composite modeling suggests that strength-ductility combinations between first and second generation AHSS can be obtained by microstructures containing various amounts of high ductility metastable austenite and another high strength phase, such as martensite or strengthened ferrite (e.g. ultra fine grained ferrite) [18]. The basic design approach has been to reduce the amount of costly alloying elements in second generation AHSS or modify the microstructures of first generation AHSS through new alloying and processing techniques [6]. The new third generation AHSS designs include modified TRIP, medium Mn TRIP, TRIP/TWIP, bainitic, and Q&P steels [6]. This work
focuses on Q&P and medium Mn TRIP steels, which will be discussed in more detail in Sections 2.2 and 2.3, respectively.

### 2.2 Quenched and Partitioned Steels

The premise of Q&P steels is the partial formation of martensite during a quench step followed by enrichment of the remaining austenite by carbon during the partitioning step [21]. Quenched and partitioned sheet steels were initially developed based on TRIP steel chemistries, which contain Si and Al to suppress carbide formation and C and Mn as austenite stabilizers [22–24]. For Q&P steels, carbide formation must be suppressed through alloying additions such that C diffusion into the austenite is maximized [21]. Silicon is effective at suppressing carbide formation, and Mo increases austenite volume fractions [25, 26]. Although Al also prevents carbide formation, it results in lower austenite fractions in Q&P steels possibly due in part to increasing bainite transformation kinetics [26]. Manganese additions enhance the austenite stability and suppress the formation of ferrite, pearlite, and bainite during processing [27]. The alloying and processing of the Q&P steel studied in this work is based on the work of De Moor et al. on a 0.3C-3Mn-1.6Si (wt.%) alloy, which will be discussed throughout this section [28].

The Q&P processing and properties discussed in this section will be focused on full austenitization and two step partitioning (partitioning at a temperature higher than the quench temperature) of the steel, since these processing steps produce austenite-martensite microstructures with mechanical behavior of interest for this work. The Q&P process, shown schematically in Figure 2.2a, is conceptually based on the constrained paraequilibrium (CPE) model for carbon diffusion between ferrite and austenite in the absence of carbide formation to obtain an equal carbon chemical potential in both phases, which results in carbon enrichment of the austenite [21]. In Figure 2.2a, $A_{c3}$ is the temperature at which full austenitization will occur upon heating, $M_S$ is the martensite start temperature, $M_F$ is the martensite finish temperature, QT is the quench temperature, PT is the partitioning temperature, and the C designations refer to carbon content. After austenitizing, the steel is quenched to partially form martensite. Following the quench step, carbon diffuses from the martensite to the austenite during the partitioning step thereby stabilizing the austenite at room temperature. For typical partitioning temperatures and times ($PT \leq 450^\circ C$ and $Pt \leq 1000$ s), only carbon diffusion occurs during the partitioning step, while the larger substitutional elements such as Si and Mn remain relatively immobile [27]. The fraction of martensite that is initially formed depends on the martensite start temperature ($M_s$) and the quench temperature (QT). The modified Andrews equation can be used to predict the $M_s$ temperature for Q&P
alloying schemes and is given by Equation 2.1, where C, Mn, Si, and Al are in weight percent [29–31]. The modified Andrews equation has Si and Al terms added and Cr, Ni, Mo terms removed, which reflects the common alloying approaches for experimental TRIP-aided steels [31–33]. The volume fraction of martensite \( f_m \) that is formed by a quench step can be estimated by the Koistinen-Marburger relationship given by Equation 2.2 [34].

\[
M_s(\degree C) = 539 - 423C - 30.4Mn - 7.5Si + 30Al \\
\]  
\( f_m = 1 - e^{-1.1\times10^{-2}(M_s - QT)} \)  

The fraction of retained austenite produced by the Q&P process can be calculated by applying the Koistinen-Marburger relationship to the initial and final quench steps and assuming complete carbon partitioning to austenite during the partitioning step and a stationary austenite-martensite interface [21, 22]. A plot of austenite and martensite fractions as a function of quench temperature for a 0.2C-1.6Mn-1.6Si-0.04Al (wt.%) steel is provided in Figure 2.2b. \( M_{QT} \) and \( \gamma_{QT} \) are the fractions of martensite and austenite formed at the quench temperature, \( \gamma_{FINAL} \) is the fraction of austenite at room temperature after Q&P processing, and \( M_{FRESH} \) is the fraction of fresh martensite formed upon quenching to room temperature. There is a peak quench temperature that corresponds to the maximum amount of retained austenite in the microstructure. At the peak quench temperature, carbon partitioning into the austenite depresses the \( M_s \) temperature to room temperature and no new martensite will form upon the final quench. Below the peak quench temperature, a large fraction of austenite is transformed to martensite during the initial quench step, and there will be more than enough carbon in the austenite to stabilize it at room temperature after the partitioning step. Above the peak quench temperature, a large fraction of austenite remains after the initial quench and carbon partitioning cannot stabilize the full remaining austenite fraction, which results in fresh martensite forming during the final quench to room temperature [22]. For this work, a low quench temperature was selected to avoid fresh martensite in the microstructure such that only partitioned martensite contributed to the mechanical behavior.

An example of experimentally obtained retained austenite amounts compared with the Q&P model for a 0.3C-3Mn-1.6Si (wt.%) alloy is shown in Figure 2.2c [28]. The peak value of retained austenite obtained experimentally is usually below the model prediction due to incomplete carbon partitioning [35]. Atom probe tomography (APT) has shown that during partitioning the amount of carbon in solution in the martensite decreases due to carbon enrichment of the austenite, carbon atom clustering in the martensite, and carbide formation.
Figure 2.2  (a) Schematic of a quenching and partitioning processing route. The microstructure sketches illustrate the fully austenitic starting microstructure, followed by the partial formation of martensite due to the initial quench step, next the carbon content of the austenite increases while the carbon content of the martensite decreases during the partitioning step, and finally fresh martensite can form upon the final quench.  (b) Retained austenite and martensite fractions as a function of quench temperature for a 0.2C-1.6Mn-1.6Si-0.04Al (wt.%) steel. The curves are based on full austenitization and full carbon partitioning from martensite to austenite. Figures from Streicher et al. [22].  (c) Experimental XRD results (data points) and model prediction (curve) of retained austenite as a function of quench temperature for a 0.3C-3Mn-1.6Si (wt.%) Q&P steel that was austenitized at 820 °C for 120 s and partitioned at 400 °C for various times. Data plotted from De Moor et al. [28].
in the martensite [36]. The retained austenite fraction tends to increase with partitioning time up to a peak value and decreases at longer partitioning times [35]. The increase in the austenite fraction corresponds to an increase in austenite carbon content as measured from the austenite lattice parameter from XRD [35, 36]. Electron backscatter diffraction (EBSD) results suggest that the increase in the austenite content can be due to the growth of austenite present after quenching [36]. During *in situ* hot stage TEM experiments of a 1C-3Mn-1.5Si (wt.%) Q&P steel, the width of a film-like austenite grain between martensite laths grew approximately 15% during partitioning at 400 °C for 1000 s [37]. Kinetic modeling of the retained austenite volume fraction as a function of partitioning time suggests that the movement of the austenite-martensite interface may lead to slightly higher than predicted maximum retained austenite volume fractions or a second maximum retained austenite fraction at long partitioning times [38]. The decrease in austenite content at long partitioning times often corresponds with a decrease in the carbon content of austenite, which could be due to carbide precipitation at austenite-martensite interfaces [35, 39]. Carbon partitioning into austenite and carbide formation may occur simultaneously [36, 40, 41]. For example, eta transition carbides precipitated in a 0.38C-1.54Mn-1.48Si (wt.%) steel during partitioning at 400 °C for 300 s [40, 41]. The partitioning temperature also affects carbide precipitation with epsilon transition carbides readily precipitated at low partitioning temperatures [39, 42]. For example, a 9260 grade steel that was austenitized and quenched to 190 °C had extensive epsilon carbide precipitation after partitioning at 250 °C, but at a higher partitioning temperature of 400 °C, the dominant carbide was cementite, which was effectively suppressed through alloying [39]. Precipitation of the transition carbide at the lower partitioning temperature contributed to a lower retained austenite fraction. For this work, the partitioning time was kept short to avoid carbide precipitation as much as possible.

The Q&P processing steps produce a microstructure of retained austenite present as fine laths or films between martensite laths, which has been observed with transmission electron microscopy (TEM) [37, 39, 43]. Electron backscatter diffraction has shown that the retained austenite can also be present in a blocky morphology, and the amount of blocky austenite tends to increase with increasing austenite volume fraction [36, 44]. An example of a Q&P microstructure with film-like austenite in a martensitic matrix from a 0.3C-3Mn-1.6Si (wt.%) steel that was austenitized at 820 °C for 120 s, quenched to 200 °C, and partitioned at 400 °C for 30 s is shown in Figure 2.3a [28]. Figure 2.3b shows representative stress-strain curves for this alloy. The continuous yielding, rapid work hardening upon yielding, and high strength are characteristic of the mostly martensitic microstructure with high initial dislocation density [24, 45]. In a study of a 0.17C-1.65Mn-0.38Si-1.11Al-0.08P (wt.%) steel that was processed by quench and tempering (Q&T), Q&P, and austempering, the tensile
work hardening behavior of the Q&P condition depended heavily on partitioning time and temperature but was generally intermediate between the work hardening behavior of the Q&T and austempering conditions [43]. The tensile strength and elongation as a function of partitioning time for the 0.3C-3Mn-1.6Si (wt.%) steel are plotted for several different quench temperatures in Figures 2.3c and 2.3d. The trends in the data are similar for all
of the quench temperatures. With increasing partitioning time, tensile strength decreases and elongation increases. These trends in mechanical behavior are similar to the mechanical behavior response that occurs with increasing martensite tempering time [24]. Thomas et al. suggested that a decrease in strength with increasing partitioning time may be due to a combination of carbon partitioning from martensite to austenite and martensite tempering that lowers the strength of the martensite [46]. The variation of the austenite volume fraction with partitioning time may also impact the mechanical properties, especially elongation. In a Q&P study by Santofimia et al. on 0.2C-3.5Mn-1.5Si (wt.%) and 0.2C-3.5Mn-0.5Si-0.2Al (wt.%) steels, an increase in partitioning time from 10 s to 100 s led to a decrease in yield and tensile strengths and an increase in uniform and total elongations [47]. The steel partitioned at 100 s contained twice as much retained austenite as the steel partitioned for 10 s, which may have contributed to the difference in mechanical properties.

2.3 Medium Mn Transformation Induced Plasticity Steels

The Mn content and processing steps, especially the intercritical annealing temperature, have strong effects on the microstructure and resulting mechanical properties of medium Mn TRIP steels [18, 48]. Alloys containing 5 wt.% Mn or less often have martensite in their final microstructures for all intercritical annealing temperatures [49–51]. In contrast, microstructures containing only ferrite and austenite are readily obtained for steels with 10 wt.% Mn, but the high stacking fault energy (SFE) of the austenite can result in both strain-induced transformation to martensite (TRIP) and mechanical twinning (TWIP) [52]. Microstructures that contain primarily ferrite and austenite and exhibit TRIP deformation behavior have been produced using medium Mn chemistries in the range of 6 – 7 wt.% [49, 51, 53–55]. A 7 wt.% Mn steel with a ferrite-austenite microstructure was selected for this work to minimize the number of as-heat treated microstructural constituents contributing to the flow behavior and to focus on the strain-induced austenite to martensite deformation mode.

Manganese is the major austenite-stabilizing element used in medium Mn TRIP steels due to the large amount of austenite that can be stabilized and the high strength-ductility combinations that can be obtained through careful processing at low alloying cost [6, 18, 49, 53]. The C level is often kept low in medium Mn TRIP steel due to welding concerns [56]. Silicon is another alloying element in many medium Mn TRIP chemistries [49–51, 54, 57]. Silicon is a strong solid solution strengthener of ferrite, suppresses cementite formation, reduces the $M_s$ temperature of austenite, and decreases the austenite stacking fault energy (SFE), which promotes the strain-induced austenite to martensite transformation [20].
A schematic of the basic processing route to obtain an equiaxed ferrite-austenite microstructure in a steel with 0.1 – 0.2 wt.% C and 6 – 7 wt.% Mn is provided in Figure 2.4a [51, 53–55]. Following hot rolling, cooling, and cold rolling, the microstructure consists of heavily deformed martensite, which is critical for obtaining an ultrafine grain microstructure of equiaxed ferrite and metastable austenite grains during subsequent intercritical annealing [53, 54]. Cold working the steel substantially increases the amount of room temperature stable austenite that forms during the intercritical anneal for a set amount of time as demonstrated in Figure 2.4b for a 0.053C-21Ni (wt.%) steel, because increasing cold work increases the nucleation rate during the intercritical anneal and leads to faster partitioning of the alloying elements [53]. A medium Mn TRIP steel that is cold-rolled prior to intercritical annealing requires less intercritical anneal time, on the order of minutes versus hours for the hot rolled condition, for high levels of retained austenite due to fast diffusion paths for Mn created by the deformed martensite [58]. Further, the cold worked microstructure promotes the development of an ultrafine grain microstructure of equiaxed austenite and ferrite as shown by the micrographs of annealed microstructures of a 0.11C-5.7Mn (wt.%) steel with and without prior cold work in Figures 2.4c and 2.4d [53]. Without cold work, the annealed microstructure resembles the former acicular martensite structure due to martensite plate interfaces and prior austenite grain boundaries being preferential nucleation sites, but cold work increases the number of nucleation sites and the driving force for nucleation leading to equiaxed ferrite and austenite grains [53].

The intercritical annealing step can be as short as a few minutes [54], although many studies have used hours or days [49, 51, 53, 59]. The high nucleation rate due to extensive cold work and the nucleation of two phases at the intercritical annealing temperature (IAT) lead to an ultrafine grain size and very slow grain growth due to mutual impingement of the grains from each phase [53]. For example, at an intermediate IAT for 6 – 7 wt.% Mn TRIP, the average grain size can be as small as a few hundred nanometers after annealing for a few minutes and only increase to approximately one micron after annealing for close to a day [49, 53, 58]. The potentially large amount of austenite that is stable at room temperature in medium Mn TRIP steels is attributed to the partitioning of Mn and C to austenite during intercritical annealing and an ultrafine austenite grain size [53, 54, 58, 59]. The peak amount of retained austenite that can be obtained at a particular IAT depends on balancing chemistry with grain size [59, 60]. For a constant intercritical annealing time, austenite grain size increases with increasing IAT, and the Ms temperature increases with increasing austenite grain size meaning that the austenite stability will decrease with increasing IAT for a constant intercritical annealing time [60]. At very short intercritical annealing times (on the order of minutes) for a constant IAT, the austenite particle size is very fine, but
Figure 2.4  (a) Example of the basic processing sequence to obtain an equiaxed ferrite-austenite microstructure in a medium Mn TRIP steel with 0.1 – 0.2 wt.% C and 6 – 7 wt.% Mn [51, 53–55]. (b) Curves showing the effect of cold work on austenite transformation time during an intercritical anneal of a 0.053C-21Ni (wt.%) steel. Replotted from Miller [53]. TEM micrographs of the ferrite-austenite microstructures of a 0.11C-5.7Mn (wt.%) steel with an initially martensitic microstructure after hot rolling that were (c) annealed for 16 h at 600 °C and (d) cold rolled in the range of 60 – 85 % then annealed for 16 h at 600 °C. Micrographs from Miller [53].

The volume fraction of austenite can be less than the equilibrium volume fraction due to limited time for diffusion [59]. At very long intercritical annealing times (on the order of hours to days) for a constant IAT, the austenite grains experience grain growth, which can reduce austenite stability and lead to less retained austenite in the final microstructure of medium Mn TRIP with approximately 5 wt.% or less Mn [50, 59]. For 6 – 7 wt.% Mn TRIP steels, it is possible to retain large quantities of austenite and produce a microstructure with only austenite and ferrite at long intercritical annealing times [49, 51, 53], but the long
annealing time may lead to grain growth that reduces the stability of the austenite in the final microstructure.

The maximum amount of austenite that can be obtained in medium Mn TRIP steels depends on the IAT and manganese stabilization of the austenite [33]. The IAT determines the equilibrium amount of austenite that can form, the equilibrium alloy composition of austenite, and the diffusion kinetics of carbon and manganese, which are in turn responsible for the final microstructure [33]. Gibbs provided a detailed methodology for predicting retained austenite in the final microstructure of a 7.1Mn-0.1C-0.12Si (wt.%) steel, which has a similar chemical composition to the medium Mn TRIP steel used in this work, based on the theoretical modeling approach proposed by De Moor et al. [33, 51]. The basic approach starts with utilizing ThermoCalc® to predict the equilibrium austenite fraction and the equilibrium concentration of C, Mn, and Si in the austenite as a function of IAT. Figure 2.5a shows the calculated equilibrium concentration of C, Mn and Si in the austenite as a function of IAT for the 7 wt.% Mn TRIP. Figure 2.5b shows the Mn content across an austenite grain from an EDS scan of a sample annealed at 600 °C for 196 hr. There is strong agreement between the predicted and experimentally measured Mn content of the austenite. In Figure 2.5a, the decrease in Mn content with increasing temperature and the decrease in C content above approximately 570 °C decrease the stability of the austenite formed at higher IAT. The peak in the carbon concentration curve in Figure 2.5a corresponds to the predicted temperature for carbide formation/dissolution, and the decrease in carbon content with decreasing temperature is due to carbide formation predicted by ThermoCalc®. Figure 2.5c shows the predicted intercritical austenite, athermal martensite, and retained austenite as a function of IAT. The equilibrium intercritical austenite fraction predicted by ThermoCalc® is indicated by the dashed black line in Figure 2.5c and is overlaid by the retained austenite curve below 610 °C. Based on the predicted equilibrium composition of the austenite, the Ms temperature was calculated as a function of the IAT using the Grange and Stewart equation [61]. When the calculated Ms temperature was above room temperature, the amount of athermal martensite that forms upon cooling was calculated using Equation 2.2. The complete model for the amount of retained austenite in the final microstructure, shown in Figure 2.5c, was constructed based on the ThermoCalc® prediction for the fraction of austenite formed at a particular IAT and by accounting for any austenite that would transform to athermal martensite upon cooling. Figure 2.5d shows the predicted and experimentally measured retained austenite fractions as a function of IAT for the 7 wt.% Mn TRIP. The experimental measurements of retained austenite show a similar trend as the model, but the experimental data do not correlate exactly with the model. The samples annealed at 575 – 625 °C had ferrite-austenite microstructures, while the sample annealed
Figure 2.5  Figures showing steps for predicting retained austenite amount in the final microstructure of 7.1Mn-0.1C-0.12Si (wt.%) TRIP steel, and experimental results from annealing at 196 hr for various IAT. (a) ThermoCalc® prediction of the equilibrium Mn, C and Si content of the austenite as a function of IAT. (b) EDS line scan showing Mn content measured across an austenite grain with neighboring ferrite grains from a sample annealed at 600 °C. (c) Prediction of the amount of retained austenite in the microstructure after intercritical annealing based on the ThermoCalc® prediction of equilibrium austenite fraction at specific IAT and estimates of the amount of athermal martensite to form upon cooling. The predicted room temperature microstructural constituents are indicated in the shaded regions above the plot. (d) Experimental results and model predictions of retained austenite as a function of IAT. Open symbols are from neutron diffraction and the shaded symbol is from XRD. Figures (a), (c), and (d) from Gibbs [51]. Figure (b) from De Cooman et al. [62].
at 650 °C contained a small amount of martensite in addition to ferrite and austenite and the sample annealed at 675 °C contained martensite, ferrite and a very small amount of austenite. Slight differences in IAT produced microstructures with widely varying amounts of ferrite, austenite, and sometimes martensite, which contributed in part to substantial differences in mechanical behavior.

Figure 2.6a shows engineering stress-strain curves for the various IAT for the 7.1Mn-0.1C-0.12Si (wt.%) TRIP steel from Gibbs [51]. *In situ* neutron diffraction measurements of austenite as a function of strain for each of the IAT are shown in Figure 2.6b. The following trends in the flow curves with respect to IAT have also been observed for other medium Mn TRIP steels [50, 54, 60, 63]:

- At the lowest IAT, 575 °C, the YS is high, and there is extensive yield point elongation (YPE). The TE is high, but working hardening rate is very low leading to a low UTS. The high austenite stability contributes to the low work hardening rate.
- With increasing IAT, the YS and YPE decrease while the UTS increases.
- The flow curves at intermediate IAT, 600 and 625 °C, exhibit serrated plastic flow.
- Austenite stability decreases with increasing IAT. At 575 °C, the austenite is almost completely stable, and at 650 °C, most of the austenite transforms rapidly immediately after yielding at less than 5% strain. The decrease in austenite stability with increasing IAT contributes to the increase in work hardening rate with increasing IAT.
- At the two highest IAT, 650 and 675 °C, the high UTS and low TE are due in part to the athermal martensite in the as-heat treated microstructures.

In low carbon steel, the appearance of an upper yield point followed by a drop to a lower yield point and YPE is associated with static strain aging during which carbon and nitrogen atoms diffuse to and pin edge dislocations [64]. The dislocation pinning results in a low initial mobile dislocation density, which requires a high stress to unpin the dislocations or generate new dislocations. When the mobile dislocation density increases, the stress drops from the upper yield point to the lower yield point. In low carbon steels, yielding can occur in a discrete deformed band, a Lüder’s band, that forms at a stress concentration. In a tensile specimen, the Lüder’s band often initiates at a fillet and forms at an angle to the tensile axis. Yielding proceeds in a discontinuous manner as the Lüder’s band travels along the specimen length. Yield point elongation increases with decreasing grain size in steel alloys, and ultra fine grain sizes (approximately 1 µm or less) are associated with low
Figure 2.6 (a) Stress-strain curves for a 7.1Mn-0.1C-0.12Si (wt.%) TRIP steel that was cold reduced 50 – 60 % and then annealed for 196 hr at the intercritical annealing temperatures labeled on the plot. (b) In situ neutron diffraction measurements of austenite as a function of strain for the same conditions shown in (a). Figures from Gibbs [48].

work hardening rates [53, 65]. The grain size and austenite stability in austenite-ferrite microstructures determines the YPE and work hardening behavior, as shown by the stress-strain curves in Figure 2.7 for a 21 wt.% Ni steel with two different grain sizes less than 1 μm and stable austenite and a 6 wt.% Mn steel also with a grain size less than 1 μm with austenite that undergoes strain-induced transformation to martensite [53]. A type I flow curve occurs when the grain size is extremely fine (on the order of 0.6 μm or smaller) and the specimen necks and fractures during Lüder’s deformation. A type II flow curve occurs for larger grain sizes and the specimen work hardens following Lüder’s deformation, which leads to higher total elongation. For type I and type II flow curves, the austenite is highly stable against a strain-induced transformation to martensite. A type III flow curve occurs when a significant amount of austenite undergoes a strain-induced transformation to martensite, which causes a high rate of work hardening.

In medium Mn TRIP steels, YPE and YS decrease with increasing IAT as demonstrated by the stress-strain curves in Figure 2.6a. At higher IAT, the formation of athermal martensite during cooling leads to high initial dislocation density and continuous yielding. De Cooman et al. used transmission electron microscopy (TEM) to examine the as heat treated and post-yielding microstructures of the 600 and 650 °C conditions of the 7 wt pct
Figure 2.7  Different types of stress-strain curves that can occur for ultra fine grain austenite-ferrite microstructures in steel. The ”Type I” and ”Type II” curves come from a 21 wt.% Ni steel, and the ”Type III” curve comes from a 6 wt.% Mn steel. Figure from Miller [53].

Mn TRIP steel in Figure 2.6 [62]. The TEM micrographs for the 600 °C condition are shown in Figure 2.8; the initial microstructure and the tensile stress-strain behavior of this condition are similar to the 7 wt.% Mn TRIP steel studied in this work. The 600 °C annealed microstructure had ferrite grains with very low initial dislocation density and carbide precipitates and austenite grains with some stacking faults and a small amount of annealing twins, shown in Figures 2.8a and 2.8b. Following the Lüder’s deformation to 6.8% strain, the ferrite had a high dislocation density and the austenite had a high density of planar stacking faults, shown in Figures 2.8c and 2.8d. After straining to 10%, the austenite contained a high density of thin, parallel $\varepsilon$ martensite plates and $\alpha'$ martensite that nucleated at $\varepsilon$ martensite, shown in Figure 2.8e. These observations indicate that both ferrite and austenite deformed by dislocation glide during the Lüder’s deformation, and beyond the YPE, strain-induced transformation of austenite to $\varepsilon$ martensite and then to $\alpha'$ martensite contribute to plastic deformation. In contrast, the 650 °C annealed microstructure contained ferrite with high initial dislocation density and austenite with stacking faults that was partially transformed to $\varepsilon$ martensite and $\alpha'$ martensite. Yielding was associated with a rapid decrease in the austenite fraction, lack of YPE, and an increase in stacking faults and $\varepsilon$ martensite in austenite at 2% strain. This behavior suggests that stress-assisted transformation of the austenite may have contributed to yielding. The retained austenite in the 650 °C sample may have had a range
of $M_\sigma$ temperatures above and below room temperature, because approximately 40% of the retained austenite transformed around yielding (possibly stress-assisted transformation) and the remainder transformed during plastic deformation (strain-induced transformation).

Figure 2.8 TEM micrographs of a 7.1Mn-0.1C-0.12Si (wt.%) TRIP steel cold reduced 50 – 60 % and then annealed for 196 hr at 600 °C. (a) as-annealed ferrite grain with a low angle grain boundary indicated by the arrow and a carbide indicated by the circle, (b) as-annealed austenite grain with stacking faults, (c) ferrite grain at the end of YPE (6.8% strain) showing a high dislocation density, (d) austenite grain at the end of YPE (6.8% strain) showing a high density of planar stacking faults, and (e) austenite grain at 10% strain showing a high density of thin $\epsilon$ martensite plates and $\alpha'$ martensite nucleated at the $\epsilon$ martensite. Figures from De Cooman et al. [62].
The serrated plastic flow following discontinuous yielding for the 600 °C and 625 °C IAT conditions in Figure 2.6a has been observed for other TRIP steels containing 4.5 – 10 wt.% Mn and seems to be associated with an intermediate IAT range that produces mostly ferrite-austenite ultra fine grained microstructures with a large amount of retained austenite (greater than approximately 10 vol.%) that undergoes strain-induced transformation to martensite [50, 51, 55, 66, 67]. There are four possible origins for serrated flow identified by Rodriguez that may be relevant to medium Mn TRIP steels tested at room temperature [68]:

- A sudden increase in the plastic strain rate of the specimen, $\dot{\epsilon}_p$, will result in a stress drop in order to maintain the imposed strain rate, $\dot{\epsilon}$. The $\dot{\epsilon}_p$ due to dislocation glide is given by Equation 2.3, in which $\rho_m$ is the mobile dislocation density, $b$ is the Burgers vector, and $\bar{v}$ is the average dislocation velocity [64]. A sudden change in $\rho_m$ or $\bar{v}$ would cause a sudden change in $\dot{\epsilon}_p$. For example, rapid dislocation generation leads to a rapid increase in $\rho_m$. The imposed strain rate necessitates a decrease in $\bar{v}$ such that $\dot{\epsilon}_p$ will match the imposed $\dot{\epsilon}$, and the decrease in $\bar{v}$ leads to a drop in stress [45, 64].

$$\dot{\epsilon}_p = \rho_m b \bar{v} \tag{2.3}$$

- The pinning and unpinning of mobile dislocations by solute atoms during plastic deformation, which is known as dynamic strain aging (DSA). DSA involves sudden changes in $\rho_m$, which in turn causes sudden changes in $\dot{\epsilon}_p$.

- Deformation by mechanical twinning.

- A stress or strain induced phase transformation.

The serrations are most likely not due to mechanical twinning (as a major deformation mechanism, but twins may occur as shear band intersections where $\alpha'$ martensite can nucleate), since it is well documented that the austenite undergoes transformation to martensite in medium Mn TRIP steels [51, 53–55, 62, 67]. Specifically, no deformation twins were observed by TEM for the 600 °C condition of the 7 wt.% Mn TRIP steel that exhibits serrated plastic flow in Figure 2.6a [62]. The occurrence of twinning as a deformation mechanism in austenite is dependent on SFE and tends to be observed in steels containing 15 – 25 wt.% Mn, 2 – 4 wt.% Al and Si, and less than 1 wt.% C [69, 70]. Both twinning and austenite transformation to martensite were observed for a 10Mn-0.3C-3Al-2Si (wt.% ) alloy [52], which has much higher Mn, C, Al and Si than the TRIP7Mn studied in this work. Even in TWIP steels, serrations in the flow curve are often not attributed to mechanical twinning but instead to a form of DSA, and the serrations readily change appearance and disappear.
with changes in temperature or strain rate while the twinning mechanism is still operative [71–80].

There are some cases for steel in which flow curve serrations have been attributed to the austenite to martensite transformation. Richman and Bolling showed that at temperatures between \( M_s \) and \( M_x \), the yielding behavior of Fe-Ni-C austenitic steel was associated with a low yield point and serrated flow behavior at yielding due to the stress-assisted austenite to martensite transformation [81]. Further, they showed that for a 15.9Ni-0.96C (wt.%) steel serrated flow behavior occurred at two different temperature regimes above \( M_x \). The serrations in the higher temperature regime were attributed to DSA while the serrations in the lower temperature regime were attributed to the strain-induced austenite to martensite transformation. It is critical to note that the serrations due to DSA were observed at a strain rate of \( 10^{-4} \), but the serrations due to the strain-induced austenite to martensite transformation were only observed at a strain rate of \( 10^{-5} \) and disappeared at a strain rate of \( 10^{-4} \), which is the lowest strain rate at which most tensile tests are performed. Therefore, the serrations observed at strain rates of \( 10^{-4} \) and higher for Fe-Mn-C steels may be attributed to a cause other than the strain-induced austenite to martensite transformation. Cai et al. suggest that strain-induced austenite to martensite transformation is responsible for serrated flow curves of austenite-ferrite microstructures in Fe-Mn-Al-C steels [82]. They propose that serrated flow can occur when the microstructure consists of a soft ferrite matrix and austenite with varying degrees of stability that transforms in a discontinuous, discrete manner. They suggest that the transformation of a discrete quantity of austenite followed by work hardening of the surrounding ferrite throughout the gage length is responsible for the large, step-like serrations, but there is no discussion of the fine serrations that can occur within the larger, step-like serrations. Their proposed mechanism for the serrated plastic flow does not address the discrete deformation bands associated with serrated plastic flow in Fe-C-Mn TRIP and TWIP steels, and it does not account for the sensitivity of the serrations to temperature and strain rate.

The origin of serrations in Fe-C-Mn steels has been attributed in several studies to a form of DSA that can occur at room temperature and low strain rates [71–80, 83–85]. In most cases, DSA arises due to solute atoms that are able to diffuse to and pin dislocations when the dislocations are temporarily held up at local obstacles, which is referred to as the waiting time for a dislocation [86]. At room temperature, the bulk diffusivity of solute atoms in steels is low, and DSA in steels is typically observed at elevated temperature for a quasi-static strain rate, such as 100 – 200 °C for AISI 1020 steel and 250 – 650 °C for 316 stainless steel [24, 68]. Dastur and Leslie proposed that the reorientation of C in C-Mn couples at the core
of dislocations could be responsible for DSA in austenitic Hadfield steel (1.0 – 1.4 wt.% C and 10 – 14 wt.% Mn) [83]. The Hadfield steel in their study (1.13C-11.4Mn-0.2 Si-0.17Ni-0.16Cr-0.08Mo wt.%) exhibited twinning with a decrease in twin density with increasing temperature and did not exhibit a strain-induced transformation to α or ε martensite. Serrations in the flow curve were observed in the range of -25 to 300 °C, and they were attributed to DSA rather than mechanical twinning because twinning was prevalent below -25 °C in the absence of serrations and very few mechanical twins were observed at 225 °C although high work hardening and serrations were present. Dastur and Leslie calculated that the activation energy for the onset of serrations was 0.6 – 0.7 that of the activation energy for bulk diffusion of C, but the activation energy for serrated flow agreed with the activation energy for short-range diffusion of C in dislocation cores. Thermodynamic and internal friction data indicate the presence of C-Mn point defect pairs in Fe-Mn-C alloys [87, 88]. It is highly probable for C and Mn to exist as couples in austenite, because Mn provides lower free energy interstitial sites for C [76]. Dastur and Leslie suggested that the C of C-Mn couples can reorient and pin dislocations when the lattice is strained, which leads to an increase in dislocation density and an increase in work hardening [83]. In a recent study by Lee et al., serrated flow observed at room temperature in Fe-Mn-C alloys that exhibit TWIP behavior was attributed to a DSA mechanism involving the interaction of C-Mn point defect complexes with stacking faults [76]. In the stacking fault region, the packing is HCP, and the C atoms that occupy the octahedral interstitial sites in the FCC lattice are transferred to tetrahedral interstitial sites after the passage of a partial dislocation. The large lattice distortion associated with C at a tetrahedral site makes it favorable for the C atoms to make a single diffusive jump to a nearby octahedral site, which has a much lower activation energy than bulk C diffusion. The reorientation of the C in the C-Mn couple to an energetically favorable orientation pins the stacking fault and partial dislocations and requires an increase in stress for dislocations to resume glide. A low SFE with a large distance between partial dislocations allows more time for the reorientation of the C-Mn couple within the stacking fault region before the trailing partial dislocation removes the stacking fault; therefore, low SFE promotes DSA. The reorientation of C-Mn couples in stacking fault regions is a possible mechanism occurring in the austenite of medium Mn TRIP steels.

Discontinuous serrated plastic flow due to DSA is characterized by the formation and propagation of localized Portevin-Le Châtelier (PLC) deformation bands and occurs at specific ranges of temperatures, strain rates, and strain [86]. The appearance of serrations during discontinuous plastic flow can be categorized by type as shown in the schematic in Figure 2.9 [68]. Type A serrations are characterized by an abrupt rise and then drop in load followed by a long, relatively flat portion of the flow curve during which a deformation band
travels along the gage length. Repeated initiation and propagation of deformation bands leads to periodic serrations in the stress-strain curve, and these type of serrations tend to occur at low temperature or high strain rates. Type B serrations appear as fine oscillations that occur in rapid succession and arise due to discontinuous plastic deformation from DSA within a deformation band. These serrations can occur at high temperatures and low strain rates and can also occur in conjunction with type A serrations. Type C serrations are load drops below the overall level of the flow curve due to the unlocking of dislocations, and they occur at higher temperatures and lower strain rates than A and B type serrations. Type D serrations are due to the propagation of deformation bands that appear as plateaus with a lack of work hardening in the flow curve, similar to the appearance of Lüder’s band propagation, and they can occur with or without type B serrations. Type E serrations follow type A serrations and occur at high strain levels with little or no work hardening.

Figure 2.9 Schematic showing various types of DSA serrations originally from Rodriguez [68] and re-drawn by Robinson and Shaw [86].

There is a critical strain level, $\epsilon_c$, for the onset of serrated plastic flow, and it is affected by temperature and strain rate [68]. The value of $\epsilon_c$ is determined by the solute diffusivity, the mobile dislocation density as a function of strain, and the factors affecting the ability of solute atoms to pin dislocations as a function of strain, e.g. the density of forest dislocations and the vacancy concentration [89]. For the case of C-Mn complexes pinning partial dislocations in austenite, $\epsilon_c$ depends on the C-Mn reorientation time, which is very short, and the interaction time of C atoms with stacking faults, which depends on the stacking fault width, density of C-Mn complexes, dislocation glide velocity, and stacking fault density [76].
Examples of the types of serrations at different temperatures, the effects of temperature and strain rate on flow behavior, and a strain rate-temperature map showing the occurrence of serrations for DSA in austenitic Hadfield steel (1.0 – 1.4 wt.% C and 10 – 14 wt.% Mn), discussed previously, are shown in Figure 2.10 [83]. Figure 2.10a shows the change in serration type from A to B to C with increasing temperature. The $\epsilon_c$ decreases and then increases with increasing temperature. As shown by Figure 2.10b, negative strain rate sensitivity occurs in the temperature range for which serrated plastic flow occurs, and there is positive strain rate sensitivity at temperatures beyond the DSA regime. The tensile properties as a function of temperature are shown in Figure 2.10c. The yield strength gradually declines with increasing temperature, while the tensile strength increases with increasing temperature until a peak is reached in the DSA regime, and then tensile strength declines rapidly beyond the DSA regime. The trend in elongation is similar to tensile strength with an increase and peak in elongation in the DSA regime. Uniform and total elongation are equal up to 300 °C. At 300 °C and above, elongation declines but reduction in area increases significantly. The change in elongation and reduction in area behavior at 300 °C occurs at the end of the DSA regime and is associated with a change in fracture behavior from intergranular fracture with voids at grain boundaries to a dimpled fracture surface. The peak in uniform elongation for DSA associated with C-Mn couples differs from traditional DSA in which uniform elongation decreases in the DSA regime. Dastur and Leslie suggest that the high uniform elongation may be due to the high concentration of C and Mn in solution necessary to produce DSA. The concentration of solute atoms is high enough that dislocations can be effectively pinned even as the dislocation density increases with increasing strain and necking is delayed. The strain rate-temperature map in Figure 2.10d shows that serrated plastic flow due to DSA occurs at room temperature for Hadfield steel. Room temperature is near the lower temperature boundary for serrated flow. For a constant temperature along the lower boundary, an increase in strain rate can cause a transition from serrations in the flow curve to no serrations. The transition between serrations to no serrations for room temperature tests of Hadfield steel likely occurs just above 0.01 s$^{-1}$.

2.4 Strain-Induced Austenite to Martensite Transformation

The Gibbs free energies for austenite and martensite as a function of temperature are illustrated in Figure 2.11a [90]. The terms \(M_s\), \(M^\sigma_s\), and \(M_d\) are temperatures that correspond to major changes in the chemical and mechanical driving forces necessary for the austenite to martensite transformation. Below \(M_s\), austenite transforms spontaneously to martensite, because the difference in the chemical free energy, \(\Delta G^{\gamma\rightarrow\alpha'}\), is high enough to provide a driving
Figure 2.10 Mechanical behavior data for an austenitic Hadfield steel (1.13C-11.4Mn-0.2Si-0.17Ni-0.16Cr-0.08Mo wt.%). (a) Serration types observed at various temperatures for tests at $3 \cdot 10^{-4}$ s$^{-1}$. (b) Variation of strain rate sensitivity with temperature. (c) Tensile mechanical properties as a function of temperature. (d) Strain rate-temperature map showing the regimes for serrated plastic flow. Figures from Dastur and Leslie [83].
force for martensite to form at pre-existing nucleation sites in the austenite [91]. As temperature increases above $M_s$, the chemical driving force for the transformation decreases and an additional mechanical driving force, $U'$, from an applied stress is necessary for martensite nucleation [91]. The dashed line below the solid lines for the Gibbs free energies of austenite and martensite in Figure 2.11a represents the additional mechanical driving force necessary for the transformation to occur. Figure 2.11b is a schematic developed by Olson and Cohen to illustrate the relationship between the critical stress to initiate $\alpha$ martensite nucleation and temperature [91]. In the temperature range between $M_s$ and $M_s^\sigma$ the transformation is initiated by elastic stress in the austenite and is termed stress-assisted. In this temperature regime, yielding occurs below the yield stress of austenite, $\sigma_\gamma$, due to the initiation of the martensite transformation [81]. Above $M_s^\sigma$, yielding occurs by slip in the austenite when $\sigma_\gamma$ is reached, and the applied stress must be equal to or greater than $\sigma_\gamma$ to initiate the transformation to martensite [81]. The strain-induced transformation occurs in the temperature range between $M_s^\sigma$ and $M_d$. In this temperature regime, plastic deformation produces potent nucleation sites at shear band intersections for martensite nucleation [91]. These shear band intersections can be formed by the intersection of $\epsilon$ martensite plates, $\epsilon$ martensite plates and twins, $\epsilon$ martensite plates and active slip systems, twins, dense stacking fault bundles, or twins and stacking fault bundles [92–96]. Above $M_d$, austenite is stable against plastic deformation-induced transformation to $\alpha$ martensite [91].

![Gibbs free energies of austenite and martensite](image1)

![Schematic illustrating the critical stress required for the initiation of the austenite martensite transformation](image2)

Figure 2.11  (a) Gibbs free energies of austenite and martensite as a function of temperature. Figure from Curtze et al. [90]. (b) Schematic illustrating the critical stress required for the initiation of the austenite martensite transformation. Schematic redrawn from Olson and Cohen [91].
Numerous models have been developed to describe the kinetics of the strain-induced austenite to martensite transformation, including the Angel-Ludwigson-Burger, Olson-Cohen, Burke-Matsumura et al., Sugimoto et al., Sherif et al., Mukherjee et al., and Stringfellow et al. models [97–105]. These models range in complexity from the Sugimoto et al. model that is a simple equation with only one constant related to the stability of austenite to the Stringfellow et al. model that involves multiple equations and numerous variables to account for the effects of temperature, strain, and stress state [102, 105]. Mukherjee et al. applied the Olson-Cohen, Burke-Matsumura et al., and Mukherjee et al. models to seventeen TRIP steel tensile test data sets from eight different studies and found an $R^2$ value above 0.9 for all three models with the majority of data set and model combinations resulting in an $R^2$ equal to or greater than 0.97 [106]. For this thesis, the Olson-Cohen mechanism-based model will be considered. This model has constants that can be obtained just from experimental strain and retained austenite measurements, and provides good fits to experimental data for TRIP and Q&P steels [43, 48, 99, 106, 107].

The Olson-Cohen model was developed based on the concept that shear band intersections are the dominant nucleation sites for martensite [99]. The model is expressed by Equation 2.4, where $f_{\alpha'}$ is the volume fraction of transformed martensite, $\alpha$ is a parameter that reflects the rate of shear band formation, $\beta$ is a term related to the probability of forming martensite nuclei at a shear band intersection, and $n$ is a fixed exponent term that relates to the number of intersections in the austenite. The parameter $\alpha$ is dependent on stacking fault energy and increases with decreasing stacking fault energy, because low stacking fault energy promotes the shear band deformation mode. Olson and Cohen also suggest that increasing strain rate should also promote the shear band deformation mode and would likely increase $\alpha$. The parameter $\beta$ is related to the chemical driving force for the austenite to martensite transformation, since the chemical driving force controls martensite nucleation. A value of 2 is the lower bound for the parameter $n$ based on the assumption that austenite grains are equiaxed and that shear bands are randomly oriented with a length equal to the austenite grain size. The parameter $n$ can have a value higher than 2, which would reflect a condition in which the number of shear band intersections is initially low due to non-random orientations of the shear bands and the number of intersections would increase more rapidly with increasing strain.

$$f_{\alpha'} = 1 - \exp\left(-\beta(1 - \exp(-\alpha \epsilon))^n\right)$$

Both $\alpha$ and $\beta$ depend on temperature. Stacking fault energy increases with increasing temperature, which leads to a decrease in $\alpha$ [99, 108]. As temperature increases, the chemical
driving force for martensite nucleation is reduced, which in turn increases the stress required for the transformation to occur [91]. The reduction in chemical driving force for the austenite to martensite transformation leads to a decrease in $\beta$ with increasing temperature [99]. Therefore, increasing temperature increases the stability of the austenite and reduces the transformation rate to martensite. The effect of temperature on the austenite to martensite transformation rate in TRIP steel is demonstrated by the plot in Figure 2.12a of martensite fraction as a function of true strain from tensile tests of 0.2C-1.7Mn-1.5Al (wt.%) TRIP steel at temperatures ranging from 7 to 100 $^\circ$C [107]. The data points are experimental measurements and the curves were fit to the data from the Olson-Cohen model. The $\alpha$ and $\beta$ parameters as a function of temperature are shown in Figure 2.12b.

Figure 2.12  (a) Experimental data and fitted curves from the Olson-Cohen model for the volume fraction of transformed martensite as a function of strain for a 0.2C-1.7Mn-1.5Al (wt.%) TRIP steel with a 1 % temper roll at various temperatures. The experimental data came from interrupted tensile tests at a strain rate of $10^{-3}$ s$^{-1}$. The fits were done with $n = 2$. (b) The Olson-Cohen model parameters $\alpha$ and $\beta$ as a function of temperature. Figures from Samek et al. [107].

### 2.4.1 Factors Affecting the Stability of Retained Austenite

Austenite stability is affected by size, morphology and chemical composition of the austenite as well as temperature and the other microconstituents present in the microstructure. Some of these factors affect the SFE and/or $M_s$ temperature of the austenite. Austenite stability increases with increasing SFE or decreasing $M_s$ temperature. A lower $M_s$ tem-
perature means the chemical driving force for the strain-induced austenite to martensite transformation is lower (at a constant test temperature) [91, 109]. As SFE increases, the distance between the partial dislocations decreases and cross-slip of screw dislocations is easier, which means deformation by dislocation glide is easier [45]. Therefore, increasing SFE leads to less planar slip and the formation of fewer shear band intersections, which are necessary as martensite nucleation sites for the strain-induced austenite to martensite transformation.

As discussed in Section 2.4, temperature has a potent effect on austenite stability. As the test temperature increases, the chemical driving force for the austenite to martensite transformation is reduced and a larger mechanical driving force is necessary for the transformation to occur [91, 109]. Further, increasing temperature increases SFE [108]. A thermal contribution aids the constriction of an extended screw dislocation allowing it to cross-slip, which means that cross-slip becomes easier with increasing temperature [45]. As shown by the data in Figure 2.12a, small increases in temperature on the order of 5 – 10 °C can have an observable stabilizing effect on austenite in TRIP steel [107].

Steel chemistry and processing determine the amount of austenite that can be retained at room temperature, and the chemical composition of the retained austenite affects its stability [20, 33]. Specific alloying concepts for Q&P and medium Mn TRIP are discussed in more detail in Sections 2.2 and 2.3. For both steel types, the major alloying elements for obtaining retained austenite in the final microstructure are C, Mn, Si and/or Al. Carbon and Mn are austenite stabilizers, while Si and/or Al are necessary to suppress carbide formation and allow C to remain in solution [20, 24]. Carbon, Mn, and Si decrease the $M_s$ temperature with C having an order of magnitude greater effect than other alloying elements, and Al increases the $M_s$ temperature [24, 29, 31]. The SFE increases linearly with increasing C and Al [70]. With increasing Mn content, SFE initially decreases until a minimum in SFE is reached in the range of 10 – 20 wt.% Mn and then SFE increases with increasing Mn [70]. Silicon decreases SFE, but it has a more minor effect on SFE than C, Mn or Al [20, 70, 110]. During processing, the partitioning of alloying elements to austenite is not necessarily completely uniform across all of the austenite grains. A recent in situ study of austenite grains during deformation with EBSD and electron probe microanalysis (EPMA) suggests that composition differences between individual austenite grains may be partly responsible for the gradual transformation of austenite with increasing strain, which is beneficial for delaying necking [111].

Several studies have demonstrated that the $M_s$ temperature can be depressed by a fine austenite grain size formed during austenitizing [112–119]. For example, Yang and Bhadeshia showed with dilatometry experiments that the $M_s$ temperature of a 0.13C-5Ni-
2.27Mn (wt.%) steel decreased with decreasing austenite grain size, and they developed a geometry-based model to predict the reduction in $M_s$ temperature due to austenite grain size [119]. The model is based on the idea proposed by Fisher et al. that the amount of martensite transformation at a given temperature is proportional to the austenite mean grain diameter cubed [112]. The model shows that the $M_s$ temperature decreases more rapidly for finer grain sizes (below approximately 10 $\mu$m) than larger grain sizes, and the predicted decrease in $M_s$ temperature for a 1 $\mu$m grain size of the steel used for their study is 71 $^\circ$C. Recently, Jimenez-Melero et al. demonstrated the relationship between $M_s$ temperature and retained austenite grain size with synchrotron X-ray diffraction experiments performed on a 0.2C-1.5Mn-0.3Si-1.8Al (wt.%) TRIP steel with a ferrite, bainite, and metastable austenite microstructure [120]. They tracked seventy individual austenite grains while cooling the steel to 100 K and found a reduction in the $M_s$ temperature by approximately 100 K when the austenite grain volume decreased from 20 $\mu$m$^3$ to 10 $\mu$m$^3$. Carbon content also had an effect on $M_s$ temperature, since finer grains tended to have higher C contents although spread in the C content was high. The large amount of spread in the C content and grain size data was attributed to other austenite stabilization factors such as morphology, dislocation density and/or crystallographic orientation. Several studies have shown that the presence of retained austenite at room temperature in Mn-C steels is partly dependent on an ultrafine austenite grain size [53, 58, 121]. Other studies of TRIP steels have found that the finer the austenite grain size the more stable the austenite grains were to a strain-induced transformation to martensite [122, 123]. During plastic deformation, the larger austenite grains tended to transform at lower strains while the finer austenite grains were present at higher strains. The increase in austenite stability with decreasing grain size is attributed to fewer potential martensite nucleation sites in smaller austenite grains [124].

Retained austenite can be present as equiaxed grains or as thin plates or films. Film or plate-like austenite usually occurs within bainite or martensite and equiaxed austenite usually occurs in the presence of equiaxed ferrite [20, 53, 125, 126]. Bhadeshia and Edmonds found that larger quantities of austenite could be retained at room temperature when the austenite had a film morphology as opposed to a blocky morphology in bainitic steels, and the film austenite was more resistant to transformation under an applied stress than the blocky austenite [125]. Sugimoto et al. and Chiang et al. showed that the presence of austenite as lamellar grains, plates, or films led to greater austenite stability versus an equiaxed morphology in TRIP steel that was processed to obtain different austenite morphologies, although it is challenging to completely isolate the effect of morphology from the effects of austenite chemistry and the neighboring microconstituents [126, 127]. For a TRIP steel containing ferrite, bainite, and blocky and film austenite, Timokhina et al. found that blocky austenite at
polygonal ferrite boundaries and at polygonal ferrite/bainite boundaries transformed at low strains whereas film austenite between bainite laths was very stable against transformation up to 20% strain [128]. Recently, Xiong et al. studied a Q&P steel containing both blocky and film austenite, and they found that the high C blocky austenite transformed rapidly at the onset of plastic deformation whereas untransformed low C film austenite could be observed up to 12% strain [129].

The relative strength of the microconstituents adjacent to the retained austenite can affect the austenite to martensite transformation due to stress partitioning among the microconstituents. In a study of first generation TRIP steels containing austenite in similar quantities, grain sizes, and carbon contents, Jacques et al. showed that the austenite to martensite transformation rate was reduced in the microstructures containing athermal martensite versus the microstructures containing only ferrite, bainite, and retained austenite [130]. Following an isostrain composite model, the athermal martensite strengthens the microstructural matrix leading to a lower mean stress in the austenite, which postpones the austenite to martensite transformation. A similar stress shielding effect was observed for solid solution strengthening of the ferrite by Si [130]. The study by Jacques et al. compared steels that had different matrix hardnewes, but the concept of stress shielding may also apply to hardness differences between individual microconstituents. The presence of film austenite adjacent to hard bainite as opposed to the presence of equiaxed austenite adjacent to softer equiaxed ferrite was suggested as another reason besides morphology for the observed higher stability of film austenite than blocky austenite in the TRIP steel studied by Chiang et al. [126]. However, in a study of TRIP steel containing ferrite, bainite, and austenite, Zhang and Findley showed that while the hardness of bainite is initially higher than that of ferrite, work hardening of both constituents led to a negligible difference in hardness beyond 4% strain [131]. For a steel with multiple microconstituents besides austenite, stress shielding may only have a significant impact on the austenite to martensite transformation rate at low strains when the hardness disparity between microconstituents is the greatest. This idea is supported by a study by Zhao et al. on a Q&P steel containing ferrite, martensite and austenite in which the austenite to martensite transformation rate below 2% strain was altered by tempering the steel, which reduced the hardness difference between the ferrite and martensite [132].

2.5 High Strain Rate Tensile Deformation

The strain rates that will be examined in this study are in the range of $0.0001 - 200 \text{ s}^{-1}$. The range of $1 - 1000 \text{ s}^{-1}$ is considered a low dynamic regime in which mechanical resonance
of the specimen and the testing machine can affect the test results [133]. The propagation of elastic waves during the test at high strain rates manifests as oscillations in the load data [134]. A discussion of the load oscillations and how they are minimized and addressed in the data is given in Section 3.4.4.

The literature review on high strain rate tensile deformation behavior will be divided into several sub-sections. First, the basic deformation mechanisms of BCC and FCC metals that contribute to the observed mechanical behavior at various strain rates are reviewed. The adiabatic heating that occurs at high strain rates will be discussed in terms of fundamental heat transfer equations and a model developed to predict adiabatic heating of sheet-type tensile specimens. The effect of strain rate on various AHSS including first generation DP and TRIP steels will be presented. Finally, studies on the effect of strain rate on the austenite-to-martensite transformation in austenitic stainless steels and TRIP steels will be discussed.

2.5.1 Deformation Mechanisms

Strain rate sensitivity refers to how flow stress at a specific level of strain responds to a change in strain rate, and for most metals, flow stress increases with increasing strain rate (i.e. positive strain rate sensitivity) [64]. Increasing strain rate has a similar effect on flow stress as decreasing temperature, because there are thermal and athermal components to flow stress. The thermal component arises from short-range obstacles to dislocation movement that can be overcome through thermal assistance [45]. The athermal component arises from long-range obstacles that cannot be overcome by thermal energy [45]. The primary short-range barrier for body centered cubic (BCC) metals is the Peierls Nabarro stress, and it is dislocation forests for face centered cubic (FCC) metals. The difference in the primary short range barriers contributes to a difference in strain rate sensitivity between BCC and FCC metals. Solute atoms can also be short-range barriers. Long range barriers include other dislocations, grain boundaries, dislocation substructure, and precipitates. Short range barriers are easier for a dislocation to overcome with increasing temperature and time. Additional thermal energy added to a material helps dislocations overcome short-range energy barriers, which allows the material to deform at a lower flow stress [133]. Likewise, with increasing strain rate, dislocation segments have less time to overcome short-range obstacles with thermal assistance and flow stress increases. Therefore, strong, positive strain rate sensitivity is expected at high strain rates when thermally and time-dependent short range barriers make a more substantial contribution to flow stress. When long range dislocation interactions, which are not thermally-dependent, dominate the flow stress, it is expected
that there will be little or no observed strain rate sensitivity. Negative strain rate sensitivity can occur in steels due to the effects of strain rate on DSA and the austenite to martensite transformation. Dynamic strain aging occurs in specific temperature and strain rate ranges [68]. As DSA diminishes and eventually disappears with increasing strain rate (at a constant test temperature), flow stress decreases and strain rate sensitivity is negative [135–137]. For metastable austenitic stainless steels, the specimen self-heating that occurs with increasing strain rate increases the austenite stability and thereby, reduces the austenite to martensite transformation rate and leads to negative strain rate sensitivity [8–13]. The negative strain rate sensitivity observed for metastable austenitic stainless steels is discussed in more detail in Section 2.5.4.

Campbell and Ferguson performed experiments to determine the effect of temperature and strain rate on the lower yield strength of annealed mild steel [138]. Their results are summarized in the plot in Figure 2.13. They identified three key regions for the behavior of the steel:

- **Region I** is the low to medium strain rate ($10^{-3} - 2 \cdot 10^1 \text{ s}^{-1}$) and low to intermediate temperature regime (approximately 293 – 713 K). There is little or no strain rate sensitivity due to athermal contributions outweighing thermally-dependent contributions to flow stress. In other words, dislocation movement is controlled by long-range interactions.

- **Region II** is the low to medium-high strain rate ($10^{-3} - 5 \cdot 10^3 \text{ s}^{-1}$) and low temperature regime (approximately 195 – 373 K). There is positive strain rate sensitivity, and flow stress behaves according to a semi-logarithmic relationship with strain rate. Temperature also has a greater effect on flow stress in region II than region I. Time and thermal contributions have a strong effect on flow stress in region II meaning that short-range interactions are strong inhibitors of dislocation movement.

- **Region IV** incorporates very high strain rates ($5 \cdot 10^3 \text{ s}^{-1} - 10^5 \text{ s}^{-1}$) and low to intermediate temperatures (approximately 293 – 713 K). Flow stress rapidly increases with strain rate, which is attributed to lattice viscosity controlling dislocation velocity.

The increase in the strain rate sensitivity between regions I and II of Figure 2.13 comes from the transition between athermal deformation mechanisms controlling the flow stress in region I to thermally-dependent deformation mechanisms controlling the flow stress in region II. The increase in the strain rate sensitivity parameter for yield strength was found to occur at approximately 10 s$^{-1}$ for IF steels, 50 s$^{-1}$ for austenitic stainless steels, between
1 and 100 s$^{-1}$ for annealed ferritic stainless steels, between 1 and 10 s$^{-1}$ for DP and TRIP steels, and close to 10 s$^{-1}$ for a plain carbon steel with a tempered martensite microstructure [9, 134, 139–141]. There are steels and conditions for which no distinct increase in strain rate sensitivity is observed either due to strong positive strain rate sensitivity or a lack of strain rate sensitivity. Steels that are mostly or completely comprised of ferrite, for example HSLA and 18 Cr-Cb™ ferritic stainless steel, can exhibit strong positive strain rate sensitivity that increases very gradually at higher strain rates in the range of 0.001 – 1000 s$^{-1}$ [134, 139]. In these steels, the contribution of long-range dislocation interactions to flow stress is small compared to the contribution of short-range dislocation interactions, which leads to strong positive strain rate sensitivity across all strain rates. Plain carbon steels with martensitic microstructures in the as-quenched or low tempering temperature ($<$ 500 °C) conditions exhibit little or no strain rate sensitivity up to 100 s$^{-1}$, which is attributed to dominant athermal contributions to flow stress particularly from high dislocation density [141].

There are different ways to quantify strain rate sensitivity. As suggested by the linear fits in Figure 2.13, there is often a semi-logarithmic relationship between flow stress and strain rate for steel. A semi-logarithmic strain rate sensitivity parameter $\beta$ can be calculated for constant true strain and temperature from Equation 2.5 where $\sigma$ is true flow stress, $\dot{\varepsilon}$ is true
strain rate and $k$ is a constant [64, 134].

$$\sigma = \beta \log(\dot{\varepsilon}) + k|_{\varepsilon,T} \tag{2.5}$$

A more widely used expression for the relationship between true flow stress and true strain rate is given by Equation 2.6 where $m$ is the strain rate sensitivity parameter and $C$ is a constant [64]. The strain rate sensitivity parameter $m$ can be calculated as the slope of a log $\sigma$ vs log $\dot{\varepsilon}$ plot with data from multiple tensile tests or $m$ can be calculated from a change in strain rate during a tensile test ($m = \log(\sigma_2/\sigma_1)/\log(\dot{\varepsilon}_2/\dot{\varepsilon}_1)$) [64].

$$\sigma = C (\dot{\varepsilon})^m|_{\varepsilon,T} \tag{2.6}$$

Strain rate sensitivity can also be examined by comparing dynamic-static strength ratios (flow stress at a high strain rate divided by flow stress at a low strain rate) for different steels [140, 142]. A dynamic-static strength ratio above one indicates positive strain rate sensitivity, and the higher the ratio above one the greater the strain rate sensitivity. Dynamic-static strength ratios are easily calculated and provide a general idea of a steel’s strain rate sensitivity, but they do not show subtle changes in strain rate sensitivity with increasing strain rate. For example, a dynamic-static strength ratio calculated from the room temperature YS of mild steel at 0.001 and 100 s$^{-1}$ from the data in Figure 2.13 would indicate a strong, positive strain rate sensitivity, but it would not show any indication of the transition from low strain rate sensitivity to high strain rate sensitivity at 0.1 s$^{-1}$.

Larour et al. conducted an extensive literature review of strain rate studies on first generation AHSS and found several overarching relationships between the strain rate sensitivity parameter, $m$, and characteristics of the steels [143]. The $m$ value typically decreases with increasing plastic strain with the sharpest decline following yielding and then an almost constant $m$ close to the UTS. The decline in $m$ is most dramatic for low strength steels. For example, $m$ declines from 0.059 at 0.002 true plastic strain to 0.022 at 0.010 true plastic strain for mild steel in the strain rate range of 0.001 – 1 s$^{-1}$; whereas, a high strength multiphase steel with a minimum UTS of 950 MPa has a nearly constant $m$ below 0.005 for true plastic strains up to 0.07 over the same strain rate range. Studies on ferritic steel grades, such as IF, HSLA, and mild steel, show that as the alloying content increases, the values of $m$ as a function of total amount of alloying elements follow a power law decrease. An increase in the total amount of alloying elements from approximately 0.3 wt.% to 1.0 wt.% causes a decline in $m$ for yield strength from 0.09 to 0.02. Figure 2.14a shows $m$ versus fraction of hard phase for multiphase first generation AHSS. As the volume percent
Figure 2.14  (a) The strain rate sensitivity parameter \( m \) at 296 K in the strain rate range 0.001 – 20 s\(^{-1}\) versus vol.% of hard phase (martensite and/or bainite) for high strength multiphase first generation AHSS. (b) Values for \( m \) at 296 K in the strain rate range 0.001 – 250 s\(^{-1}\) versus quasi-static yield strength \( (R_{\text{p0.2}}) \). The \( m \) values come from literature sources for various AHSS grades including IF, HSIF, HSLA, BH, DP, TRIP, CP, and PHS. Figures from Larour et al. [143].

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of hard phase increases, the already relatively low $m$ for yield strength decreases linearly. These results suggest that strengthening mechanisms such as work hardening, solid solution strengthening, and second phase strengthening generally lead to a decrease in strain rate sensitivity for AHSS by increasing the amount of athermal, long-range dislocation interactions that contribute to flow stress. The values of $m$ at 296 K in the strain rate range of $0.001 - 250 \text{s}^{-1}$ for a wide variety of AHSS, including IF, high strength IF, HSLA, BH, DP, TRIP, CP, and press hardened steel (PHS), are plotted versus the quasi-static YS in Figure 2.14b. This plot shows a general power law decrease in $m$ with increasing quasi-static YS. An analogous plot for the TS showed a similar trend. These plots demonstrate that increasing the strength of first generation AHSS by a variety of strengthening mechanisms leads to a significant decrease in flow stress strain rate sensitivity. This broad trend arises because most strengthening mechanisms employed in AHSS make athermal contributions to the flow stress, which reduces strain rate sensitivity.

### 2.5.2 Adiabatic Heating

During a tensile test, a portion of the work to plastically deform the specimen is transformed to heat. The temperature rise associated with plastic deformation is given by Equation 2.7 where $\beta$ reflects the fraction of work that is transformed to heat, $C_p$ is heat capacity at constant pressure, and $\rho$ is density [133].

$$\Delta T = \frac{\beta}{\rho C_p} \int_0^\epsilon \sigma d\epsilon$$  \hspace{1cm} (2.7)

The linear thermal diffusion distance, $d_t$ in a material is given by Equation 2.8 where $\alpha$ is thermal diffusivity and $t$ is time [133].

$$d_t = 2\sqrt{\alpha t}$$  \hspace{1cm} (2.8)

Equation 2.8 shows that the distance of heat diffusion in a tensile specimen is related to the square root of time. High speed tensile tests can have test durations on the order of milliseconds to seconds, which is an insufficient amount of time for heat diffusion from the interior of a sheet specimen to a free surface. Therefore, adiabatic heating of the specimen occurs. Figure 2.15 shows the maximum temperature change of Duracorr® and 18 Cr-CbTM sheet type tensile specimens as a function of strain rate. Duracorr® and 18 Cr-CbTM are ferritic stainless steels, which had thicknesses of 1.651 mm and 1.168 mm, respectively. The maximum specimen temperature was measured experimentally with a thermocouple (refers
to the maximum specimen temperature regardless of amount of strain) and predicted from a heat transfer model developed by G.P. Martins [139]. Between 0.01 and 100 s\(^{-1}\), the UTS range was approximately 625 – 750 MPa and the UE range was 20 – 25% for the Duracorr\(^{\text{®}}\) and the UTS range was approximately 570 – 690 MPa and the UE range was 20 – 25% for the 18 Cr-Cb\(^{\text{TM}}\). Adiabatic heating was responsible for a maximum temperature change of approximately 60 – 70 K in the strain rate range of 0.1 – 100 s\(^{-1}\). This type of heating at high strain rates could have a stabilizing effect on retained austenite in third generation AHSS, which could in turn affect the mechanical response of the steel during forming operations.

![Figure 2.15](image.png)

Figure 2.15  Maximum tensile specimen temperature change as predicted by a heat transfer model and measured experimentally with a thermocouple as a function of strain rate for two ferritic stainless steels. The experimentally measured temperature was the maximum temperature regardless of amount of strain. Due to thermocouple response time, experimental measurements could only be made at lower strain rates. Model developed by G.P. Martins at CSM. Figure re-plotted from Clarke [139].

### 2.5.3 Effect of Strain Rate on AHSS Deformation Behavior

Advanced high strength steels typically have multiple microconstituents that contribute to a composite-like mechanical behavior of the steel. The strain rate response of a few single phase steels will be briefly reviewed to provide insight into how individual microconstituents contribute to the observed strain rate response of multiconstituent AHSS. Then, tensile test studies in the quasi-static to low dynamic strain rate regimes (0.0001 – 1000 s\(^{-1}\)) on first generation TRIP and DP steels will be discussed.
The flow stress at or near yielding as a function of strain rate for IF steel, ferritic stainless steel, martensitic steel, and austenitic stainless steels is shown in Figure 2.16. The ferritic steels in Figures 2.16a and 2.16b and austenitic steels in Figure 2.16c show strong positive strain rate sensitivity at all strain rates with greater strain rate sensitivity above 10 s\(^{-1}\) [9, 134, 139]. The ferritic stainless steel has slightly greater positive strain rate sensitivity than the austenitic stainless steels, but the strain rate sensitivities of the IF steels and the austenitic stainless steels are very similar. Considering only the crystal structure, the BCC ferrite is expected to have greater strain rate sensitivity than the FCC austenite, because short range dislocation barriers, particularly the Peierl’s stress, are more difficult to overcome in the BCC lattice leading to a higher flow stress [45]. In Figure 2.16a, the IF-4 (0.004C-0.9Mn-0.09P wt.%) steel has higher quantities of C, Mn, and P alloying than the IF-3 (0.003C-0.2Mn-0.01P wt.%) steel, and the IF-4 steel has lower strain rate sensitivity than the IF-3 steel. The solid solution strengthening of the ferrite led to a decrease in strain rate sensitivity, which is consistent with the findings of Larour et al. that solid solution strengthening can cause a significant decrease in strain rate sensitivity [134, 143]. The martensitic steel in Figure 2.16d exhibited little or no strain rate sensitivity for the as-quenched and all tempering conditions, which is attributed to dominant long-range dislocation interactions at all strain rates [141]. The various tempering conditions do show a slight change in strain rate sensitivity with increasing degree of tempering. The as-quenched and 180 °C tempering temperature conditions have a slight negative strain rate sensitivity below 1 s\(^{-1}\). Above 1 s\(^{-1}\), the strain rate sensitivity is very low, but a trend is difficult to discern due to the scatter. The 250 and 400 °C tempering conditions exhibit no strain rate sensitivity at all strain rates. The 500 °C tempering condition exhibits a very slight positive strain rate sensitivity below 1 s\(^{-1}\), and a small positive strain rate sensitivity above 1 s\(^{-1}\). With increasing tempering temperature, the high as-quenched dislocation density decreases, the carbon in solution redistributes and carbides precipitate, and the martensite laths coarsen [24]. Overall, the contribution of long-range dislocation interactions to the flow behavior decreases with increasing degree of tempering, which is responsible for the increase in strain rate sensitivity.

True stress-strain curves over a strain rate range of 0.00089 – 490 s\(^{-1}\) for a commercial DP steel with 15 vol.% martensite are shown in Figure 2.17a [134]. Flow stress increases, uniform elongation decreases, and work hardening rates stay the same with increasing strain rate. The curves also show that the yield point becomes more pronounced with increasing strain rate. Inertial effects upon loading the specimen at high displacement rate could cause the yield point to appear more pronounced (thereby being a test artifact), but more pronounced yield points are expected with increasing strain rate (meaning it is real deformation behavior). The YS and TS as a function of strain rate are shown in Figure 2.17b. Strain rate sensitivity is
Figure 2.16  (a) Flow stress at 0.02 true strain versus strain rate from two IF steels: IF-3 (0.003C-0.2Mn-0.01P wt.%.) and IF-4 (0.004C-0.9Mn-0.09P wt.%). Figure from Bruce [134]. (b) Yield strength versus strain rate for 409 ULTRA FORM® ferritic stainless steel. Figure from Clarke [139]. (c) Yield strength versus strain rate for 304L and 309 austenitic stainless steels. Figure from Lichtenfeld [9]. (d) 2% offset flow stress versus strain rate for a 0.21C-1.46Mn-1.47Si-0.03Al wt.% steel that was quenched and tempered at various temperatures for one hour (except 180 °C for 20 min). Figure from Addessio [141].
higher for the YS than the TS, which is mostly due to more long range dislocation interactions and higher dislocation density involved in the plastic deformation at the TS. At high strain levels, such as near the TS, adiabatic heating also contributes a softening effect that opposes the stress increase due to increased strain rate. The strain rate sensitivity increases at higher strain rates for both YS and TS, which is likely a reflection of the BCC ferrite matrix. The dynamic/static strength ratios of YS and TS decrease with increasing martensite content of DP steels, as shown in Table 2.1 from results produced by Kircher [140]. This trend demonstrates the composite mechanical behavior of DP steel with respect to strain rate where the ferrite exhibits strong positive strain rate sensitivity and the martensite exhibits little or no strain rate sensitivity. The higher dynamic/static strength ratio for the YS than the TS in Table 2.1 is consistent with the higher strain rate sensitivity of the YS compared to the TS for the DP steel in Figure 2.17b. The decrease in the dynamic/static strength ratio as the volume percent of martensite increases and the lower dynamic/static strength ratio for TS than YS are also consistent with the findings by Larour et al. for the strain rate sensitivity parameter as a function of volume percent hard phase and as a function of plastic strain [143].

The difference in the tensile mechanical response of TRIP steels compared to DP steels is primarily due to the presence of retained austenite in TRIP steels. The effects of the
Table 2.1 – Dynamic/Static (D/S) strength ratios for the yield and tensile strengths of DP steels with different amounts of martensite. Dynamic data came from a strain rate of 100 s⁻¹, and static data came from a strain rate of 0.001 s⁻¹. Data from Kircher [140].

<table>
<thead>
<tr>
<th>Martensite Vol.%</th>
<th>Yield Strength D/S ratio</th>
<th>Tensile Strength D/S ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>13</td>
<td>1.43</td>
<td>1.13</td>
</tr>
<tr>
<td>28</td>
<td>1.28</td>
<td>1.10</td>
</tr>
<tr>
<td>39</td>
<td>1.20</td>
<td>1.03</td>
</tr>
<tr>
<td>49</td>
<td>1.11</td>
<td>1.04</td>
</tr>
</tbody>
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amount and stability of retained austenite on high strain rate mechanical properties have been studied. Choi et al. examined the effect of strain rate on two TRIP steels with differing carbon content (0.10C-1.52Mn-1.48Si-0.046Al-0.51Cu wt.% and 0.14C-1.51Mn-1.49Si-0.050Al-0.51Cu wt.%) [144]. Each steel was heat treated to obtain a low and a high amount of austenite. The 0.10 wt.% C steel contained 11 vol.% and < 3 vol.% retained austenite, and the 0.14 wt.% C steel contained 16 vol.% and < 3 vol.% retained austenite. Quasi-static tensile tests on the TRIP steels showed that increasing retained austenite decreases YS but increases UTS, UE, and TE. The UTS versus strain rate data for both the low and high retained austenite fractions are shown in Figure 2.18a for the 0.10 wt.% C steel and in Figure 2.18b for the 0.14 wt.% C steel. Figure 2.18 shows that the amount of retained austenite has almost no impact on the UTS strain rate sensitivity. Similarly, Kircher found that the amount of retained austenite had very little or no effect on the YS and UTS strain rate sensitivities for TRIP steel containing 9 – 16 vol.% retained austenite [140]. Figure 2.18 also shows that for the low austenite amounts, the 0.10 wt.% C steel has higher strain rate sensitivity than the 0.14 wt.% C steel, which is similar to the effect of solid solution strengthening on strain rate sensitivity found for the IF steels in Figure 2.16a. For the high austenite amounts, the 0.10 wt.% C steel has a small positive strain rate sensitivity while the 0.14 wt.% C steel has a slight negative strain rate sensitivity, but the strain rate sensitivities above 1 s⁻¹ are similar for the steels. While the amount of retained austenite had little effect on the YS and UTS strain rate sensitivities, the amount of retained austenite did affect the UE as a function of strain rate. The specimens with high levels of retained austenite exhibited a decrease in uniform elongation with increasing strain rate, while those with low levels of retained austenite showed little change in uniform elongation [144].

The stability of the retained austenite can have a greater impact on high strain rate behavior of first generation TRIP steel than the amount of the retained austenite [14]. Choi et al. examined TRIP steels with 0.10 wt.% C and 0.15 wt.% C that were each heat treated to produce two different austenite stabilities with similar amounts of austenite (approximately
Figure 2.18 UTS versus true strain rate for two TRIP steel compositions: (a) 0.10C-1.52Mn-1.48Si-0.046Al-0.51Cu wt.% and (b) 0.14C-1.51Mn-1.49Si-0.050Al-0.51Cu wt.% Each steel was heat treated to obtain a high and low amount of austenite. Figures from Choi et al. [144].

7 vol.% and 11 vol.% austenite, respectively) [14]. For both the 0.10 wt.% C and 0.15 wt.% C steels at strain rates above 10 s$^{-1}$, the lower austenite stability condition had a higher positive strain rate sensitivity than the same steel chemistry with higher austenite stability. Below 10 s$^{-1}$, the UTS was the same for the low and high stability conditions for each steel chemistry. At most of the strain rates, the UE was higher for the high stability condition. These results suggest that at low strain rates differences in austenite stability affect ductility with little impact on strength, but at high strain rates, the austenite stability still affects ductility and also affects flow stress strain rate sensitivity. The effect of strain rate on the austenite to martensite transformation in first generation TRIP steels will be discussed in Section 2.5.4.

Several literature studies compare the high strain rate mechanical properties of commercially produced equivalent strength DP and TRIP steels [90, 145–148]. The stress-strain curves for first generation DP and TRIP steels of equivalent strength at low and high strain rates from a study by Kim et al. are shown in Figure 2.19a [148]. The curves demonstrate the difference in work hardening behavior between DP and TRIP steels. DP steels have high work hardening rates at low strain levels. In contrast, TRIP steels have higher work hardening rates at larger strain levels and have prolonged work hardening due to the strain-induced transformation of retained austenite to martensite. The result is that TRIP steels
have higher uniform and total elongations with lower yield strength than equivalent tensile strength DP steels. The TRIP780 and DP780 exhibited similar positive strain rate sensitivities for YS and UTS. For both steels, the UE and TE increase slightly with increasing strain rate up to 1 s\(^{-1}\). Above 1 s\(^{-1}\), the TE increases more rapidly with increasing strain rate while the UE decreases. Therefore, plastic instability is reached sooner and greater localized deformation occurs with increasing strain rate. Fractography of commercial first generation DP and TRIP steels showed similar dimpled, ductile fracture surfaces at low and high strain rates [147, 148]. In a study by Huh et al., the YS strain rate sensitivity for equivalent strength DP and TRIP steels increases with increasing strain rate up to 200 s\(^{-1}\), although the increase is greater for the DP steels above 1 s\(^{-1}\) as shown in Figure 2.19b [147]. Similar trends in strain rate sensitivity were observed for flow stress measured at various offsets up to necking, but the flow stress became less sensitive to strain rate with increasing amount of plastic deformation. These four grades of steel were also tested at various strain rates by Oliver et al. [146]. The DP600 and TRIP600 exhibited similar YS strain rate sensitivities, but the DP600 had slightly higher UTS strain rate sensitivity. The DP800 YS strain rate sensitivity was slightly higher than the TRIP800, but the UTS strain rate sensitivities were similar. Peixinho et al. found that DP600 had greater YS and UTS strain rate sensitivity than TRIP600 for strain rates 1 – 200 s\(^{-1}\) [145]. In a study by Curtze et al., both the YS and UTS strain rate sensitivities were found to be lower for TRIP700 than DP600 at all strain rates in the range of 0.001 – 1000 s\(^{-1}\) [90]. For equivalent strength levels, the YS and UTS strain rate sensitivities are either very close between DP and TRIP steels or the DP grade
has slightly higher strain rate sensitivity. A variety of factors such as alloying, austenite stability, and the amounts of ferrite, martensite, and/or bainite contribute to the strain rate sensitivities of the DP and TRIP steels, which is why the strain rate sensitivities may be similar between these two classes of steel in some studies and differ in other studies. The flow stress strain rate sensitivity does generally decline with increasing nominal strength level of the commercial DP and TRIP grades, presumably due to increased alloying and/or larger quantities of less strain rate sensitive microconstituents such as martensite [90, 146–148].

Curtze et al. also examined the combined effects of temperature (-100 – 235 °C) and strain rate (0.001 – 1250 s⁻¹) on DP600 and TRIP700 [90]. The YS, UTS, UE, and TE as a function of temperature at various strain rates are shown in Figures 2.20a and 2.20b for DP600 and in Figures 2.20c and 2.20d for TRIP700. The DP600 had a chemical composition of 0.12 wt.% C, 0.75 wt.% Si+Cr+Mo, and 1.5 wt.% Al+Mn. The TRIP700 had a chemical composition of 0.2 wt.% C, 0.4 wt.% Si+Cr+Mo, and 2.9 wt.% Al+Mn, and it contained 12 vol.% retained austenite. The YS of the DP600 was more sensitive to temperature than the YS of the TRIP700, but the UTS of the TRIP700 was more sensitive to temperature than the UTS of the DP600 due to the effect of temperature on the austenite to martensite transformation (e.g. stress-assisted transformation below $M_s$ and strain-induced transformation above $M_s$). The sensitivity of the retained austenite to temperature in the TRIP700 also led to an increase in UE and TE on the order of 10% strain from -50 °C to 115 °C followed by a decrease in UE and TE with increasing temperature; whereas, the DP600 exhibited very little temperature sensitivity of the UE and TE above -50 °C. At most temperatures, the UE and TE decreased with increasing strain rate for the TRIP700, but the DP600 UE only decreased slightly with increasing strain rate and the TE was relatively insensitive to strain rate. There were similarities for both steels in the overall trends for YS and UTS with respect to both temperature and strain rate: the UTS was more sensitive to temperature than the YS, the temperature sensitivity of the UTS was similar at all strain rates, and YS was more sensitive to temperature at the higher strain rates (750 and 1250 s⁻¹).

### 2.5.4 Effect of Strain Rate on the Strain-Induced Austenite to Martensite Transformation

According to the Olson-Cohen model for the austenite to martensite transformation, increasing strain rate should promote shear band formation and in turn increase the transformation rate [99]. However, increasing the specimen temperature is expected to increase stacking fault energy making slip bands less planar and thus, making shear band formation more difficult [99]. Increasing temperature will also reduce the chemical driving force
Figure 2.20  Mechanical properties from tensile tests in the temperature range -100 – 235 °C and the strain rate range 0.001 – 1250 s⁻¹: (a) DP600 YS and UTS, (b) DP600 UE and TE, (c) TRIP700 YS and UTS, and (d) TRIP700 UE and TE. Figures from Curtze et al. [90].

for transformation [99]. Studies of austenitic stainless steels have shown a pronounced reduction in the overall austenite to martensite transformation with increasing strain rate, which was attributed to adiabatic heating [8–13]. For austenitic stainless steels, the effect of adiabatic heating seems in most cases to outweigh any increase in the mechanical driving force for the transformation at strain rates up to 1000 s⁻¹. Thus, adiabatic heating could potentially have a stabilizing effect on retained austenite in AHSS if a significant amount of the austenite to martensite transformation occurs at high strain levels when adiabatic heating is greatest. The extent of adiabatic heating will differ between austenitic stainless steels and other AHSS with less alloying and mostly BCC/BCT microstructures, because
the amount of plastic work (area under the stress-strain curves), thermal conductivities, and heat capacities differ. For 304 stainless steel, the heat capacity is 500 J/(kg·°C), the thermal conductivity is 16.2 W/(m·°C), and the density is 8.03 g/cm³, which means that the thermal diffusivity is $4.0 \times 10^{-6} \text{ m}^2/\text{s}$ [149]. For HSLA steel, the heat capacity is 481 J/(kg·°C), the thermal conductivity is 89 W/(m·°C), and the density is 7.87 g/cm³, which means that the thermal diffusivity is $2.4 \times 10^{-5} \text{ m}^2/\text{s}$ [150]. Equation 2.7 indicates that the higher density and heat capacity of the 304 stainless steel will produce a smaller change in temperature compared to the HSLA steel for constant $\beta$ and area under the stress-strain curve. Since the thermal diffusivity of HSLA steel is an order of magnitude greater than the thermal diffusivity of 304 stainless steel, the thermal diffusion distance in HSLA steel is more than double the distance in 304 stainless steel (for a constant length of time) according to Equation 2.8. Therefore, heat dissipation can occur more readily for HSLA steel, and nearly complete adiabatic heating will occur at a slightly higher strain rate for HSLA steel than for 304 stainless steel (assuming specimen dimensions are the same) independent of heat capacity. Finally, the areas under the stress-strain curves can vary widely between different steels (due to different YS, UTS and UE) and lead to different amounts of heating according to Equation 2.7.

Type 304L stainless steel is a fully austenitic metastable stainless steel that exhibits gradual austenite to martensite transformation throughout plastic deformation at a quasi-static strain rate. Figure 2.21a shows the volume percent of martensite that forms as a function of strain at various strain rates for 304L stainless steel from a study by Lichtenfeld [9]. There is a substantial reduction in the austenite to martensite transformation rate between the quasi-static strain rate, 0.000125 s⁻¹, and the two higher strain rates, 0.125 s⁻¹ and 100 s⁻¹. The estimated maximum temperature increase calculated from the area under the true stress-strain curve was 59 °C for 0.125 s⁻¹ and 65 °C for 100 s⁻¹. Therefore, the reduction in the transformation rate was attributed to adiabatic heating stabilizing the austenite. The YS, UTS, and UE as a function of strain rate are plotted in Figure 2.21b. The YS has positive strain rate sensitivity throughout the strain rate range. At yielding, the microstructure is almost entirely austenite at all strain rates. In contrast, the amount of austenite contributing to the UTS changes with increasing strain rate. The UTS exhibits negative strain rate sensitivity up to approximately 0.125 s⁻¹ followed by positive strain rate sensitivity at higher strain rates. There is a drop in UE between 0.00125 and 0.0125 s⁻¹, which likely corresponds to a significant difference in austenite stabilization between the two strain rates. At the low strain rates, the maximum specimen temperature rises rapidly with increasing strain rate, which means that the degree of austenite stabilization increases substantially with increasing strain rate up to 0.125 s⁻¹. The decrease in the austenite to
Figure 2.21  (a) Martensite content as a function of strain at various strain rates and (b) yield strength and ultimate tensile strength as a function of strain rate for 304L stainless steel. Austenite content measured by XRD with Cu Kα radiation. Figures re-plotted from Lichtenfeld [9].

Martensite transformation rate causes the UTS and UE to decrease with increasing strain rate up to 0.125 s\(^{-1}\). At a quasi-static strain rate and different test temperatures, the UTS and TE of 304 stainless steel decrease as temperature increases from room temperature [151]. Adiabatic conditions were reached at 0.125 s\(^{-1}\) in the study by Lichtenfeld, which means that the specimen heating was directly proportional to the area under the stress-strain curves at 0.125 s\(^{-1}\) and higher strain rates. Since the areas under the stress-strain curves were similar at the high strain rates, the adiabatic heating and, therefore, the degree of austenite stabilization was similar at 0.125 s\(^{-1}\) and higher strain rates. The positive UTS strain rate sensitivity at higher strain rates reflects the strain rate sensitivity of the mostly austenitic microstructure. The UE increases slightly with increasing strain rate, but the average UE remains between 42\% – 46\% at high strain rates when most of the austenite is stable in contrast to 60\% at quasi-static strain rates when most of the austenite transforms to martensite. The effects of strain rate on the austenite to martensite transformation and the YS and TS of 304L stainless steel were similarly observed for 304 stainless steel by Rodríguez-Martínez et al., for 301L stainless steel by Takagi et al., for 301LN2B stainless steel by Larour et al., and for 301LN and 304 stainless steels by Talonen et al. [10–13]. Depending on the amount of transformation and the degree of austenite stabilization due to adiabatic heating, the trends in UE and TE differed between the studies. For example, the trend in UE for 301L stainless steel studied by Takagi et al. is very similar to the UE trend for
304L stainless steel in Figure 2.21b, but Larour et al. found the UE and TE both increased with increasing strain rate for 301LN2B stainless steel [9, 11, 13]. In the study of 301LN2B stainless steel by Larour et al., the difference in $\alpha'$ martensite at UTS between 0.0001 and 1 s$^{-1}$ was just over 20 vol.% [13]. Larour et al. suggested that adiabatic heating enhanced ductility with increasing strain rate. In the study of 304L stainless steel by Lichtenfeld, the difference in $\alpha'$ martensite at UTS between 0.000125 and 0.125 s$^{-1}$ was nearly 60 vol.% [13]. The stabilization of a large portion of the austenite due to adiabatic heating led to an observable drop in ductility.

Hecker et al. found that in 304 stainless steel slightly more austenite transformed to martensite below 0.25 strain at 1000 s$^{-1}$ than at 0.001 s$^{-1}$ as shown in Figure 2.22 [8]. Above 0.25 strain, the austenite transformation rate remained very low at 1000 s$^{-1}$ but increased significantly at 0.001 s$^{-1}$. Transmission electron microscope observations showed that the density of shear band intersections at 1000 s$^{-1}$ was approximately double the density of shear band intersections at 0.001 s$^{-1}$ for all strain levels up to 0.30 strain [152]. The rate of shear band formation is expected to increase with increasing strain rate, because deformation by shear band formation is promoted at high strain rates [99]. The greater quantity of shear band intersections formed at the high strain rate is responsible for the slightly greater amount of martensite formed at 1000 s$^{-1}$ for strains less than 0.25 [8]. Beyond 0.25 strain, adiabatic heating suppressed the austenite to martensite transformation at 1000 s$^{-1}$ [8]. From the TEM observations, the martensite still nucleated at shear band intersections at 1000 s$^{-1}$, but adiabatic heating appeared to restrict the repeated nucleation and coalescence of martensite embryos observed at the low strain rate and limited the transformed martensite to the vicinity of the shear band intersections [152]. Lichtenfeld found that for 304L stainless steel the austenite to martensite transformation rate was lower at 100 s$^{-1}$ than at 0.00125 s$^{-1}$ for 0.10 and greater strain [9]. In both studies, the amount of transformed martensite below 0.10 strain at low and high rates is very similar. Based on these studies of 304 series stainless steels, the increase in the mechanical driving force due to increased strain rate appears to be very small and may only have an observable effect on the austenite transformation rate at strain rates above approximately 100 s$^{-1}$.

Adiabatic heating does not seem to have a significant stabilizing effect on retained austenite in first generation TRIP steels for strain rates up to 500 s$^{-1}$ [14–16, 153]. Choi et al. examined the effect of strain rate on the austenite to martensite transformation in two TRIP steels, 0.10C-1.5Mn-1.5Si (wt.%) and 0.15C-1.5Mn-1.5Si (wt.%), that were each heat treated to obtain retained austenite with low and high stabilities [14]. The retained austenite volume fractions as a function of strain at 0.01, 10, and 200 s$^{-1}$ are shown for the 0.10 wt.% C steels
in Figure 2.23a and for the 0.15 wt.% C steels in Figure 2.23b. For both steel chemistries, the low stability condition is designated LS and the high stability condition is designated HS in Figure 2.23. There is no significant difference in the austenite transformation rate between the three strain rates for either the low or high stability 0.10 wt.% C steels. For both the low and high stability 0.15 wt.% C steels, the transformation rate is slightly greater between approximately 8% and 14% strain at 200 s\(^{-1}\) than at 0.01 s\(^{-1}\). Above 18% strain, the austenite is completely transformed to martensite at all of the strain rates. For all four steels, adiabatic heating likely had little effect on the austenite transformation, since the transformation occurred early and rapidly during plastic deformation. Figure 2.23c shows UTS versus strain rate, and Figure 2.23d shows UE versus strain rate. There is positive UTS strain rate sensitivity from 0.01 s\(^{-1}\) to 600 s\(^{-1}\) for all four steels, and all four steels exhibit similar trends in UE. The UE is almost constant or decreases very slightly with increasing strain rate up to 10 s\(^{-1}\) then increases to a peak value at 200 s\(^{-1}\). The increase in UE between 10 and 200 s\(^{-1}\) is attributed to suppressed necking from greater strain rate hardening, and above 200 s\(^{-1}\) the decrease in UE may be due to a softening effect from adiabatic heating. Neither the UTS nor the UE exhibit a significant drop with increasing strain rate, which is consistent with adiabatic heating having little or no stabilizing effect on the austenite transformation rate.

Jiménez et al. examined the combined effects of temperature (-100 to 450 °C) and strain rate (0.0004, 0.01, and 0.4 s\(^{-1}\)) on austenite to martensite transformation in a TRIP700 steel (0.22C-1.64Mn-1.51Al-0.05Si-0.012P wt.%) [15]. The UTS as a function of temperature at
Two experimental TRIP steels 0.10C-1.5Mn-1.5Si (wt.%), labeled 10C, and 0.15C-1.5Mn-1.5Si (wt.%), labeled 15C, were each heat treated for low stability (LS) and high stability (HS) austenite. Retained austenite as a function of strain at various strain rates plotted for (a) 10LS and 10HS and (b) 15LS and 15HS. (c) UTS and (d) UE plotted as a function of strain rate. Figures from Choi et al. [14].

The three strain rates is shown in Figure 2.24a. With increasing temperature, the UTS decreases to a minimum above 100 °C and then increases to a maximum above 200 °C before decreasing again. Below 100 °C, strain rate does not affect the UTS, but between 150 and 350 °C, UTS decreases with increasing strain rate. Increasing strain rate also shifts the UTS peak to higher temperatures. There are two peaks in the uniform plastic elongation (UPE) as a function of temperature shown in Figure 2.24b. The first peak occurs for all three strain rates at approximately 60 °C and the second peak follows the trend of the high
temperature UTS peak. The UPE decreases with increasing strain rate at temperatures up to 300 °C with the largest differences around 200 °C. Figure 2.24c shows the austenite volume fraction as a function of plastic strain at different temperatures and strain rates. The austenite to martensite transformation rate is unaffected by strain rate at 20 °C and 100 °C, but increasing temperature has a substantial stabilizing effect on the austenite. Stabilization of the austenite at elevated temperature contributes to the decline in UTS and the decline in UPE with increasing temperature up to 150 °C. Figure 2.24d shows the austenite volume fraction as a function of plastic strain at 200 °C and various strain rates. The austenite stability is high at 0.4 s\(^{-1}\) at 200 °C, but there is a substantial decrease in austenite stability with decreasing strain rate. The austenite to martensite transformation rate at 0.0004 s\(^{-1}\) is greater at 200 °C than at 22 °C. The strong effect of strain rate on the austenite to martensite transformation at 200 °C is attributed to a dynamic strain aging effect that

![Figure 2.24](image)

**Figure 2.24** Tensile test results for TRIP700 tested at 0.0004, 0.01, and 0.4 s\(^{-1}\) and various temperatures (-100 – 450 °C): (a) UTS as a function of temperature, (b) UPE as a function of temperature, (c) austenite volume fraction as a function of plastic strain at -60, 20, and 100 °C, and (d) austenite volume fraction as a function of plastic strain at 200 °C. Figures from Jiménez et al. [15].
causes dislocation pinning by C atoms in the austenite, which promotes deformation by the strain-induced austenite transformation rather than by dislocation glide at low strain rates. As strain rate increases at 200 °C, the DSA effect diminishes, and austenite is more stable against strain-induced transformation to martensite. The decrease in DSA and the austenite to martensite transformation rate with increasing strain rate lead to the decrease in UTS and UPE with increasing strain rate at elevated test temperatures near 200 °C.

The lack of an effect of strain rate on the austenite to martensite transformation up to 0.30 plastic strain at 20 °C for the TRIP700 is a significantly different result than the stabilizing effect of adiabatic heating observed for metastable austenitic stainless steels [9–11, 15]. There may be several possible reasons for the difference in the austenite behavior. Jiménez et al. used miniature tensile specimens, which may have allowed for more heat dissipation during the test. As discussed at the beginning of this section, the thermal conductivity is likely higher in the TRIP700 steel compared to the austenitic stainless steels, and the higher thermal conductivity would contribute to greater heat dissipation during the test. Changes in the austenite content as a function of strain are also more obvious and dramatic in the austenitic stainless steels, since they are fully austenitic prior to deformation compared to the 15 vol.% of austenite initially present in the TRIP700. Finally, other austenite stabilization factors such as morphology and size may have significant contributions to the austenite stability in the TRIP700 steel that outweigh the small temperature increase from adiabatic heating at 0.4 s⁻¹.
CHAPTER 3
EXPERIMENTAL METHODS

This chapter starts with an experimental design overview that includes discussion on the selection of tensile test conditions and the tensile test matrices. Then, background information is given on the four steels used in this work: QP3Mn, TRIP7Mn, DP980, and QP980. Development of the QP3Mn steel Q&P heat treatment is described in detail including dilatometry results. The rest of the chapter is dedicated to describing the various experimental techniques that were used. The microstructural characterization techniques include light optical microscopy (LOM) and scanning electron microscopy (SEM) of etched microstructures and electron backscatter diffraction (EBSD) of QP3Mn and TRIP7Mn. Austenite volume fraction was measured by X-ray diffraction (XRD) in the as-received steels and for various interrupted tensile test conditions. In situ temperature measurements during tensile testing were made with spot welded thermocouples and a non-contact infrared thermal camera. The tensile test methodology is divided into separate sections according to the type of test frame that was used. A portion of the low strain rate tensile testing was performed on low speed screw-driven test frames. The more complicated high strain rate tensile tests were performed on a servo-hydraulic frame, and the tensile test details, specimen preparation techniques, and data analysis methods are discussed.

3.1 Experimental Design

The experimental designs are organized into separate sections for the main QP3Mn and TRIP7Mn study and the smaller DP980 and QP980 study. The main QP3Mn and TRIP7Mn study involved tensile testing at various strain rates at room temperature, at a quasi-static strain rate at various elevated temperatures, and at various strain rates for test temperatures below and above room temperature. The effects of temperature and strain rate on the austenite to martensite transformation rate were examined with XRD measurements of the austenite volume fraction from interrupted tensile tests. These tests addressed the research objectives in Section 1.1. The DP980 and QP980 study was intended to help understand differences in tensile behavior at various strain rates between first and third generation AHSS and to examine tensile test conditions with various tensile deformation histories to help predict the high strain rate performance of formed parts. Three pre-strain
conditions were examined. The austenite volume fraction was measured with XRD as a function of strain at various strain rates for the QP980.

3.1.1 QP3Mn and TRIP7Mn Steels Study

A medium Mn TRIP steel and a quenched and partitioned (Q&P) steel were selected for this study, because they have very different quasi-static mechanical properties with just two as-processed microstructural constituents in each steel. A 7 wt.% Mn TRIP steel was selected as it was available, and 7 wt.% Mn can produce a ferrite-austenite microstructure containing austenite that undergoes strain-induced transformation to martensite. A lower amount of Mn would increase the chance of forming martensite in the as-processed microstructure, and a higher amount of Mn could potentially increase the austenite stacking fault energy high enough that begins to deform by both strain-induced transformation to martensite and twinning. The specific 0.3C-3Mn-1.6Si (wt.%) chemistry was selected for the Q&P steel based on high retained austenite fractions (> 10 vol.%) and high strength-ductility combinations that were obtained in a previous study by De Moor et al. [28].

The tensile test matrix for various strain rates at room temperature (22 – 26 °C) is provided in Table 3.1. Ten different strain rates in the range of 0.0001 s^{-1} – 200 s^{-1} were selected for tensile testing. Specimen temperature was measured during the tensile test for strain rates up to 10 s^{-1}. For four of the strain rates, tensile tests were interrupted at various amounts of strain to measure the retained austenite with XRD. Tensile tests were also conducted at three different strain rates at either -10 °C or 70 °C to investigate how altering the austenite stability (by altering test temperature) would affect high strain rate deformation behavior. As indicated in Table 3.1 for the 70 °C at 0.0005 s^{-1} tests, two of the tests to fracture and all of the interrupted tests overlap with the quasi-static elevated temperature tensile test matrix in Table 3.2 meaning that the same test conditions appear in both matrices but the actual tests were not duplicated. One extra tensile test at 0.0005 s^{-1} at 70 °C was completed for the test matrix in Table 3.1.

To further investigate strain rate and adiabatic heating effects on flow stress, two additional types of tests were performed at room temperature (22 – 26 °C). The first test involved interrupting a tensile specimen after a certain amount of plastic strain at 0.1/s, allowing the specimen to cool, and then pulling the specimen to fracture at 0.1 s^{-1}. This test was performed at 0.1 s^{-1}, because it is the lowest strain rate at which both the QP3Mn and TRIP7Mn specimens experience near adiabatic heating, which will be discussed in Section 5.1.1. At 0.1 s^{-1}, the effect of adiabatic heating on flow stress is expected to be the most...
Table 3.1 – Tensile test matrix at various strain rates for QP3Mn and TRIP7Mn. Tests were conducted at room temperature (22 – 26 °C) unless otherwise stated. For the interrupted conditions, five tests were performed for QP3Mn and six or seven tests were performed for TRIP7Mn (indicated by 5/6 or 5/7). Two QP3Mn and three TRIP7Mn specimens were interrupted and then subsequently pulled to fracture at 0.0005 s⁻¹ and 0.1 s⁻¹.

<table>
<thead>
<tr>
<th>Strain Rate (s⁻¹)</th>
<th>0.0001</th>
<th>0.0005</th>
<th>0.001</th>
<th>0.01</th>
<th>1</th>
<th>10</th>
<th>50</th>
<th>100</th>
<th>200</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fracture</td>
<td>3</td>
<td>3</td>
<td>3</td>
<td>3</td>
<td>3</td>
<td>3</td>
<td>3</td>
<td>3</td>
<td>3</td>
</tr>
<tr>
<td>Interrupted</td>
<td>-</td>
<td>5/6</td>
<td>-</td>
<td>-</td>
<td>5/6</td>
<td>-</td>
<td>5/6</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>T = -10 °C, Fracture</td>
<td>-</td>
<td>3</td>
<td>-</td>
<td>-</td>
<td>3</td>
<td>3</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>T = -10 °C, Interrupted</td>
<td>-</td>
<td>5/6</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>5/6</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>T = 70 °C, Fracture</td>
<td>-</td>
<td>3*</td>
<td>-</td>
<td>-</td>
<td>3</td>
<td>3</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>T = 70 °C, Interrupted</td>
<td>-</td>
<td>5/7*</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>5/6</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Interrupted, Fracture</td>
<td>-</td>
<td>2/3</td>
<td>-</td>
<td>-</td>
<td>2/3</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

*These tests overlap with the quasi-static elevated temperature tensile test matrix.

As strain rate increases, specimen self-heating contributes a thermal effect to flow behavior and can potentially affect austenite stability. Elevated temperature tensile tests at a quasi-static strain rate were conducted to isolate the effect of temperature on flow stress and the austenite to martensite transformation rate. The test temperatures were selected based on specimen temperatures measured during high strain rate tensile testing. The test matrix is provided in Table 3.2. Interrupted tensile tests were performed at various strain intervals up to the UTS. The TRIP7Mn temperature conditions with very high uniform elongations (70, 85, and 100 °C) were interrupted at seven different strain levels compared to the six interrupted tests at the other temperature conditions. Fewer temperature conditions were tested for the QP3Mn, because the specimen self-heating was lower for the QP3Mn specimens than the TRIP7Mn specimens during high strain rate tensile testing.
### Table 3.2 – Tensile test matrix for elevated temperature testing at 0.0005 s\(^{-1}\) for QP3Mn and TRIP7Mn.

<table>
<thead>
<tr>
<th>Temperature</th>
<th>40 °C</th>
<th>55 °C</th>
<th>70 °C</th>
<th>85 °C</th>
<th>100 °C</th>
<th>115 °C</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fracture, QP3Mn</td>
<td>2</td>
<td>2</td>
<td>2</td>
<td>2</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Interrupted, QP3Mn</td>
<td>5</td>
<td>5</td>
<td>5</td>
<td>5</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Fracture, TRIP7Mn</td>
<td>2</td>
<td>2</td>
<td>2</td>
<td>2</td>
<td>2</td>
<td>2</td>
</tr>
<tr>
<td>Interrupted, TRIP7Mn</td>
<td>6</td>
<td>6</td>
<td>7</td>
<td>7</td>
<td>7</td>
<td>6</td>
</tr>
</tbody>
</table>

### 3.1.2 DP980 and QP980 Steels Study

The tensile test matrix for DP980 and QP980 tests at various strain rates at room temperature (22 – 26 °C) is provided in Table 3.3. The tensile test matrix for DP980 and QP980 is smaller than that for QP3Mn and TRIP7Mn, because it was a smaller side study that was meant to provide high strain rate tensile data for comparison to the QP3Mn and TRIP7Mn data. Additionally, the DP980 and QP980 study was designed to evaluate the influence of pre-conditioning treatments. Three pre-strain conditions were incorporated into the test matrix to investigate how these steels behave at various strain rates with prior deformation, as would be the case for a formed part that is subsequently involved in a crash scenario. Pre-straining was performed in uniaxial tension at a quasi-static strain rate to keep the stress state simple and to avoid temperature effects from specimen heating at higher strain rates, although sheet steel would be subjected to higher strain rates and complex stresses in actual forming operations. Since automotive sheet is often formed and subjected to a bake hardening heat treatment during vehicle assembly, a pre-strained condition (5% PS) and a pre-strained then tempered condition (5% PS + 180 °C, 20 min) were examined to simulate forming and forming followed by bake hardening. An elevated temperature pre-strain condition (5% PS at 80 °C) was also included to study the effect of increasing the austenite stability during pre-straining. Pre-straining was performed on a screw-driven frame and will be described in Section 3.4.2. After the pre-straining step, the specimens were strained to fracture on the high strain rate frame. X-ray diffraction was performed on samples from each of the QP980 interrupted tensile tests.

### 3.2 Materials

The Q&P chemistry, designated as QP3Mn, was provided by ArcelorMittal as a laboratory heat that was 50% cold-rolled and had a thickness of 1.1 mm. The 45 kg (100 lb) vacuum induction processed laboratory heat ingot was hot rolled to a 20 mm plate, air cooled, and hot rolled again to a 2.8 mm thickness. A coiling simulation was conducted at
Table 3.3 – Tensile test matrix for DP980 and QP980. Tests were conducted at room temperature (22 – 26 °C). Both materials were tested for each condition except as noted for the interrupted condition.

<table>
<thead>
<tr>
<th>Strain Rate (s⁻¹)</th>
<th>0.0001</th>
<th>0.001</th>
<th>0.01</th>
<th>0.1</th>
<th>1</th>
<th>10</th>
<th>50</th>
<th>100</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fracture</td>
<td>2</td>
<td>2</td>
<td>2</td>
<td>2</td>
<td>2</td>
<td>2</td>
<td>2</td>
<td>2</td>
</tr>
<tr>
<td>QP980, Interrupted</td>
<td>-</td>
<td>6</td>
<td>-</td>
<td>6</td>
<td>-</td>
<td>6</td>
<td>-</td>
<td>6</td>
</tr>
<tr>
<td>5% Pre-Strain</td>
<td>-</td>
<td>2</td>
<td>-</td>
<td>-</td>
<td>2</td>
<td>2</td>
<td>-</td>
<td>2</td>
</tr>
<tr>
<td>5% Pre-Strain at 80 °C</td>
<td>-</td>
<td>2</td>
<td>-</td>
<td>-</td>
<td>2</td>
<td>2</td>
<td>-</td>
<td>2</td>
</tr>
<tr>
<td>5% Pre-Strain, 180 °C + 20 min</td>
<td>-</td>
<td>2</td>
<td>-</td>
<td>-</td>
<td>2</td>
<td>2</td>
<td>-</td>
<td>2</td>
</tr>
</tbody>
</table>

700 °C in an argon atmosphere. Next, both surfaces were ground to remove the decarburized layer and reduce the thickness to 2.0 mm. Finally, the steel was cold rolled approximately 50%. The cold-rolled QP3Mn steel was subjected to quenching and partitioning salt pot heat treatments that are described in detail in Section 3.2.1. The medium Mn TRIP chemistry, designated as TRIP7Mn, was initially hot rolled, then batch annealed, then cold rolled to a final thickness of 1.4 mm, and finally batch annealed at an intercritical annealing temperature by Baosteel as part of an industrial trial. The TRIP7Mn steel was tested in the as-received condition. The commercially available DP980 and QP980 steels were produced by Baosteel. Both were tested in the as-received condition, and both were 1.6 mm thick. The sheet steel chemistries are shown in Table 3.4.

Table 3.4 – Sheet steel chemistries in weight percent.

<table>
<thead>
<tr>
<th>wt.%</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>Al</th>
<th>N</th>
<th>S</th>
<th>P</th>
</tr>
</thead>
<tbody>
<tr>
<td>QP3Mn</td>
<td>0.28</td>
<td>2.56</td>
<td>1.56</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.049</td>
<td>0.0037</td>
<td>0.002</td>
<td>0.01</td>
</tr>
<tr>
<td>TRIP7Mn</td>
<td>0.14</td>
<td>7.14</td>
<td>0.23</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.056</td>
<td>0.012</td>
<td>0.002</td>
<td>0.007</td>
</tr>
<tr>
<td>QP980</td>
<td>0.2</td>
<td>1.79</td>
<td>1.52</td>
<td>-</td>
<td>0.03</td>
<td>-</td>
<td>0.039</td>
<td>0.004</td>
<td>0.003</td>
<td>0.0062</td>
</tr>
<tr>
<td>DP980</td>
<td>0.09</td>
<td>2.29</td>
<td>0.29</td>
<td>-</td>
<td>0.54</td>
<td>0.2</td>
<td>0.024</td>
<td>0.0042</td>
<td>0.0014</td>
<td>0.014</td>
</tr>
</tbody>
</table>

3.2.1 Quenching and Partitioning Heat Treatments

The QP3Mn tensile specimens in the cold-rolled condition were subjected to quenching and partitioning heat treatments. Heat treatment experiments with various conditions were performed to select one Q&P heat treatment for all of the tensile specimens, and the details and results of the heat treatment experiments are presented in Appendix A. The heat treatment profile shown in Figure 3.1 was chosen to produce an austenite and martensite microstructure with over 10 vol.% austenite, greater than 10% total elongation, and greater than 1500 MPa tensile strength. The heat treatment parameters were based on the work
of De Moor et al. on an alloy with similar amounts of carbon, manganese, and silicon [28]. The samples were held at an austenitizing temperature of 830 °C for 120 s, quenched to 180 °C (approximately 65 °C/s cooling and held 10 s in the quench salt pot), and held at a partitioning temperature of 400 °C for 10 s. Salt pots were used for all three processing steps, and three tensile specimens were heat treated at a time.

![Quenching and partitioning heat treatment profile for QP3Mn tensile specimens.](image)

Thermal simulations were performed for the QP3Mn steel to identify the $A_c3$, $M_s$, and $M_f$, to check that complete austenitization occurred for the selected hold time and temperature, and to make sure that only martensite formed upon cooling at different rates, since slight differences in transfer time between the austenitizing and quenching salt pots for different batches of specimens can result in different cooling rates. Dilatometry was performed with a Gleeble® 3500 and a push-rod contact quartz dilatometer. Details of the Gleeble® 3500 set-up and testing procedures are provided by Araújo [44]. The dilatometry specimens were in the cold rolled condition and had a width of 6 mm, length of 15 mm, and thickness equal to the sheet thickness. The dilatometer was centered along the length of the specimen, and the dilatometer measured the change in the width dimension. During the salt pot heat treatments and the dilatometry experiments, the specimen temperature was measured with a K-type thermocouple spot welded to the center (widthwise and lengthwise) of the specimen. Heating and cooling rates were calculated from the thermocouple measurements. Three separate dilatometry specimens were heated at a rate of 78 °C/s, which is the heating rate of the specimens in the salt pots, to 900 °C and held for 120 s, after which the power was turned off and the specimens were cooled to room temperature. The cooling rate from 900 °C to approximately 350 °C was 115 °C/s and decreased to 28 °C/s between approxi-
mately 350 °C and 100 °C. The cooling rates measured in the quench salt pot were 340 °C/s between 707 °C and 376 °C and 18 °C/s between 357 °C and 192 °C. A fourth dilatometry specimen was run with the same heating rate and hold step and a 26 °C/s controlled cooling rate to simulate a long delay in transferring a tensile specimen from the high temperature salt pot to the quench salt pot. The slow cooled dilatometry curve was identical to the fast cooled curves; only martensite forms at both cooling rates. Figure 3.2a shows a representative dilatometry curve for the dilatometry specimens with the fast cooling rate. There are changes in curvature corresponding to \( A_{c1} \) and \( A_{c3} \) upon heating and for \( M_s \) and \( M_f \) upon cooling. The \( A_{c3} \), \( M_s \), and \( M_f \) were measured from points where the dilatometry curve deviates from linearity and the slope of the curve changes sign (e.g. the slope changes from negative to positive near \( A_{c3} \)) as indicated in Figure 3.2a. The average transformation temperature values from the three curves are: \( A_{c3} = 844 \pm 2 \) °C, \( M_s = 351 \pm 7 \) °C, and \( M_f = 155 \pm 10 \) °C. The \( M_s \) temperature calculated from the chemistry and the Andrews equation in Equation 2.1 is 333 °C. Andrews also proposed an empirical equation for \( A_{c3} \); the equation \( A_{c3} = 910 - 203\sqrt{C} + 44.7Si - 30Mn + 700P + 400Al \) (elements are in weight percent) was used to calculate an \( A_{c3} \) temperature of 822 °C for QP3Mn [29]. The experimentally measured \( A_{c3} \) and \( M_s \) temperatures are close to 20 °C higher than the predicted values. The measured \( A_{c3} \) is slightly higher than the austenitizing temperature used for the heat treatments. However, \( A_{c3} \) increases with increasing heating rate, and therefore, the hold for 120 s at 830 °C is likely long enough for full austenitization.

To check whether full austenitization could occur during the salt pot austenitization hold, a fifth dilatometry specimen was run with a heating rate of 78 °C/s to 820 °C and held for 120 s. The temperature 820 °C was selected as an estimated lower bound on the austenitization temperature, since the temperature of the salt pot fluctuates slightly during the heat treatment due to cooling when the lid is removed. After the austenitizing hold, the dilatometry specimen was heated to 950 °C, held for 30 s, and then the power was turned off and the specimens were cooled to room temperature. The resulting dilatometry curve is shown in Figure 3.2b. The transition between the hold at 820 °C and the heating to 950 °C is abrupt and the curve up to 950 °C is linear, which suggests that most if not all of the sample was completely austenitized. Based on the dilatometry results, the microstructure after the salt pot heat treatment at 830 °C is expected to be completely austenite. The observed microstructure from the Q&P salt pot heat treatments, which is shown and discussed in Section 4.1, appears to be uniform and completely comprised of martensite and austenite.
Figure 3.2  (a) QP3Mn dilatometry curve showing $A_{c1}$ and $A_{c3}$ for a 78 °C/s heating rate. Upon cooling to room temperature, only martensite forms as indicated by $M_s$ and $M_f$. (b) QP3Mn dilatometry curve for an austenitizing hold at 820 °C for 120 s that suggests that the microstructure was completely austenitized.

3.3 Microstructure Characterization

Several characterization techniques were used to examine the four steel microstructures. Optical microscopy was used for the DP980 and QP980 to obtain large fields of view at low magnification to quantify the ferrite and martensite volume fractions. The DP980, QP980, and QP3Mn microstructural features were examined at higher magnification with an SEM. The TRIP7Mn microstructure was examined with EBSD to differentiate between the ferrite and austenite grains. The QP3Mn microstructure was also examined with EBSD to identify austenite morphologies. The volume fraction of retained austenite was measured by XRD.

3.3.1 Metallography

Specimens examined with LOM and SEM were polished through 1 μm diamond slurry and etched with 2 % nital for 15 s. For the QP980 and DP980, constituent fractions were measured by point counting along concentric circles overlaid on optical micrographs. A minimum of 1425 points were counted for each material. For the QP980, the constituents were identified as ferrite or martensite/austenite. The austenite fraction measured by XRD was subtracted from the martensite/austenite constituent fraction to determine the amount of martensite. The DP980 and QP980 microstructures were examined with an FEI Quanta
600i E-SEM in secondary electron imaging mode. The QP3Mn microstructure was examined with a JEOL JSM-7000F FE-SEM in secondary electron imaging mode.

### 3.3.2 Electron Backscatter Diffraction

The TRIP7Mn and QP3Mn EBSD specimens were polished through 1 µm diamond slurry and then polished on a vibratory polisher for 4 hours with 0.02 µm colloidal silica. The EBSD patterns were collected with TSL OIM™ Data Collection 7 using a 0.03 – 0.05 µm step size at a 20 keV operating voltage, and TSL OIM™ Analysis 7 was used to analyze the patterns. A JEOL JSM-7000F FE-SEM and a FEI Helios Nanolab 600i FIB-SEM were used for EBSD. For the TRIP7Mn grain size measurements, the data set was cleaned up using neighbor phase correlation with a 0.1 minimum confidence index.

### 3.3.3 X-Ray Diffraction

Samples for XRD were lightly ground with 400 and 600 SiC paper to remove the surface oxide layer. Samples were thinned in 10 parts deionized water, 10 parts 30% hydrogen peroxide, and 1 part 48% hydrogen peroxide for 10 to 20 minutes to remove material damaged by grinding. The XRD scans were performed with a Phillips X’pert X-ray diffractometer using nickel-filtered copper Kα radiation with a one degree incident beam slit, operating at 45 kV and 40 mA, and using an Xcelerator detector. Samples were scanned over a 2-theta range of 40°–105° with a 200 s dwell time and 0.05° step size. The samples were oriented such that the rolling direction was parallel to the X-ray beam. Repeated measurements of retained austenite in as-processed samples of QP980 fall within 1 vol.% of each other.

PANalytical X’Pert HighScore Plus software was used to calculate the integrated intensities and peak positions from Profit peak fits of the raw data with the Kα2 peaks removed. The retained austenite volume fraction was calculated by the SAE method with Equation 3.1 [154]. In Equation 3.1, \( I \) is the integrated intensity for a given diffraction peak and \( R \) is a parameter based on physics principles that is proportional to the theoretical integrated intensity. Four ferrite peaks, \( \{110\}, \{200\}, \{211\}, \{220\} \), and four austenite peaks, \( \{111\}, \{200\}, \{220\}, \{311\} \), were used for every calculation.

\[
V_\gamma = \frac{\left(\frac{1}{q} \sum_{j=1}^{q} \frac{I_{\gamma j}}{R_{\gamma j}}\right)}{\left(\frac{1}{p} \sum_{i=1}^{p} \frac{I_{\alpha i}}{R_{\alpha i}}\right) + \left(\frac{1}{q} \sum_{j=1}^{q} \frac{I_{\gamma j}}{R_{\gamma j}}\right)}
\] (3.1)
3.4 Tensile Testing

The different tensile specimen geometries are provided. Various tensile test parameters and set-ups are described. The high strain rate tensile test system and data analysis methods are discussed in detail. Finally, in situ specimen temperature measurement methods are discussed.

3.4.1 Tensile Specimen Geometries

The four tensile specimens geometries used in this work are shown in Figure 3.3. All tensile specimens for all four steels were machined with the tensile axis parallel to the rolling direction. All tensile tests to fracture and the interrupted tensile tests for QP3Mn and TRIP7Mn at 0.1 s\(^{-1}\) and lower strain rates were performed using a tensile specimen design from Addessio [141]. The tensile specimens that were deformed at a quasi-static strain rate and an elevated temperature in an oil bath required a pin hole in one of the grips. Interrupted tensile testing on the high strain rate frame was performed with a modified version of a tensile specimen design from Choi et al. [155]. Premature fracture at the notches occurred prior to plastic deformation of the gage length for the QP3Mn specimens. The issue was traced to very low hardness in the notched region post heat treating in the salt pots. The interrupted tensile specimen was re-designed with much less severe notches, and the heat treatment technique was changed such that the specimens went into the salt pots notch-end first. The changes resulted in more uniform hardness throughout the grip with the notches and eliminated premature failure at the notches.

The high hardness of the steels in this work led to challenges with gripping the specimens securely during testing. To reduce slipping in the grips, 80 grit SiC paper was used to grind a crosshatch pattern into the grip portions of the tensile specimen prior to testing. The high strength of the steels also led to slipping in the interruption device used for interrupting tests on the high strain rate frame. Therefore, both notches of the high strain rate interrupted tensile specimens were modified with a 0.5 mm cut from a diamond saw at the narrowest point of the notched region (the QP3Mn specimen design with shallow notches was strong enough to also require these cuts). In Figures 3.3c and 3.3d, the diamond cuts are represented by lines perpendicular to the notches. The cuts from the diamond saw produced very consistent fractures at the notches with no slipping in the interruption device.
Figure 3.3  (a) Regular tensile specimen geometry used for most tensile tests. (b) Regular tensile specimen geometry with a pin hole for submerged tensile tests in an oil bath. (c) Geometry for interrupted tensile testing on the high strain rate frame. (d) A second interrupted tensile specimen design with less severe notches that was used for QP3Mn to prevent premature fracture at the notch. For both interrupted tensile specimen designs the lines perpendicular to the notches indicate where small cuts were made with a diamond saw prior to testing. The specimen thickness for all designs is the sheet thickness.

3.4.2 Low Strain Rate Tensile Testing

A portion of the tensile testing for all four steels was performed with low speed electro-mechanical screw-driven test frames. All of the QP3Mn and TRIP7Mn tensile tests to
fracture and interrupted tensile tests at strain rates below 1 s\(^{-1}\) were completed on screw-driven frames to obtain better load resolution and more complete strain data than possible with the high strain rate frame. The DP980 and QP980 5% pre-straining was performed on screw-driven frames to obtain accurate, repeatable pre-strain amounts with an extensometer.

The QP3Mn and TRIP7Mn tensile tests at strain rates of 0.0001 – 0.1 s\(^{-1}\) in the experimental matrix in Table 3.1 were performed on a screw-driven MTS Alliance RT/100 test system with a MTS 25.4 mm +12.7 mm/-2.5 mm extensometer for the room temperature tests and a Shepic 25.4 mm + 12.7 mm extensometer for the tests at -10 °C and 70 °C. The displacement rates and data acquisition (DAQ) rates for the various strain rates are provided in Table 3.5.

Strain rate change and interrupted tensile tests were performed on the QP3Mn and TRIP7Mn to try to separate the effects of strain rate and adiabatic heating on flow stress. These tests were performed on a screw-driven MTS Alliance RT/100 test system with a MTS 25.4 mm +12.7 mm/-2.5 mm extensometer at room temperature. During the strain rate change tests, the displacement rate of the test was changed several times between 0.0127 and 2.54 mm/s to alter back and forth between 0.0005 and 0.1 s\(^{-1}\) strain rates. The purpose of the strain rate change test was to observe the flow stress response to a change in strain rate from 0.0005 s\(^{-1}\) to 0.1 s\(^{-1}\) at a constant temperature, strain level and dislocation structure. The interrupted tensile tests involved pulling the tensile specimen to a specific level of strain (4% and 6% for QP3Mn and 15%, 20%, and 25% for the TRIP7Mn), unloading the specimen, allowing the specimen to cool to 25 °C, and then pulling the specimen to failure. These tests were performed at 0.005 s\(^{-1}\) and 0.1 s\(^{-1}\). The purpose of the interrupted tensile test was to compare the effect of strain rate plus adiabatic heating and strain rate alone on flow stress. The low strain rate, 0.0005 s\(^{-1}\), was selected due to very little self heating of the specimen during the tensile test (\(\leq 5 ^\circ C\)). The high strain rate, 0.1 s\(^{-1}\), was selected because the specimen experiences a significant amount of heating during the test but the strain rate sensitivity is low.

In Table 3.1, some of the 70 °C tensile tests overlap with the quasi-static elevated temperature tensile test matrix in Table 3.2. Only the three 0.1 s\(^{-1}\) and one extra 0.0005 s\(^{-1}\) 70 °C tensile tests were performed in air on the Alliance frame. Data for the other two 0.0005 s\(^{-1}\) tests to fracture and the 0.0005 s\(^{-1}\) interrupted tests at 70 °C came from the submerged tensile tests. The methods for producing the -10 and 70 °C temperature conditions in air are described in Section 3.4.6.

A submerged tensile set-up was used for the QP3Mn and TRIP7Mn quasi-static elevated temperature experiments in Table 3.2. The tensile tests were performed in Paratherm oil
Table 3.5 – Tensile test parameters for low strain rate testing.

<table>
<thead>
<tr>
<th>Strain Rate (s⁻¹)</th>
<th>Displacement Rate (mm/s)</th>
<th>Data Acquisition Rate (Hz)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.0001</td>
<td>0.00254</td>
<td>1</td>
</tr>
<tr>
<td>0.0005</td>
<td>0.0127</td>
<td>10</td>
</tr>
<tr>
<td>0.001</td>
<td>0.0254</td>
<td>10</td>
</tr>
<tr>
<td>0.01</td>
<td>0.254</td>
<td>100</td>
</tr>
<tr>
<td>0.1</td>
<td>2.54</td>
<td>1000</td>
</tr>
</tbody>
</table>

on the screw-driven Instru-Met test system with a 0.0127 mm/s displacement rate, a 10 Hz DAQ rate, and a Shepic 25.4 mm + 12.7 mm extensometer. The specimens were submerged for at least 3 min prior to testing. The measured test temperatures were within ±1 °C of the target temperature for all the temperatures examined, except for the 40 °C condition for which the measured test temperatures were within ±1 °C of 42 °C.

The pre-straining for the three DP980 and QP980 pre-strained conditions in Table 3.3 was performed on two screw-driven test frames. The 5 % pre-straining at room temperature was performed on the Alliance frame at a 0.0212 mm/s displacement rate with a MTS 25.4 mm +12.7 mm/-2.5 mm extensometer. Specimens for the 5 % PS + 180 °C, 20 min condition were subsequently heat treated in an open air furnace at 180 °C for 20 min. The 5 % pre-straining at 80 °C was performed with the submerged tensile test set-up on the Instru-Met frame at a 0.0212 mm/s displacement rate with a Shepic 25.4 mm + 12.7 mm extensometer. The samples were submerged in Paratherm oil heated to 82 °C ± 1 °C for 5 min prior to straining.

Following pre-straining for all three test conditions, the DP980 and QP980 specimens were kept below 0 °C prior to being pulled to fracture on the high strain rate frame to minimize strain aging at room temperature. Strain gages for tensile testing on the high strain rate frame had to be attached after pre-straining. The epoxy for the strain gages required a two hour cure at 75 °C, which was found to not have a significant impact on strain aging.

### 3.4.3 High Strain Rate Tensile Testing

Tensile tests to fracture and interrupted tensile tests above 0.1 s⁻¹ strain rate for the QP3Mn and TRIP7Mn were performed on the high strain rate frame. All of the DP980 and QP980 tensile tests to fracture and the QP980 interrupted tensile tests were performed on the high strain rate frame. Most of the high strain rate methodology used in this work was developed by previous users [134, 139, 141, 155]. A new technique that was introduced for
high strain rate tensile testing from this work is the use of two strain gages to measure load at strain rates above 10 s\(^{-1}\). The two strain gage technique for measuring load is described in Section 3.4.4.

A MTS 810 high rate servo-hydraulic testing system provides high strain rate tensile testing capability. The system is comprised of a 500 kN (110,000 lb) capacity frame with a 50 kN (11,000 lb) capacity actuator. A two servo valve system allows the actuator to move at low and high rates. Low actuator velocities are provided by a 38 L/min (10 gal/min) servo valve with a 38 L/min (10 gal/min) hydraulic pump. High actuator velocities are provided by a 1514 L/min (400 gal/min) servo valve with a 19 L (5 gal) capacity oil accumulator. The peak velocity is 13.5 m/s at zero load, and the velocity is specified to be 10 m/s at 50% of the peak load capacity of the actuator. The system is controlled by a MTS TestStar\textsuperscript{TM} digital controller.

A schematic of the fixturing for the high rate servo-hydraulic tensile test machine is provided in Figure 3.4. A 400 kN (90,000 lb) Kistler\textsuperscript{®} piezoelectric quartz crystal load washer (model: 9071A, SN: 1188687) is located between the upper grip and the machine crosshead. A quartz crystal load washer is used to measure load for high strain rate tensile tests, because a load washer has a quicker response time and much lower mass than a traditional load cell. The amount of mass between the load washer and the tensile specimen is directly proportional to the amplitude of the load washer ringing, which refers to oscillations in the load data due to resonance of the load washer at high strain rates. Load ringing will be discussed in greater detail in Section 3.4.4. The grips are made of a titanium housing with knurled steel wedges and are specifically designed to minimize mass between the load washer and the specimen. The specialized titanium grips can be used with thin sheet specimens that have a maximum thickness of 2.03 mm (0.08 in) and a maximum width of 15.2 mm (0.60 in). Heavier steel grips were used for the DP980 and QP980 tensile tests, because new knurled steel wedges had to be machined for the titanium grips. The lower grip is connected with a slack adapter, which allows the actuator to fully accelerate to the selected testing velocity. The slack adapter allows for up to approximately 203 mm (8.0 in) of travel before engaging the lower grip and loading the tensile specimen.

Data acquisition rates up to 2.5 MHz were attained through the National Instruments\textsuperscript{TM} oscilloscope software program VirtualBench Scope\textsuperscript{TM} and a National Instruments\textsuperscript{TM} 12-bit data acquisition board with four data channels. The National Instruments\textsuperscript{TM} data acquisition system can record time, actuator displacement, load from the load washer, and two strain measurements. Alternatively, an extra strain channel can be measured instead of actuator displacement for tests at strain rates above 10 s\(^{-1}\) during which one high elongation strain
Figure 3.4  Schematic of high strain rate tensile test fixturing for the MTS 810 high rate servo-hydraulic frame. Redrawn from Bruce [134].

gage is used to measure strain and two elastic strain gages are used to measure load. The displacement rates and DAQ rates used for tensile tests at 1 s\(^{-1}\) and above are provided in Table 3.6. The data acquisition rates were selected to obtain over 2,000 data points during the tensile test. For a few tests with low total elongations, the number of data points collected during the test was 1000 – 2000.

Table 3.6 – Tensile test parameters for high strain rate testing.

<table>
<thead>
<tr>
<th>Strain Rate (s(^{-1}))</th>
<th>Displacement Rate (mm/s)</th>
<th>Data Acquisition Rate (Hz)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>25.4</td>
<td>10,000</td>
</tr>
<tr>
<td>10</td>
<td>254</td>
<td>200,000</td>
</tr>
<tr>
<td>50</td>
<td>1270</td>
<td>1,000,000</td>
</tr>
<tr>
<td>100</td>
<td>2540</td>
<td>1,000,000</td>
</tr>
<tr>
<td>200</td>
<td>5080</td>
<td>2,500,000</td>
</tr>
</tbody>
</table>

A stopper fixture was designed by Choi et al. to perform interrupted tests during which the specimen can be pulled to a specific amount of strain without stopping the actuator during a test [155]. A photo of an interrupted geometry tensile specimen with the interruption
device installed is provided in Figure 3.5. The gap distance between the moveable grip and the guide is set according to the target amount of strain for the specimen. During the tensile test, the tensile specimen reduced section deforms until the moveable grip on the interruption device engages with the frame of the device. Then, the specimen deforms and fractures at the notch in the bottom grip section.

Figure 3.5 A high strain rate interrupted geometry tensile specimen with the interruption device installed. Interruption device designed by Choi et al. [155].

3.4.4 High Strain Rate Load Measurement

Before any tensile tests were performed, the load washer required service from Kistler®, because over time the load washer developed a non-linear electrical signal with respect to applied load. The load washer must be compressed when it is installed in the upper grip configuration on the high strain rate frame. A 50 kN (11,000 lb) MTS load cell (model: 20F-02, SN: 113438) and digital force reader (model: 9840, SN: 10294) were used to calibrate the load washer. The calibration of the load washer can differ each time the upper grip configuration is re-assembled and installed on the frame, and the calibration difference can be on the order of 20%. Thus, the calibration was performed five times after installing the upper grip configuration; the resulting calibration curve with all of the data is shown in Figure 3.6.

Figure 3.7a shows engineering stress-strain curves from the high rate frame and the screw-driven Alliance frame for QP3Mn at 0.1 s⁻¹. The stress-strain curves from the two different frames show good agreement. Figure 3.7b shows engineering stress-strain curves from the high rate frame and the screw-driven Alliance frame for TRIP7Mn at 0.0005 s⁻¹.
Figure 3.6  Load washer calibration curve that was used for high strain rate tensile testing.

The yield point elongations and total elongations show good agreement. The flow stress beyond 10% strain is higher for the test performed on the high strain rate frame. At UTS, the difference between the tests on the two frames is 50 MPa. A portion of the difference may be due to inherent scatter between tensile specimens. Among sets of three TRIP7Mn tests with each set run at a constant displacement rate on the screw-driven Alliance frame, the UTS range was 20 – 30 MPa. The load measurement with a quartz crystal load washer during tensile tests on the high strain rate frame has greater noise than a standard load cell, which can be observed in Figures 3.7a and 3.7b. The noise in the load reading from the load washer is ± 29 N compared to ± 3 N for the load cell used on the screw-driven Alliance frame.

To achieve strain rates above 1 s\(^{-1}\), the actuator accelerates through a slack adapter before engaging the end of the slack adapter and loading the specimen. When the actuator hits the end of the slack adapter, an impulse load is transmitted through the load train. The impulse load generates mechanical resonance throughout the load train of the tensile frame and results in oscillations in the load washer data [133]. Examples of load oscillations in stress-strain curves at various strain rates are shown in Figure 3.8a from Bruce [134]. The load washer oscillation frequency is approximately 2500 Hz at all the strain rates, but the amplitude increases with increasing strain rate [134]. Load ringing frequency is inversely proportional to mass, and the amplitude is directly proportional to mass [134]. Therefore,
Figure 3.7 Engineering stress-strain curves from tensile tests performed on the high rate hydraulic frame and the screw-driven Alliance frame: (a) QP3Mn tests at 0.1 s\(^{-1}\) and (b) TRIP7Mn tests at 0.0005 s\(^{-1}\).

decreasing the mass between the tensile specimen gage section and the load measurement device will improve the load data by decreasing the stress wave amplitude and increasing the frequency. A significant reduction in load ringing can be obtained through the use of an elastic strain gage attached to one of the tensile specimen grip sections [134]. Recently,
Yang et al. showed that servo-hydraulic high rate tensile test set-ups can subject the tensile specimen to a bending moment during a high rate test, which contributes to the oscillations in the load data [156]. They demonstrated that the bending artifacts in the load data can be accounted for and removed through the use of two elastic strain gages with one strain gage on each side of the grip. Figure 3.8b shows an example of readings from two elastic strain gages during a QP980 tensile test at 500 s\(^{-1}\); the oscillations due to the bending moment were removed by averaging the data from the two gages [156].

For the QP3Mn and TRIP7Mn tensile tests, an additional elastic strain gage channel was added to the high rate frame in order to record data from two elastic strain gages during tests at 50 s\(^{-1}\) and higher. The QP980 and DP980 tensile tests at 50 s\(^{-1}\) and 100 s\(^{-1}\) were completed prior to the publication of the work of Yang et al. and the addition of a second elastic strain gage channel to the high rate frame [156]. Therefore, the tests were conducted with one elastic strain gage according to the methodology developed by Bruce [134]. The QP980 and DP980 load data likely has artifacts from bending during the test, but the oscillations in the
load data from the one elastic strain gage were small. For example, for one of the QP980 tests at 100 s\(^{-1}\), the peak-valley difference in the load oscillation near UTS was 12 MPa. For all specimens, the elastic strain gage (Vishay Precision Group Micro-Measurements CEA-06-125UN-120) was attached to the grip section at a distance of 6 mm from the fillet as shown in the image of a strain gaged tensile specimen in Figure 3.9. For the QP3Mn and TRIP7Mn specimens, the second elastic strain gage was attached on the opposite side of the grip so as to mirror the first gage. The elastic strain gages were attached and wired according to application procedures from Vishay Precision Group Micro-Measurements [157].

![Figure 3.9](image) Tensile specimen with a high elongation strain gage attached to the gage length and an elastic strain gage attached to the grip section.

The methodology used for data analysis with either one or two elastic strain gages was the same except that the data from a test with two gages were averaged. Prior to testing at a high rate, the tensile specimen was loaded elastically at 0.0005 s\(^{-1}\) up to a load corresponding to less than 75\% of the yield load. The elastic loading provided calibration data for the two elastic strain gages. The specimen was then unloaded, and the high rate tensile test was performed. The data analysis involved calculating a calibration constant for each elastic strain gage from the quasi-static elastic loading data, applying those calibration constants to the elastic strain gage data from the high rate test, and averaging the load data from the two elastic strain gages. The averaged elastic strain gage data were often very close to the average of the load washer data during uniform plastic deformation as shown in Figure 3.10a. The averaged load washer data are shown to compare only the uniform plastic deformation portion of the curve to the averaged elastic strain gage data. The load washer data was averaged using the same number of data points throughout the test, which results in the non-representative elastic loading and necking data. In some tests, there was a significant difference between the averaged load washer data and the averaged elastic strain gage data as shown in Figure 3.10b. For these cases, the calibration values for the strain gage data were adjusted slightly by the same order of magnitude to obtain better agreement between the averaged strain gage data and the averaged load washer data. Previous users of the high
Figure 3.10 Examples of load measurements at 100 s\(^{-1}\) for QP3Mn. One specimen (a) shows good agreement and the other specimen (b) shows poor agreement between the averaged elastic strain gage data and the averaged load washer data in the uniform plastic deformation regime. For example, at time equal to 0.0052 s the difference between the averaged elastic strain gage data and the averaged load washer data is 51 N (8 MPa) for (a) versus 337 N (50 MPa) for (b). The load washer data were averaged over the same number of points for the entire test, which produced the disparities between the averaged data and the raw data in the elastic and necking regimes. (Color image - see PDF version)
rate frame at CSM noted that the load data from one elastic strain gage did not always match the average of the load washer data [134, 139]. The mismatch may be due in part to bending effects in the single elastic strain gage data, but the mismatch was also observed for some specimens with two elastic strain gages in the current work. Therefore, the mismatch may be due to various factors in the test set-up including the initial flatness of the tensile specimen and the alignment of the grips after tightening.

### 3.4.5 High Strain Rate Strain Measurement

Strain was measured with a high elongation strain gage (Vishay Precision Group Micro-Measurements EP-08-250BG-120) attached to the specimen gage section with epoxy (Vishay Precision Group Micro-Measurements M-Bond A12). The high elongation strain gage was attached and wired according to application procedures from Vishay Precision Group Micro-Measurements [157, 158]. The strain gage can potentially measure up to 20% strain, but the gage frequently detached at much lower strains. Strain data were acquired with a quarter-bridge Wheatstone bridge circuit. The non-linearity error of the Wheatstone bridge becomes substantial when measuring plastic strains, because the percent error is a similar magnitude to the percent strain measurement (e.g. approximately 10% error for a 10% strain reading) [159]. The strain data were corrected for the Wheatstone bridge non-linearity using Equation 3.2 where \( e \) is corrected strain, \( F \) is the gage factor, and \( e_i \) is the strain gage reading after converting voltage to strain [159]. For consistency, the data from the elastic gage on the grip section were also corrected for Wheatstone bridge non-linearity.

\[
e = \frac{2e_i}{2 - F \cdot e_i}
\]  

(3.2)

Figure 3.11a shows strain measured over time by a high elongation strain gage during a tensile test. During loading of the tensile specimen, the strain versus time data are mostly linear once the sample begins to plastically deform. A deviation from linearity indicates that the strain gage debonded, as noted in the example in Figure 3.11a. The strain gage debonded prior to or at the onset of necking for all of the tests performed on the high strain rate frame in this study. Therefore, the strain data were extrapolated up to the point of fracture of the specimen. Figure 3.11a shows the strain data extrapolated using the engineering strain rate and time after the strain gage debonded. The frame was run in displacement control meaning that velocity was constant during the test. The constant velocity leads to the engineering strain rate being approximately constant throughout plastic deformation of the specimen, which can be shown mathematically and experimentally. Assuming that all deformation
occurs in the specimen gage length, \( l_0 \), then the velocity of the actuator, \( v \), is equal to the displacement of the gage length, \( \Delta l \), divided by the change in time, \( \Delta t \). Equation 3.3 shows that engineering strain rate, \( \dot{e} \), is equal to \( v \) divided by \( l_0 \). Figure 3.11b shows the extensometer data from a QP3Mn tensile test. The strain rate is low during elastic loading, and it increases and becomes constant throughout plastic deformation until fracture.

\[
\dot{e} = \frac{\Delta l}{l_0} \frac{1}{\Delta t} = \frac{v}{l_0}
\]  

(3.3)

Figure 3.11  (a) Engineering strain and engineering stress versus time for a QP3Mn tensile test at 1 s\(^{-1}\). The strain was measured with a strain gage and results were extrapolated after the strain gage debonded. (b) Engineering strain and engineering stress versus time for a QP3Mn tensile test at 0.0005 s\(^{-1}\). The strain was measured by an extensometer.

3.4.6 Tensile Test Set-Ups for Various Temperatures and Various Strain Rates

The -10 and 70 °C tensile tests for the QP3Mn and TRIP7Mn at 0.0005 s\(^{-1}\) (except as noted for 70 °C in Section 3.1.1) and 0.1 s\(^{-1}\) were performed on the screw-driven Alliance frame with a Shepic 25.4 mm + 12.7 mm extensometer. The tests at 10 s\(^{-1}\) were performed on the high strain rate frame. An EXAIR® compressed air Adjustable Spot Cooler with two 25.4 mm flat nozzles provided the -10 °C condition. The spot cooler set-up on the high strain rate frame is shown in Figure 3.12a. The spot cooler was connected to the house compressed air, and the nozzles were aligned with both faces of the tensile specimen gage section. The temperature at either end of the gage section was measured to be 3 °C higher.
than the temperature in the center of the gage section. Two model 4085 SpotIR infrared heat lamps with OMEGA® CN79000 temperature controllers were used for the 70 °C condition. The infrared heat lamp set-up on the high strain rate frame is shown in Figure 3.12b. The temperature control for the infrared heat lamps required K-type thermocouples spot welded on each tensile specimen. These thermocouple wires were spot welded on both sides of the specimen in the center (widthwise and lengthwise) of the fillet region of the stationary grip. The temperature at either end of the gage section was measured to be 5 °C lower than the temperature in the center of the gage section. For each specimen tested at either temperature, K-type thermocouple wires were spot welded to the center of the gage section to make sure that the specimen reached within 70 – 72 °C or -10 – -12 °C prior to testing.

![Figure 3.12](image)

**Figure 3.12** (a) Compressed air spot cooler for low temperature testing on the high strain rate frame. (b) Infrared heat lamps for high temperature testing on the high strain rate frame.

### 3.4.7 Specimen Temperature Measurement During Tensile Testing

The surface temperature of the tensile specimen during tensile testing for selected conditions was recorded with K-type thermocouple wires that were spot welded to the specimen. The thermocouple wires were spot welded in the center, widthwise and lengthwise, of the gage section. For the strain rate range of 0.001 – 0.01 s⁻¹, temperature was recorded with an OMEGA® TC-08 thermocouple reader with a data acquisition rates up to 5 Hz. At 0.1 – 10 s⁻¹ strain rates, temperatures for the QP3Mn and TRIP7Mn were measured with a Vishay thermocouple reader capable of recording at 10,000 Hz.
Non-contact temperature measurements were made with a FLIR® A655sc infrared thermal camera at data acquisition rates up to 200 Hz. The calibration range of the camera was -40 to 150 °C. The maximum temperature that could be reported by the software was 160.2 °C, which was exceeded at the fracture surface in most cases. The gage section of the specimen was cleaned with acetone and then spray painted in matte black to maximize emissivity. Temperature measurements were made at 0.001 – 100 s⁻¹ strain rates.

3.4.8 Data Analysis

All tensile test data for all four materials were analyzed in terms of engineering stress and strain. Some of the tensile data were also converted to true stress, $\sigma$, and true strain, $\epsilon$, using Equations 3.4 and 3.5, where $s$ is engineering stress and $e$ is engineering strain [64].

$$\sigma = s(1 + e) \quad (3.4)$$

$$\epsilon = ln(1 + e) \quad (3.5)$$

The work hardening behaviors of the TRIP7Mn and QP3Mn steels were analyzed by fitting the Hollomon strain hardening model in Equation 3.6 to the true stress-strain curves. In Equation 3.6, $n$ is the Hollomon strain hardening exponent (a power law exponent) and $K$ is a constant [160].

$$\sigma = Ke^n \quad (3.6)$$

For the QP3Mn, the model was fit from 0.01 true strain up to necking. For the TRIP7Mn, the model was fit from 0.1 true strain up to necking. Examples of the model fit to the true stress-strain data are shown in Figure 3.13a for QP3Mn and Figure 3.13b for TRIP7Mn. For the majority of tests, the $R^2$ value was $\geq 0.98$ for the QP3Mn and $\geq 0.99$ for the TRIP7Mn, which suggests that the Hollomon strain hardening model is a reasonable model for the work hardening behavior of these steels. The TRIP7Mn engineering stress-strain curves have a sigmoidal shape for several of the different test conditions in this study, but the Hollomon strain hardening model is still a good fit (although mathematically it does not account for the sigmoidal curve shape) and can be used to assess how work hardening changes with strain rate and temperature. The QP3Mn tensile tests have very high initial work hardening rates following yielding (below 0.02 strain), which leads to the deviation of the model from the data at low strains.
Figure 3.13  True stress-strain curves for (a) QP3Mn and (b) TRIP7Mn showing the fit of the Hollomon strain hardening model to experimental data from tests at 0.0001 and 100 s\(^{-1}\). The solid lines are the experimental data, and the dashed lines are the model fits. The model was fit from 0.01 true strain up to necking for the QP3Mn and from 0.1 true strain up to necking for the TRIP7Mn.
CHAPTER 4
RESULTS AND DISCUSSION

The results for the four steels in this study will be presented in three separate sections for QP3Mn, TRIP7Mn, and QP980 and DP980. Each section contains micrographs and descriptions of the microstructure followed by a summary of the tensile data and retained austenite measurements as a function of strain at various strain rates. For the QP3Mn and TRIP7Mn, the tensile test results include tests at room temperature and various strain rates, at a quasi-static strain rate and various temperatures, and at different temperatures and strain rates. For the QP980 and DP980, all of the tensile tests were performed at room temperature and at various strain rates. A small pre-strain study is included in the QP980 and DP980 tensile test results.

4.1 QP3Mn Results

A representative FE-SEM micrograph of the QP3Mn microstructure is shown in Figure 4.1. The etched microstructure in Figure 4.1 is consistent with a mostly martensitic microstructure with interspersed lath-like austenite [28]. An EBSD image quality map, inverse pole figure, and phase map are shown in Figure 4.2. The image quality map in Figure 4.2a shows the finely spaced boundaries of lath martensite [161]. The inverse pole figure in Figure 4.2b shows that the martensite is present in many different orientations and does not have a strong texture. The phase map in Figure 4.2c shows austenite throughout the microstructure.

Figure 4.1 FE-SEM image of the QP3Mn microstructure. Etched with 2% nital for 15 s.
as thin green lines (film austenite), and some of the austenite has a blocky morphology and appears as fine grains. The austenite content from the EBSD phase map is 9 vol.% (based on all data points). The average retained austenite content measured by XRD from four as-processed specimens is 14 ± 1 vol.%. The austenite content from XRD is higher than the amount from EBSD, because austenite films are sometimes poorly resolved with EBSD due to their small thickness dimension. The thickness of the films can be close to the EBSD spot size, e.g. films with thicknesses on the order of 100 nm were observed by TEM in Q&P steels [37, 39]. The martensite volume fraction formed during the first quench step to 180 °C is estimated to be 85 vol.% as calculated by the Koistinen-Marburger Equation 2.2 using the average $M_s$ temperature from dilatometry. The estimated martensite volume fraction and the experimentally measured average martensite volume fraction of 86 vol.% (by XRD) are very close, which may indicate that most of the austenite was stabilized during partitioning and very little or no fresh martensite formed during the quench to room temperature.

Figure 4.2 EBSD scan of the QP3Mn as-heat treated microstructure: (a) image quality map, (b) inverse pole figure map, and (c) phase map. (Color image - see PDF version)
4.1.1 QP3Mn Tensile Properties and Retained Austenite Measurements at Various Strain Rates

Example engineering stress-strain curves for QP3Mn tested at room temperature (22 – 26 °C) in air and at various strain rates are shown in Figure 4.3a. All of the curves exhibit continuous yielding with initially high work hardening rates post-yielding. Beyond approximately 2% strain, the curves have low work hardening rates, and UTS values are high (greater than 1500 MPa). There is almost no change in UTS with increasing strain rate, but the low strain continuous yielding region is more abrupt and occurs at higher flow stresses (leading to higher yield strength) for the 1 and 10 s\(^{-1}\) strain rates. As strain rate increases from a quasi-static rate, the specimen temperature at UTS progressively increases as shown by Figure 4.3b. For tests at 0.1 s\(^{-1}\) and greater strain rates, the specimen temperature at UTS is approximately constant with increasing strain rate, which indicates that adiabatic heating conditions are reached at approximately 0.1 s\(^{-1}\). The specimen temperature is nearly constant in the range of 0.1 – 10 s\(^{-1}\), because the area under the stress-strain curve up to UTS is similar for those strain rates. The amount of self-heating at UTS is predicted from the area under the stress-strain curve for strain rates up to 100 s\(^{-1}\) in Section 5.1.1. The temperature data in Figure 4.3a shows the variation of stress-strain properties with strain rate for QP3Mn.

![Figure 4.3](image-url)

Figure 4.3 (a) Representative engineering stress-strain curves for QP3Mn tested at room temperature (22 – 26 °C) in air at various strain rates. (b) QP3Mn specimen temperature at UTS measured with a K-type thermocouple (temperature could only be measured at 10 s\(^{-1}\) or lower strain rates). The initial room temperature was higher by 3 °C for the high strain rate tests due to ambient temperature differences in the laboratories.
Figure 4.3b shows that the specimen temperature increases by approximately 35 °C by the
time UTS is reached at the high strain rates (0.1 s\(^{-1}\) and higher). The increase in specimen
temperature provides a small thermal contribution that assists dislocations in overcoming
short-range barriers (which should lead to a decrease in flow stress), but the tensile data
from quasi-static tensile tests at elevated temperature presented in Section 4.1.2 suggest
that this temperature rise may not produce an observable softening effect on flow stress.
The specimen temperature increase plotted as a function of strain with the corresponding
stress-strain curve for select low strain rate conditions are provided in Appendix B. Thermal
camera images from a test at 1 s\(^{-1}\) are also provided in Appendix B. The thermal data in
Appendix B shows the continuous heating of the tensile specimen gage length throughout
uniform deformation.

Representative true stress-strain curves plotted up to uniform elongation from the
QP3Mn room temperature tensile tests in air at various strain rates are shown in Figure
4.4a. The curves are offset by 0.06 strain for clarity. The average true YS, UTS, and UE
for all of the strain rates tested are summarized in Appendix C. The work hardening rate
versus true strain for four of the true stress-strain curves are plotted in Figure 4.4b. At
all strain rates, the QP3Mn exhibits very high work hardening rates immediately following
yielding (over 40,000 MPa at 0.01 or less strain), and the work hardening rates rapidly drop
to 2000 MPa near UTS (between 0.07 and 0.1 strain). Between 0.01 and 0.03, the specimen
tested at 0.001 s\(^{-1}\) had the highest work hardening rates. Above 0.03, the work hardening
rates were very close for the tests at 0.001, 0.1, and 100 s\(^{-1}\). The specimen tested at 10 s\(^{-1}\)
exhibited the lowest work hardening rates throughout plastic deformation among the four
strain rates. The Hollomon strain hardening exponent \(n\) was calculated from 0.01 true strain
up to necking and is shown as a function of strain rate in Figure 4.4c (\(n\) was averaged from
three tests at each strain rate). Examples of the curve fitting to calculate \(n\) are provided in
Figure 3.13a in Chapter 3. There is an overall decreasing trend in \(n\) from 0.0001 s\(^{-1}\) to 1 s\(^{-1}\).
The lowest \(n\) values occur for 1 s\(^{-1}\) and 10 s\(^{-1}\), and then above 10 s\(^{-1}\) the \(n\) values are slightly
higher. The decrease in \(n\) means that the overall work hardening decreases with increasing
strain rate up to 1 s\(^{-1}\). One factor that could affect work hardening behavior is the specimen
self heating at high strain rates. The amount of specimen self heating significantly increases
between 0.0001 s\(^{-1}\) to 0.1 s\(^{-1}\), and the heating may have a softening effect on flow stress
that contributes to the decrease in work hardening with increasing strain rate. The work
hardening behavior is similar at 1 and 10 s\(^{-1}\). Above 10 s\(^{-1}\), the values of \(n\) are the same or
close to the values of \(n\) at 0.1 s\(^{-1}\) and lower strain rates. The positive strain rate sensitivity
above 10 s\(^{-1}\), which can be observed by the higher flow stresses at strain rates above 10 s\(^{-1}\)
in the true stress-strain curves in Figure 4.4a, is associated with greater work hardening in
Figure 4.4  (a) Example true stress-strain curves for QP3Mn tested at room temperature (22 – 26 °C) in air at various strain rates. The curves are offset by 0.06 strain.  
(b) Work hardening rate \((d\sigma/d\varepsilon)\) versus total true strain for four strain rates. 
(c) The Hollomon strain hardening exponent, \(n\), as a function of engineering strain rate for QP3Mn.
tensile tests at 30 – 200 s\(^{-1}\) than in tests at 1 and 10 s\(^{-1}\). Another factor that may influence the trend in \(n\) is the flow stress strain rate sensitivity from yielding to UTS. Yield strength is usually more sensitive to strain rate than UTS for AHSS [143], which means that as strain rate increases the YS increases more rapidly than the UTS. The difference between YS and UTS would decrease with increasing strain rate and could lead to a decrease in the calculated value of \(n\) with increasing strain rate.

The average engineering UTS and 0.2% offset YS data are plotted versus strain rate in Figure 4.5a. The uncertainty of the flow stress measurement is estimated to be 7 MPa. There is significant scatter in the YS data that obscures the trends in YS with respect to strain rate. From 0.0005 to 0.1 s\(^{-1}\), the error bars overlap, and the strain rate sensitivity appears to be low. However, the difference in the average YS between 0.0001 and 1 s\(^{-1}\) is 217 MPa, which is a strength increase of 20%. The YS is nearly constant between 1 and 10 s\(^{-1}\), and then the average YS drops between 10 and 30 s\(^{-1}\) before increasing up to 200 s\(^{-1}\) (although the errors bars overlap between 30 and 200 s\(^{-1}\)). The difference in the average YS between 0.001 and 100 s\(^{-1}\) is 136 MPa, which is an overall strong, positive strain rate sensitivity between low and high strain rates. This change in YS can be compared to the changes in YS for the austenitic stainless steels and martensitic steels in Figures 2.16c and 2.16d. For the same strain rate range of 0.001 and 100 s\(^{-1}\), the 304L stainless steel average YS increased by 131 MPa and the 309 stainless steel average YS increased by 181 MPa, while the as-quenched 0.2 wt.% C steel YS increased by approximately 100 MPa and the 0.2 wt.% C steel quenched and then tempered at 400 °C for one hour had no change in YS [9, 141]. The change in YS for QP3Mn between 0.001 and 100 s\(^{-1}\) is closer to the behavior of austenitic stainless steel than as quenched or quenched and tempered martensitic steel. There are some major differences between austenitic stainless steel, fully martensitic steel, and Q&P steel that affect the strain rate sensitivity. The austenitic stainless steel contains austenite that is heavily alloyed and present as large, equiaxed grains [9]; whereas, the austenite in the Q&P steel is less alloyed and is present as very fine laths or blocky grains. For the 0.2 wt.% C steel, the as-quenched martensite has a high dislocation density and a relatively large amount of carbon in solution, and tempering the martensite at 400 °C for one hour causes carbide precipitation and elimination of low angle boundaries [24, 141]. The partitioned martensite in the QP3Mn is expected to have a low amount of solute carbon, few precipitates, and potentially less recovery compared to the tempered martensite. In contrast to the YS behavior, there is very little change in UTS with increasing strain rate. At strain rates below 1 s\(^{-1}\), there is almost no strain rate sensitivity, and there is a slight positive strain rate sensitivity between 1 and 30 s\(^{-1}\) with no significant change in UTS between 30 and 200 s\(^{-1}\).
Figure 4.5 Average tensile mechanical properties of QP3Mn versus strain rate: (a) UTS and 0.2% YS and (b) TE and UE. The error bars represent one standard deviation for three tests.

The large difference in strain rate sensitivities between the YS and UTS at higher strain rates may be due to a few factors. At the YS, most of the retained austenite is untransformed. The austenite is expected to have greater strain rate sensitivity than the martensite, which tends to have very little strain rate sensitivity due to the pervasiveness of long-range dislocation interactions. The initial austenite content is only 14 vol.%; therefore, the positive YS strain rate sensitivity must also be related to the behavior of the partitioned martensite (e.g. the partitioned martensite may have low positive strain rate sensitivity and the austenite may have high positive strain rate sensitivity, and the composite effect is an overall positive strain rate sensitivity that is between the strain rate sensitivities of the individual constituents). At the UTS, the microstructure is almost completely martensitic and is deformed, which are both associated with substantial long-range dislocation interactions during deformation leading to low strain rate sensitivity. Another difference between YS and UTS at high rates is temperature. The specimen is close to room temperature at the 0.2% YS, and it increases in temperature by approximately 35 °C at the UTS at high strain rates. As discussed previously, this temperature rise may not produce an observable softening effect on flow stress, but the thermal contribution does aid dislocations in overcoming short-range barriers, while increasing strain rate has the opposing effect of reducing the time for dislocations to overcome short-range barriers. The possibility of entering a DSA regime is also an important consideration for assessing the effect of adiabatic heating on flow stress, because entering a DSA regime would result in an increase rather than a decrease in flow stress. As
will be discussed for the quasi-static, elevated temperature tensile test data in Section 4.1.2. QP3Mn may enter a DSA regime near 70 °C. Based on the quasi-static, elevated temperature test data, DSA is not a factor in deformation behavior in the temperature range 22 – 55 °C.

The average engineering UE and TE are plotted versus strain rate in Figure 4.5b. The UE and TE for each test were both measured from the stress-strain curve (at strain rates above 0.1 s\(^{-1}\) a portion of the elongation data must be extrapolated from the strain gage data for each test). The uncertainty of the elongation measurement is estimated to be 0.1%. Both UE and TE decrease with increasing strain rate up to 1 s\(^{-1}\) and then increase at the higher strain rates. The trend in the UE is similar to the trend in \(n\) with respect to strain rate. The decrease in elongation on the order of 3% strain between 0.0001 s\(^{-1}\) and 1 s\(^{-1}\) is attributed to the lower work hardening rates and thus instability being reached sooner at the higher strain rates. The UE reaches a minimum at 1 s\(^{-1}\) and then increases up to 100 s\(^{-1}\). The only significant increase in average \(n\) value is between 10 and 30 s\(^{-1}\). The steady increase in average UE by a total of 3% over the strain rate range 1 – 100 s\(^{-1}\) may be due to strain rate hardening that causes a resistance to necking and leads to an increase in uniform elongation [14, 162]. In the strain rate range 0.001 – 200 s\(^{-1}\), a decrease in elongation to a minimum near 10 s\(^{-1}\) followed by an increase in elongation with increasing strain rate was also observed for a few first generation TRIP steels [14, 146]. The average value of \(n\) decreases slightly between 100 and 200 s\(^{-1}\), and the average elongation also decreases.

One contribution to the high amount of scatter particularly in the YS (e.g. 1158 ± 92 MPa at 0.01 s\(^{-1}\)) and elongation (e.g. 12.9% ± 1.2% TE at 0.0005 s\(^{-1}\)) data could be slight differences between heat treatment batches. The tensile specimens were heat treated in batches of three, and the tensile tests that were performed at each strain rate came from different batches. While the heat treatments may be a contributing factor to the scatter, there were no distinct batch-based trends observed (e.g. only one of three specimens from a single batch would have a different YS compared to the specimens from other batches).

The retained austenite volume fraction as a function of strain at 0.0005, 0.1, 10, and 100 s\(^{-1}\) is shown in Figure 4.6a. The uncertainty in the austenite measurement by XRD is estimated to be 1 vol.%. An example of how the X-ray peak intensities change with increasing strain for the 0.0005 s\(^{-1}\) condition is provided in Appendix D. There is no significant difference in the austenite transformation rate between 0.0005 s\(^{-1}\) and 100 s\(^{-1}\). Specimen temperatures as a function of strain up to UTS for various strain rates up to 0.1 s\(^{-1}\) are shown in Figure 4.6b. The heating profiles at 10 s\(^{-1}\) and 100 s\(^{-1}\) are expected to be close to the 0.1 s\(^{-1}\) data, because the areas under the stress-strain curves are similar at the high strain rates. Most of the austenite transformation occurs by 0.04 strain, and at 0.1 s\(^{-1}\) the specimen temperature only
heats up to approximately 40 °C at that strain level. The effect of temperature on austenite stability in QP3Mn is discussed in the context of the quasi-static, elevated temperature tensile tests in Section 4.1.2. Adiabatic heating increases the specimen temperature at UTS up to around 60 °C at 0.1 s\(^{-1}\), but austenite stability is the same between 0.0005 and 0.1 s\(^{-1}\) (e.g. the austenite content is the same at the two strain rates near 0.08 strain). The retained austenite stability is unaffected by the increase in specimen temperature with increasing strain during the QP3Mn high strain rate tensile tests. The apparent lack of an effect of adiabatic heating on the austenite to martensite transformation rate will be discussed further in Sections 4.1.2 and 5.1.2.

**Figure 4.6** (a) Retained austenite as a function of plastic engineering strain for QP3Mn tested at room temperature (22 – 26 °C) in air at various strain rates. (b) Specimen temperature as a function of engineering strain up to UTS measured with a K-type thermocouple.

In summary, the QP3Mn has an overall positive YS strain rate sensitivity in the range 0.0001 – 200 s\(^{-1}\), although there is significant scatter in the YS data (especially in the data sets from 0.0005 – 0.1 s\(^{-1}\) for which the error bars overlap). The austenite strain rate sensitivity may be a strong contributor to the positive YS strain rate sensitivity. The QP3Mn has almost no UTS strain rate sensitivity up to 1 s\(^{-1}\) and a slight positive strain rate sensitivity up to 200 s\(^{-1}\). The very low UTS strain rate sensitivity is consistent with long-range dislocation interactions dominating the flow stress behavior in the mostly martensitic microstructure. An increase in specimen self-heating with increasing strain rate up to 0.1 s\(^{-1}\) may contribute to lack of UTS strain rate sensitivity at strain rates up to 0.1 s\(^{-1}\). Elongation decreases with
increasing strain rate up to 1 s$^{-1}$ and then increases with increasing strain rate. The work hardening rate also decreases up to 1 s$^{-1}$ and is slightly higher above 10 s$^{-1}$. The austenite to martensite transformation rate is independent of strain rate up to 100 s$^{-1}$.

4.1.2 QP3Mn Tensile Properties and Retained Austenite Measurements at Various Temperatures

Figure 4.7a shows example engineering stress-strain curves for QP3Mn tensile tests at temperatures between -10 and 70 °C and at a quasi-static strain rate of 0.0005 s$^{-1}$. The flow curve at -10 °C exhibits the greatest amount of work hardening, and the work hardening rate is significantly higher than the work hardening rates observed at elevated test temperatures. The flow curves are very similar for 22, 40, and 55 °C in terms of yielding, work hardening behavior, and elongation. The 70 and 85 °C curves are similar to the lower temperature curves in terms of yielding and work hardening behavior up to UTS. In contrast to the lower temperature curves, the 70 and 85 °C curves exhibit less uniform elongation and show signs of discontinuous flow behavior during necking. The flow curves from the 70 and 85 °C tests with the necking region magnified are shown in Figure 4.7b. In the prolonged necking region, there are serrations that are low in magnitude and frequency in the flow curve. The serrations could be due to multiple necks forming along the gage length, although only one necked region was clearly observed post-fracture. A Q&P development study suggested the possibility of DSA for subtle serrated plastic flow behavior at room temperature observed for a few heat treatment conditions [43]. The serrations could also be related to multiple localized austenite transformation events (more than half of the austenite is untransformed when uniform elongation is reached). The possibility of austenite with widely varying stabilities at elevated temperature will be discussed in the context of the transformation behavior in Section 5.2.

The UTS and 0.2% offset YS are plotted versus temperature in Figure 4.8a. There is a large amount of scatter in the data at -10 and 22 °C (e.g. YS values for the three tests at -10 °C are 1092, 1153, and 1230 MPa and the YS values for the two tests at 85 °C are 1190 and 1195 MPa), but overall there is no significant effect of temperature on the YS or UTS for the QP3Mn steel for temperatures between 22 and 85 °C. The high scatter in the -10 °C data may be due in part to the cooling of the specimen with compressed air as opposed to the more uniform temperature environment of the oil bath for the elevated temperature tests. There is a decrease in UTS between -10 and 22 °C, which arises due to the differences in work hardening behavior (i.e. differences in the slope (ds/de) of the flow curves shown in Figure 4.8a). The UTS values at the elevated temperatures (> 22 °C) are all in the
Figure 4.7  (a) Representative engineering stress-strain curves for QP3Mn at various temperatures and a quasi-static strain rate. (b) Tensile curves for the 70 and 85 °C tests (only plotted above 1200 MPa) showing the serrated flow during necking.

range of 1558 – 1577 MPa (only 19 MPa difference for the entire range). The flow stress behavior of the QP3Mn is dominated by the mostly martensitic microstructure. Although the martensite has undergone a short partitioning heat treatment, which would reduce the carbon content and likely result in some recovery, the martensite deformation behavior is expected to be dominated by long-range dislocation interactions. Temperature sensitive short-range dislocation barriers do not appear to have a significant impact on flow stress for two reasons: 1) their contribution to flow stress is very small compared to long-range barriers and 2) the test temperature range is small with a maximum increase from room temperature of 63 °C to the highest test temperature, which amounts to a very small increase in the
thermal contribution for aiding dislocations to overcome short range barriers. Finally, the serrated flow behavior observed for the 70 and 85 °C tests may be due to DSA (possibly the C-Mn couple mechanism discussed in Section 2.3) [76, 83]. The onset of DSA with increasing temperature is associated with an increase in UTS and elongation [83]. Therefore, DSA may be factor in the UTS and elongation data for the 70 and 85 °C tests.

The UE and TE are plotted versus temperature in Figure 4.8b. The UE declines slightly with increasing temperature for a total average decrease of 3% strain from -10 to 85 °C. Most of the UE decline occurs between -10 and 55 °C, and the UE is similar between 70 and 85 °C. The TE also declines on average by 3% strain from -10 to 55 °C, after which the TE appears to increase with increasing temperature for some of the test specimens implying that some of the tests showed post-uniform elongation increases at the higher test temperatures. The increase in TE and similar UE at 70 and 85 °C coincides with the appearance of serrations in the flow curves.

![Graphs showing tensile mechanical properties of QP3Mn versus temperature](image)

Figure 4.8 Tensile mechanical properties of QP3Mn versus temperature: (a) UTS and 0.2% YS and (b) TE and UE.

Retained austenite volume fractions versus strain from tensile tests at a quasi-static strain rate of 0.0005 s⁻¹ and temperatures in the range of -10 – 70 °C are shown in Figure 4.9. For all test temperatures, the retained austenite volume fraction decreases with increasing strain, but that austenite to martensite transformation rate decreases with increasing test temperature (i.e. austenite becomes more stable with increasing test temperature). The QP3Mn austenite stability increases with increasing temperature, but the effects of the
austenite stabilization on mechanical properties are not clear. The slight decline in the elongation data up to 55 °C suggests that the austenite stabilization could contribute to reduced ductility (the greater the extent of transformation the more instability is delayed). However, the austenite stabilization does not cause a major reduction in UTS with increasing temperature for most of the temperature range. The only decrease in UTS occurs between -10 and 22 °C (an average decrease of 57 MPa), but the austenite to martensite transformation rate at -10 °C is very similar to the transformation rate at 22 °C despite a 32 °C difference in temperature. The average UE decreases from 10% to 9% strain between -10 and 22 °C. The temperature increase between 22 °C and 55 °C is 33 °C, but the temperature effects on austenite stability and mechanical properties are very different from the 32 °C increase in temperature between -10 and 22 °C. The austenite is significantly more stable at 55 °C than at -10 and 22 °C. For example, near 0.08 strain the amount of austenite at -10 °C is close to 4 vol.% and at 22 °C is just under 5 vol.%, whereas nearly 8 vol.% austenite remains at 55 °C. There is no significant difference in average UTS between 22 and 55 °C, but the average UE decreases from 9% to 8%. The austenite to martensite transformation rates are nearly the same at 70 and 85 °C, and the UTS and elongation data are also very close between 70 and 85 °C. In a study by Feng and Speer, tensile tests of TRIP780 (initially 16 vol.% austenite) and QP980 (initially 12 vol.% austenite) were performed at 20, 40 and 60 °C with a constant 0.0006 s⁻¹ strain rate [163]. Austenite stability increased with increasing temperature for both steels, although the transformation rate was nearly the same at 40 and 60 °C for the QP980. For the TRIP780, UTS decreased from 834 to 768 MPa (66 MPa difference) and UE increased from 27% to 37% (10% strain difference) with increasing
temperature from 20 to 60 °C. However, for the QP980, UTS only decreased from 1006 to 978 MPa (28 MPa difference) and UE decreased from 17% to 15% (2% strain difference) with increasing temperature from 20 to 60 °C. The low sensitivity of UTS to temperature and the decrease in UE with increasing temperature observed for the QP980 are similar to the behavior observed for QP3Mn.

In Figure 4.9, there appears to be a change in austenite transformation behavior between 22 and 40 °C. At -10 and 22 °C, the austenite transforms steadily with increasing strain. At 40 °C and above, the amount of austenite at 0.01 and 0.02 strain is constant. At 70 and 85 °C, approximately 2 vol.% austenite transforms prior to 0.01 strain, followed by the austenite being relatively stable between 0.01 and 0.04 plastic strain, and then the austenite content decreases with increasing strain. This behavior may be related to differences in austenite stabilities within the microstructure at elevated temperature and will be discussed further in Section 5.2.

The results of the tensile tests at various temperatures and a quasi-static strain rate provide insight to the role of adiabatic heating during high strain rate deformation, although the tests are not directly comparable because temperature changes as a function of strain throughout a high strain rate test and both temperature and strain rate affect deformation mechanisms. Figure 4.6b shows that the specimen temperature at 0.1 s⁻¹ is approximately 40 °C at 0.04 strain and is 61 °C at UTS (heating profiles at higher strain rates are measured or estimated to be similar as discussed in Section 5.1.1). The YS and UTS of the QP3Mn steel at 0.0005 s⁻¹ are insensitive to test temperature in the range of 22 to 70 °C. The temperature insensitivity of flow stress at a quasi-static strain rate suggests that adiabatic heating likely does not have a major softening effect on flow stress at high strain rates. There is a slight decline in UE between 22 and 70 °C for the quasi-static tensile tests and a slight decline in UE between the 0.0001 and 1 s⁻¹ tests. It is not clear that adiabatic heating is linked to the decline in UE with increasing strain rate, because UE also decreases between 0.1 and 1 s⁻¹ and the amount of heating at UTS is nearly the same. Additionally, the UE for the quasi-static tests at elevated temperature may be affected by the increase in austenite stability with increasing temperature, though the increase in austenite stability does not appear to have a significant effect on UTS. As mentioned in the discussion of the austenite versus strain data at various strain rates (Section 4.1.1), adiabatic heating does not have an observable stabilizing effect on the austenite at strain rates up to 100 s⁻¹. Most of the austenite to martensite transformation occurs by 0.04 strain, and the specimen temperature at 0.1 s⁻¹ and higher strain rates is close to 40 °C at 0.04 strain. Figure 4.9 shows only a slight increase in austenite stability between 22 and 40 °C (the difference between the two test temperatures
near 0.08 strain is approximately 1 vol.% austenite). Therefore, below 0.04 strain there is not enough specimen self-heating to have a significant effect on austenite stability. There is a much larger difference in austenite stability between the quasi-static 22 and 55 °C data (the difference between the two test temperatures near 0.08 strain is approximately 3 vol.% austenite). However, adiabatic heating above 40 °C during high strain rate tensile tests does not have an observable stabilizing effect on austenite. The dynamic temperature change during high strain rate tests does not have the same stabilizing effect as higher temperatures during isothermal, quasi-static tensile tests.

In summary, the QP3Mn YS, UTS, and work hardening are fairly insensitive to temperature in the range of 22 – 85 °C. Between -10 and 22 °C, the YS does not change significantly, but the UTS decreases due to a decrease in the work hardening rate. The UE and TE decrease slightly with increasing temperature, which may be related to stabilization of austenite. The austenite to martensite transformation rate decreases with increasing temperature for the QP3Mn, and the austenite transformation behavior is very similar at 70 and 85 °C.

### 4.1.3 QP3Mn Tensile Properties and Retained Austenite Measurements at Various Strain Rates and Temperatures

The mechanical properties and retained austenite measurements as a function of strain for the tensile tests performed at -10, 22, and 70 °C and at 0.0005, 0.1, and 10 s\(^{-1}\) strain rates are summarized in Figure 4.10. The average engineering 0.2% offset YS and UTS versus strain rate for all three test temperatures are shown in Figure 4.10a. The error bars (representing one standard deviation) of the YS data overlap for all three test temperatures at each of the three strain rates, which indicates that the difference between the average YS at all three tests temperatures is not statistically significant for each strain rate. Therefore, there are no significant differences in the YS trends with respect to strain rate for the three test temperatures. At all three temperatures, measured YS values between 0.0005 and 0.1 s\(^{-1}\) are essentially constant, and the average YS increases between 0.1 and 10 s\(^{-1}\) (positive YS strain rate sensitivity).

At all three temperatures, measured UTS values between 0.0005 and 0.1 s\(^{-1}\) are also essentially constant, but there are differences in the UTS strain rate sensitivities among the three test temperatures between 0.1 and 10 s\(^{-1}\). The UTS strain rate sensitivity is higher at -10 °C than 22 °C between 0.1 and 10 s\(^{-1}\). An increase in strain rate sensitivity with a decrease in test temperature is expected, because the decrease in test temperature reduces the thermal assistance to dislocations for overcoming short-range barriers. With decreasing
temperature, the thermal component of flow stress increases and consequently, strain rate increases. An increase in strain rate sensitivity with decreasing test temperature also occurs for mild steel [138]. Between 0.1 and 10 s\(^{-1}\), the positive UTS strain rate sensitivity is higher at 70 °C than 22 °C. The austenite is more stable at 70 °C than 22 °C, especially near UTS (as shown in Figure 4.10b of austenite versus strain). The higher quantity of austenite in the microstructure may contribute to the higher UTS strain rate sensitivity at 70 °C than at 22 °C between 0.1 and 10 s\(^{-1}\). It is unclear why the UTS strain rate sensitivity is slightly negative between 0.0005 and 0.1 s\(^{-1}\) at 70 °C but is close to zero for the other test temperatures.

Retained austenite measurements as a function of strain at 0.0005 and 10 s\(^{-1}\) for the three test temperatures are shown in Figure 4.10b. For all three test temperatures, the austenite to martensite transformation rate was similar at the low and the high strain rates. The austenite is significantly more stable at 70 °C than at 22 °C with just over 8 vol.% austenite remaining close to 0.08 strain at 70 °C compared to just under 5 vol.% remaining near 0.08 strain at 22 °C. In contrast, the austenite stability is only slightly lower at -10 °C compared to 22 °C (e.g. there is 4 vol.% austenite at -10 °C and just under 5 vol.% austenite at 22 °C close to 0.08 strain at 0.0005 s\(^{-1}\)), and the differences between the data sets are within the uncertainty of the XRD measurements. Therefore, the strain-induced austenite transformation has little or no impact on the differences in average UTS values between -10 and 22 °C at different strain rates. The stabilization of austenite at 70 °C does not appear to increase UTS strain rate sensitivity, although the stabilized austenite is expected to have greater strain rate sensitivity than fresh martensite.

Figure 4.10c shows the average engineering UE versus strain rate and Figure 4.10d shows the average engineering TE versus strain rate for all three test temperatures. Within the scatter in the UE and TE data in Figures 4.10c and 4.10d, there are no major changes in UE and TE with respect to strain rate for each temperature (there is a slight decrease in UE between 0.001 and 0.1 s\(^{-1}\) for 22 °C, but otherwise the error bars overlap between the three strain rates for each test temperature). However, both UE and TE decrease with increasing temperature. Elongation is more sensitive to changes in test temperature than changes in applied strain rate. The trends in elongation with respect to temperature and strain rate are associated with austenite stability. Austenite stability increases with increasing test temperature, and elongation decreases with increasing temperature. The austenite to martensite transformation rate is similar at low and high strain rates (meaning austenite stability is similar at low and high strain rates) for all three test temperatures, and elongation does not change significantly with increasing strain rate.
Figure 4.10  Average tensile mechanical properties of QP3Mn versus strain rate for three different test temperatures: (a) UTS and 0.2% YS, (c) UE and (d) TE. The error bars represent one standard deviation for three tests. (b) Retained austenite as a function of plastic engineering strain at a low and a high strain rate for three different test temperatures.

In summary, test temperature (-10 to 70 °C) overall has little effect on the strength and ductility trends with respect to strain rate for QP3Mn. However, the YS and UTS strain rate sensitivities between 0.1 and 10 s⁻¹ are slightly higher at -10 °C compared to 22 °C, and the YS and UTS strain rate sensitivities are slightly negative between 0.0005 and 0.1 s⁻¹ at 70 °C. Temperature had a greater effect than strain rate on elongation, which is likely
due austenite stability increasing with increasing temperature. The austenite to martensite transformation rate is not significantly affected by strain rate at any test temperature.

4.2 TRIP7Mn Results

The TRIP7Mn microstructure is shown in the image quality map, inverse pole figure map, and the phase map from an EBSD scan in Figure 4.11. The image quality map in Figure 4.11a shows the equiaxed ferrite and austenite grains. The inverse pole figure map in Figure 4.11b shows a spread in the crystal plane normal directions perpendicular to the sheet surface normal direction between <001> and <111> with a strong concentration of orientations close to <111> indicated by the high concentration of red, pink, purple, and especially blue ferrite grains for ferrite. The austenite grains in Figure 4.11b are more randomly oriented than the ferrite grains, which is shown by the greater color variation in the austenite grains. Figure 4.11c shows that the ferrite and austenite grains are similar in size and morphology and are interspersed throughout the microstructure. The ferrite grains have a grain size of 1.2 ± 0.5 μm. The austenite grains have a grain size of 0.8 ± 0.3 μm. The retained austenite content measured from EBSD is 43 vol.%, and the average amount measured from XRD for four as-received specimens is 39 ± 2 vol.%.

4.2.1 TRIP7Mn Tensile Properties and Retained Austenite Measurements at Various Strain Rates

Example engineering stress-strain curves for TRIP7Mn tested at room temperature (22 – 26 °C) in air at various strain rates are shown in Figure 4.12a. All of the curves exhibit YPE up to 8% or greater strain. Recent TEM observations during YPE of a similar 7 wt.% Mn TRIP steel showed that the ferrite has a very low initial dislocation density after annealing, and it is likely that the ferrite experiences static strain aging at room temperature, which both contribute to yielding by the nucleation and propagation of Lüders bands [62]. Additionally, fine grain size is associated with high Lüders strain during yielding in steel alloys [53]. There is no change observed in the yielding behavior up to 0.1 s⁻¹. Above 0.1 s⁻¹, the upper and lower YS increase. With increasing strain rate up to 10 s⁻¹, both UTS and ductility decrease, and UTS and elongation gradually increase with increasing strain rate above 10 s⁻¹.

Figure 4.12b shows a magnification of the serrations at 0.0005, 0.01, and 0.1 s⁻¹. At strain rates below 1 s⁻¹, the TRIP7Mn undergoes discontinuous plastic deformation after the YPE region giving the appearance of serrations in the flow curve. At 0.0005 s⁻¹, there are very fine
serrations up to approximately 25% strain, and then the serrations have the appearance of long linear segments at higher strain. For the test at 0.01 s$^{-1}$, the very fine serrations are not evident, and the serrated flow curve has the appearance of long, linear segments. By 0.1 s$^{-1}$, the serrations are nearly indistinguishable in the flow curve. The diminishing magnitude and frequency of the flow curve serrations with increasing strain rate is consistent with the behavior of materials exhibiting DSA [86]. As discussed in Section 2.3, the serrated flow behavior may be due to DSA, and it will be discussed in the context of DSA caused by C-Mn couples in the austenite [76, 83]. The lack of distinct serrations in the flow curves at 1 s$^{-1}$ and above indicates a transition to uniform plastic deformation of the gage length. Figure 4.12c shows the specimen temperature measured by a thermocouple in the center of the gage length as a function of strain for strain rates from 0.0005 to 1 s$^{-1}$. For 0.0005, 0.001, and
Figure 4.12  (a) Representative engineering stress-strain curves for TRIP7Mn tested at room temperature (22 – 26 °C) in air at various strain rates. (b) Engineering stress-strain curves for TRIP7Mn at low strain rates (only plotted above 700 MPa) showing the serrations in the flow curves. (c) Specimen temperature as a function of engineering strain up to UTS measured with a K-type thermocouple.

0.01 s⁻¹, the specimen temperature undergoes repeated heating and cooling, which gives the temperature profile a serrated appearance. The repeated heating and cooling is interpreted as being caused by discontinuous plastic flow. The thermocouple measurements at 0.1 s⁻¹ have a step-like appearance, because discontinuous plastic deformation occurs without enough time for the specimen to cool when the deformation is occurring away from the thermocouple.
The temperature profile from the thermocouple is smooth at 1 s\(^{-1}\), indicating uniform plastic deformation. The noise in the load data and the appearance of very fine oscillations at high strain rates (1 s\(^{-1}\) and above) are artifacts from high strain rate tensile testing and not indicators of serrated plastic flow. Thermal camera images provide supporting evidence that the propagation of PLC bands only occurred at strain rates below 1 s\(^{-1}\), and images from a test at 1 s\(^{-1}\) are provided in Appendix B. The specimen temperature increase plotted as a function of strain with the corresponding stress-strain curve for select low strain rate conditions are also provided in Appendix B.

Figure 4.13 shows the retained austenite volume fraction as a function of plastic strain after YPE for TRIP7Mn tested at room temperature (22 – 26 °C) in air at various strain rates (austenite was not measured during YPE). An example of how the X-ray peak intensities change with increasing strain for the 0.0005 s\(^{-1}\) condition is provided in Appendix D. The uncertainty in the austenite measurement by XRD is estimated to be 1 – 2 vol.%. The retained austenite amount drops from 40 vol.% in the as-received material to 22 vol.% post-YPE at 8% plastic strain. In Figure 2.6, the stress-strain curves and austenite as a function of strain are shown for a similar 7 wt.% Mn TRIP steel that was processed at different intercritical annealing temperatures [48]. For the 600 °C condition, very little austenite transformed during YPE, but the austenite produced by annealing at 625 °C condition exhibited lower stability and more than a quarter of the austenite transformed during YPE. The austenite stability determines how much austenite will transform during YPE. The transformation of approximately 18 vol.% austenite during YPE for the TRIP7Mn does not cause an increase in stress. An explanation for the transformation of austenite at constant stress is that it is a stress-assisted transformation rather than a strain-induced transformation [81, 91, 164]. The strain-induced transformation of austenite to martensite beyond YPE contributes to a high work hardening rate and high elongation [48]. The austenite to martensite transformation rate is similar at 0.1, 10, and 100 s\(^{-1}\), and it is reduced at these higher strain rates compared to 0.0005 s\(^{-1}\). The difference in the austenite volume fractions between 0.0005 s\(^{-1}\) and the three higher strain rates increases with increasing strain. For example, in Figure 4.13 the difference between quasi-static and high strain rates at 12% strain is 3 vol.% austenite and the difference at 30% strain is over 5 vol.% austenite.

The decrease in the austenite to martensite transformation rate at the higher strain rates is attributed to adiabatic heating stabilizing the austenite. The specimen temperature as a function of strain up to UTS for various strain rates, plotted in Figure 4.12c, shows that the TRIP7Mn specimens reach higher temperatures than the QP3Mn specimens (Figure 4.6b) that did not exhibit austenite stabilization at high strain rates. Since much higher
strains are experienced in the TRIP7Mn than the QP3Mn, the austenite is exposed to higher temperatures from adiabatic heating. In the TRIP7Mn, the austenite content at 8% plastic strain is 22 vol.% for the test at 0.0005 s$^{-1}$, and the austenite content at 9% plastic strain is 21 vol.% for the test at 0.1 s$^{-1}$. The austenite fraction data from 10 and 100 s$^{-1}$ appear to converge with the austenite fraction data from 0.0005 s$^{-1}$ at 8% strain. At 12% strain where the quasi-static and high strain rate transformation rates begin to diverge, the specimen temperature at 0.1 s$^{-1}$ is close to 55 °C compared to 27 °C at 0.0005 s$^{-1}$ (according to the thermocouple data in Figure 4.12c). By 20% strain, the specimen temperature at 0.1 s$^{-1}$ is around 70 °C and the temperature at 0.0005 s$^{-1}$ is still close to 27 °C (according to the thermocouple data in Figure 4.12c). The effect of specimen temperature on the austenite to martensite transformation kinetics is discussed in more detail in Section 5.2. The similar amounts of specimen self-heating at 0.1 and 1 s$^{-1}$ in Figure 4.12c suggest that specimen self-heating at 0.1 s$^{-1}$ and greater strain rates is likely to be the same order of magnitude, which may be responsible for the similar austenite to martensite transformation rates at 0.1, 10, and 100 s$^{-1}$. The amount of self-heating at UTS is predicted from the area under the stress-strain curve for strain rates up to 100 s$^{-1}$ in Section 5.1.1. The similar extent of adiabatic heating and austenite to martensite transformation rates at 0.1, 10, and 100 s$^{-1}$ suggest that the mechanical driving force for the strain-induced austenite to martensite transformation may not change significantly due to strain rate.

A possible consideration in comparing adiabatic heating in QP3Mn and TRIP7Mn are the differences in thermal characteristics between austenite and ferrite/martensite. As dis-
cussed in Section 2.5.4, austenite has a lower thermal conductivity than ferrite (austenitic 304 stainless steel has a thermal conductivity of 16.2 W/(m·°C) compared to a thermal conductivity of 89 W/(m·°C) for ferritic HSLA steel) [149, 150]. Since the TRIP7Mn steel contains more austenite than the QP3Mn steel, the thermal conductivity is likely lower for the TRIP7Mn. The effective thermal conductivity is dominated by the majority of ferrite and/or martensite in the microstructures of both steels meaning that the effective thermal conductivities are expected to be similar between the two steels. A more in depth discussion of the differences in thermal behavior between austenitic and ferritic steels is provided in Section 2.5.4.

Representative true stress-strain curves for TRIP7Mn tested at room temperature (22 – 26 °C) in air at various strain rates are shown in Figure 4.14a. The curves are offset by 0.15 strain for clarity. The average true YS, UTS, and UE for all of the strain rates tested are summarized in Appendix C. Figure 4.14b shows work hardening rate \((dσ/dε)\) versus true strain for tests at 0.0005, 0.1, 10, and 100 s\(^{-1}\). A maximum appears in the work hardening rate curve at each strain rate due to the sigmoidal shape of the stress-strain curve. The maximum in the work hardening rate curve decreases with increasing strain rate meaning that work hardening during plastic deformation decreases with increasing strain rate. The Hollomon strain hardening exponent \(n\) was calculated from 0.1 true strain up to necking and is shown as a function of strain rate in Figure 4.14c. Examples of the curve fitting to calculate \(n\) are provided in Figure 3.13b in Chapter 3. The Hollomon power law strain hardening model is not a direct mathematical fit to a flow curve with a sigmoidal shape, but the values of \(n\) were calculated as a way to examine the general trend in how the amount of work hardening changes with strain rate. The value of \(n\) decreases almost continuously with increasing strain rate. The decrease in the average value of \(n\) from 0.5 at 0.0005 s\(^{-1}\) to less than 0.4 at 1 s\(^{-1}\) represents a tremendous decrease in work hardening with increasing strain rate. For comparison, the average \(n\) was between 0.09 and 0.14 for QP3Mn over the same strain rate range. As discussed previously, the large reduction in the amount of work hardening with increasing strain rate up to 1 s\(^{-1}\) is likely due to a combination of factors. One factor in the work hardening rate decrease is the stabilization of austenite due to adiabatic heating at high strain rates. Since the heating rate of the tensile specimen increases over the 0.0005 – 0.1 s\(^{-1}\) strain rate range, the austenite becomes more stable with increasing strain rate up to 0.1 s\(^{-1}\). An increase in austenite stability causes a decrease in work hardening rate. The diminishing effect of DSA (as demonstrated by the diminishing serrations in the flow curve combined with a decrease in UTS in Figure 4.12b) with increasing strain rate also contributes to lower flow stress with increasing strain rate. As strain rate increases, the lower YS increases while the UTS decreases, which also contributes to lower calculated \(n\)
Figure 4.14  (a) Example true stress-strain curves for TRIP7Mn tested at room temperature (22 − 26 °C) in air at various strain rates. The curves are offset by 0.15 strain. (b) Work hardening rate \(\frac{d\sigma}{d\varepsilon}\) versus total true strain for four strain rates. (c) The Hollomon strain hardening exponent, \(n\), as a function of engineering strain rate for TRIP7Mn.

values. Finally, the large amount of specimen self-heating with increasing strain rate shown in Figure 4.12c (specimen temperature reaches over 95 °C at UTS during the 0.1 s\(^{-1}\) test and over 105 °C at UTS during the test at 1 s\(^{-1}\)) makes a small thermal contribution to assist dislocations in overcoming short-range barriers, which causes a softening effect on flow stress. In the strain rate range of 1 − 100 s\(^{-1}\), the average value of \(n\) varies between 0.37 and
0.41 with an average value of 0.39, and there is no distinct trend due to scatter in the data. The average value of \( n \) decreases to less than 0.3 at 200 \( s^{-1} \).

Figure 4.15a shows the upper and lower YS and UTS versus strain rate. The uncertainty of the flow stress measurement is estimated to be 7 MPa. The upper and lower YS show the same trends with almost no strain rate sensitivity at low strain rates and small, positive strain rate sensitivity that gradually increases at high strain rates. In contrast, the UTS exhibits negative strain rate sensitivity up to 1 \( s^{-1} \) and is close to zero between 1 and 10 \( s^{-1} \). Negative UTS strain rate sensitivity was also observed for a 7.4Mn-1.55Al-0.14C-0.2Si (wt.%) steel (35 vol.% austenite) and a 10.1Mn-1.68Al-0.14C-0.2Si (wt.%) steel (57 vol.% austenite) that were tested at room temperature in the strain rate range 0.002 – 0.2 \( s^{-1} \) [165]. The amount of transformed austenite at 20% was unaffected by strain rate for the 7.4Mn-1.55Al-0.14C-0.2Si (wt.%) steel but decreased with increasing strain rate for the 10.1Mn-1.68Al-0.14C-0.2Si (wt.%) steel. Negative strain rate sensitivity associated with DSA was observed for Hadfield steel at room temperature between 0.0001 and 0.001 \( s^{-1} \) [83]. A softening effect from adiabatic heating and a reduction in the austenite transformation rate may also contribute to the negative UTS strain rate sensitivity by reducing the work hardening rate. The contributing factors to the decrease in flow stress with increasing strain rate are discussed further in the context of strain rate change tests in Section 5.1.2. Above 10 \( s^{-1} \), there is a small positive UTS strain rate sensitivity. Above 1 \( s^{-1} \), there is no serrated plastic flow (strain rates above 1 \( s^{-1} \) are outside of the DSA regime), the adiabatic heating is nearly constant, and the austenite to martensite transformation rate does not change as shown in Figure 4.13. These factors allow for the increase in the thermal component of flow stress with increasing strain rate to cause a noticeable increase in the overall flow stress.

The average engineering UE and TE are plotted versus strain rate in Figure 4.15b. The UE and TE for each test were both measured from the stress-strain curve (at strain rates above 0.1 \( s^{-1} \) a portion of the elongation data must be extrapolated from the strain gage data for each test). The uncertainty of the elongation measurement is estimated to be 0.1%. Figure 4.15b shows that the UE and TE generally decrease with increasing strain rate from 0.0005 to 1 \( s^{-1} \). The decrease in UE is expected from the decrease in work hardening rate with increasing strain rate and its associated effects on instability. The increase in UE and TE from 10 to 100 \( s^{-1} \) in Figure 4.15b has similarly been observed for first generation TRIP [14, 147]. The increase in elongation may be due the suppression of necking by strain rate hardening. The origin for the decrease in UE and increase in TE between 100 and 200 \( s^{-1} \) is unclear.
The retained austenite measurements and the flow curves at high strain rates may support the DSA mechanism of C-Mn complexes pinning partial dislocations in austenite that is proposed in literature [76, 83]. Comparing the 0.1 and 10 s\(^{-1}\) data, the austenite to martensite transformation rate is nearly the same, but discontinuous plastic flow was observed at 0.1 s\(^{-1}\) and not at 1 or 10 s\(^{-1}\) (based on thermal camera data at 1 s\(^{-1}\) in Appendix B and the appearance of the flow curves in Figures 4.12a and 4.14a), which suggests that the serrations are not likely due to the transformation itself. Lee et al. calculated that the C-Mn reorientation time is 0.5 s based on an activation energy of 0.75 eV [76]. The total duration of the tensile test is approximately 4.4 s at 0.1 s\(^{-1}\), 0.44 s at 1 s\(^{-1}\), and 0.044 at 10 s\(^{-1}\). Therefore, neglecting temperature effects and considering only test time, it is feasible that the C-Mn complexes have enough time to reorient during the 0.1 s\(^{-1}\) test and not during the 1 or 10 s\(^{-1}\) tests.

In summary, the upper and lower YS strain rate sensitivities are very low at 1 s\(^{-1}\) and lower strain rates, and above 1 s\(^{-1}\) the YS strain rate sensitivity is positive for the TRIP7Mn. The TRIP7Mn has significant negative UTS strain rate sensitivity up to 1 s\(^{-1}\) and very slight positive strain rate sensitivity above 10 s\(^{-1}\). Elongation follows a similar trend as UTS with respect to strain rate. These trends are due to a combination of factors: disappearance of DSA with increasing strain rate, a softening effect from adiabatic heating, and stabilization of austenite due to adiabatic heating. The TRIP7Mn has a lower austenite to martensite
transformation rate above 10% strain at high strain rates (0.1 – 100 s\(^{-1}\)) due to adiabatic heating stabilizing the austenite.

4.2.2 TRIP7Mn Tensile Properties and Retained Austenite Measurements at Various Temperatures

Example engineering stress-strain curves for TRIP7Mn at various temperatures and a quasi-static strain rate are shown in Figure 4.16. Overall, Figure 4.16 shows that with increasing test temperature the lower YS decreases, work hardening rate decreases, UTS decreases, and UE and TE both increase (over most of the temperature range). With increasing temperature, the flow stress from approximately 10% strain (beyond the YPE region) to UTS decreases. Elongation increases with increasing temperature up to 100 °C before decreasing at 115 °C. The increase in elongation with increasing temperature below 100 °C differs from the trend of decreasing elongation with increasing temperature for the QP3Mn. The fine serrations in the flow curve at room temperature increase in intensity at 40 and 55 °C, and the flow curve is characterized by distinct, step-like segments, which likely correspond to the travel of PLC bands from one end of the gage length to the other end [74, 75]. The fine serrations decrease in intensity and the distinct segments of the flow curve extend over greater amounts of strain as temperature increases through 70 and 85 °C. By 100 °C, the fine serrations disappear but the curve still contains distinct, nearly linear segments that are bound by abrupt load increases and drops. At 115 °C, there are no serrations or excursions in the flow curve, and there is very little work hardening compared to the lower test temperatures. The flow curve also does not exhibit serrations at -10 °C, and the work hardening rate is the highest at -10 °C. The pattern of the serrations increasing and then decreasing in magnitude and intensity with increasing temperature was similarly observed for Hadfield steel in the temperature range of -10 to 200 °C [83]. In Figure 4.16, the critical strain at which the serrations begin increases with increasing temperature. The serrations begin immediately following the YPE at 22 °C, around 10% strain for 40 °C and 55 °C, close to 14% strain for 70 °C, close to 18% strain for 85 °C, and at 30% strain for 100 °C. Zhang and Speer examined the effect of test temperature on steel with the same chemistry as the TRIP7Mn steel, but the steel was intercritically annealed for different amounts of time to achieve different ferrite and austenite grain sizes [166]. The microstructures and quasi-static, room temperature tensile tests from the 12, 24, 48, and 96 hour hold times share similarities with the TRIP7Mn steel in this work: the austenite and ferrite grains were equiaxed and similar in size to each other, the as-heat treated microstructure consisted of 35 – 40 vol.% retained austenite, and the stress-strain curves had distinct upper and lower
yield points (lower YS equal to or greater than 600 MPa), YPE greater than 5% strain, TE over 30%, and UTS equal to or greater than 1000 MPa. Tensile tests were performed in the temperature range of -80 to 120 °C at a 0.001 strain rate, and the temperature effects on the stress-strain curves for all four heat treatments were the same as the temperature effects described for the TRIP7Mn.

The austenite volume fraction versus plastic strain data for all test temperatures are shown in Figure 4.17. The TRIP7Mn retained austenite stability is very temperature sensitive in the examined temperature range. As the austenite to martensite transformation rate decreases with increasing temperature, the work hardening rate and UTS are reduced. Since the initial microstructure consists of ferrite and 40 vol.% austenite, the austenite to martensite transformation leads to a significant, observable high work hardening rate at room temperature that decreases substantially as austenite is stabilized by increasing temperature. Further, the transformation introduces martensite into the microstructure and causes dislocations to form in the ferrite, which both have a strengthening effect. When less martensite is formed during deformation, the composite strengthening is reduced.

Figure 4.18a shows the upper and lower YS and UTS versus temperature. Both the upper and lower YS decrease slightly with increasing temperature. The lower YS only decreases by an average of 64 MPa over the entire temperature range, while the upper YS decreases

![Figure 4.16](image_url)
by an average of 106 MPa. A large decrease in the upper YS occurs between 22 and 40 °C. At the elevated temperatures, the upper and lower YS are very close to each other with the difference between them ranging between 16 to 39 MPa. At room temperature, the upper and lower YS differ by 62 to 74 MPa. A contributing factor to the slight decline in YS with increasing temperature is the increased thermal contribution that lowers the stress necessary for dislocations to overcome short-range barriers. This YS dependence on temperature is in contrast to behavior observed for QP3Mn (Figure 4.8a shows a lack of YS temperature sensitivity for QP3Mn), likely due to the large amount of ferrite in the TRIP7Mn microstructure compared to the mostly martensitic QP3Mn microstructure. There are fewer long range dislocation interactions in ferrite than martensite, which means that temperature sensitive short range barriers have greater impact on deformation and strengthening mechanisms. Figure 4.17 suggests that less austenite transforms to martensite in the YPE region with increasing temperature, which means that potentially less austenite undergoes a stress-assisted transformation in the YPE regime with increasing strain rate. However, the effect of increased austenite stability on the yielding behavior is unclear. There is a more substantial effect of temperature on UTS than YS in TRIP7Mn. The UTS declines with increasing temperature and has an average 274 MPa decrease over the entire temperature range. Multiple factors contribute to the rapid decrease. There is a small contribution from the addition of thermal energy that assists dislocations in overcoming short range barriers.
The major contributors to the UTS behavior are the change in serrated flow behavior and the stabilization of austenite with increasing temperature.

Figure 4.18 Tensile mechanical properties of TRIPMn versus temperature: (a) UTS, upper YS, and lower YS and (b) TE and UE.

Figure 4.18b shows UE and TE versus temperature. The UE and TE show similar trends of increasing with increasing temperature up to 100 °C. For both UE and TE, the average increase between 22 and 100 °C is 17% strain. However, between 100 and 115 °C, the UE and TE both decrease by over 15% strain. As more austenite is stabilized in the microstructure with increasing temperature, less martensite forms during deformation. As the formation of martensite is delayed to higher strains with increasing temperature, the change in work hardening rate is more gradual with increasing strain, which contributes to an increase in ductility. The flow curve and austenite transformation behavior at 115 °C are very similar to the room temperature tensile results of a 7 wt.% Mn TRIP steel annealed at a low intercritical annealing temperature (575 °C), shown in Figure 2.6 [48]. Both tensile curves are also very similar in appearance and magnitude to the room temperature tensile curve in Figure 2.7 for a 21 wt.% Ni containing steel with a 0.7 μm grain size and 50-50 mixture of ferrite and stable austenite (did not exhibit transformation to martensite during deformation) [53]. In all three cases, a lack of serrations in the flow curve and low work hardening are associated with very stable austenite. The disappearance of the serrations in the flow curve coincides with the substantial decrease in elongation between 100 and 115 °C, but the disappearance of the serrations do not have a direct correlation with austenite stabilization. In Figure 4.17, the change in the austenite transformation rate between 85 and 100 °C is of a similar
magnitude to the change in transformation rate between 100 and 115 °C. Yet, elongation increases by approximately 2% strain between 85 and 100 °C and decreases by over 15% strain between 100 and 115 °C. Over most of the temperature range, TRIP7Mn elongation increases with increasing temperature, while the austenite consistently becomes more stable against transformation to martensite. The elongation values are especially high at 70, 85, and 100 °C (UE ranges 47% – 53% strain and TE ranges 55% – 59% strain) with a large increase in elongation between 55 and 70 °C (the average UE increases by 6% and the average TE increases by 9%). There is also a large reduction in strain hardening between 55 and 70 °C. The trends in elongation and work hardening are likely due to the combined effects of temperature on austenite stability and DSA. Samples from tensile specimens interrupted at 20% plastic strain were selected for further analysis with EBSD for the following test conditions: 22 °C at 10 s⁻¹, 70 °C at 0.0005 s⁻¹, and 115 °C at 0.0005 s⁻¹. The EBSD scans, which are included in Appendix E, show a decrease in the amount austenite from the as-received microstructure for all three conditions, and the deformed microstructures look similar for all three conditions despite differences in test temperature and strain rate.

In summary, the TRIP7Mn upper and lower YS decrease slightly with increasing temperature for tensile tests at a quasi-static strain rate in the range of 22 to 115 °C. The upper and lower YS do not change significantly between -10 and 22 °C. From -10 to 115 °C, the TRIP7Mn UTS decreases substantially, which is related in large part to the steady decrease in the austenite to martensite transformation rate. The serrations in the flow curve increase in magnitude and frequency up to 55 °C, then the intensity decreases and the appearance of the serrations changes with increasing temperature up to 100 °C, and finally the serrations do not appear at 115 °C. The UE and TE both increase with increasing temperature up to 100 °C. At 115 °C, the lack of DSA and highly stable austenite contribute to a sharp drop in elongation and very little work hardening.

4.2.3 TRIP7Mn Tensile Properties and Retained Austenite Measurements at Various Strain Rates and Temperatures

Figure 4.19 shows representative engineering stress-strain curves from tensile tests at -10, 22, and 70 °C and 0.0005 and 0.1 s⁻¹ for TRIP7Mn. The -10 °C curves are smooth and continuous without serrations in the stress-strain curve, and increasing strain rate results in small decreases in flow stress and elongation. The 22 °C curve at 0.0005 s⁻¹ exhibits very fine serrations and long, almost linear segments at high strains. At 0.1 s⁻¹ and 22 °C, the serrations are nearly imperceptible in the flow curve (the slight load increase near 27% strain followed by a long, nearly linear segment that extends to 34% strain is an example of the
presence of discontinuous plastic flow at 0.1 s\(^{-1}\)). The increase in strain rate from 0.0005 to 0.1 s\(^{-1}\) leads to a large decrease in UTS and ductility. The 70 °C curve at 0.0005 s\(^{-1}\) exhibits serrations of greater magnitude and frequency than the 22 °C curve at 0.0005 s\(^{-1}\), which may suggest a greater effect of DSA on the flow behavior at 70 °C than 22 °C as discussed in Section 4.2.2. At 0.1 s\(^{-1}\) at 70 °C, the magnitude and frequency of the serrations is severely diminished, but there are significant sudden load drops and increases that are indicative of serrated plastic flow. The 70 °C curves show the greatest reduction in ductility and UTS with increasing strain rate compared to the -10 and 22 °C tests.

![Engineering stress-strain curves for TRIP7Mn from tensile tests at -10, 22, and 70 °C and 0.0005 and 0.1 s\(^{-1}\).](image)

Figure 4.19 Engineering stress-strain curves for TRIP7Mn from tensile tests at -10, 22, and 70 °C and 0.0005 and 0.1 s\(^{-1}\).

The average engineering upper YS, lower YS, and UTS at 0.0005, 0.1, and 10 s\(^{-1}\) from tensile tests performed at -10, 22, and 70 °C are shown in Figure 4.20. Figures 4.20a and 4.20b show that for all three test temperatures, the strain rate sensitivities of the lower and upper YS are very similar. There is little or no YS strain rate sensitivity between 0.005 and 0.1 s\(^{-1}\) (low strain rate regime), and there is a positive strain rate sensitivity between 0.1 and 10 s\(^{-1}\) (high strain rate regime). In contrast, Figure 4.20c shows that the UTS strain rate sensitivities are highly dependent on test temperature. The -10 °C condition has a very slight negative UTS strain rate sensitivity in the low strain rate regime and there is no statistically significant difference in the average UTS in the high strain rate regime between 0.1 and 10 s\(^{-1}\). The 22 °C condition has a greater negative UTS strain rate sensitivity in the low strain rate regime than the -10 °C condition and possibly a slight negative UTS strain rate sensitivity in the high strain rate regime (the average UTS decreases between 0.1 and
10 s\(^{-1}\) but the error bars overlap). The 70 \(^\circ\)C condition exhibits a very high negative strain rate sensitivity in the low strain rate regime (the average UTS decreases from 923 MPa at 0.0005 s\(^{-1}\) to 805 MPa at 0.1 s\(^{-1}\)) and a positive strain rate sensitivity in the high strain rate regime (the average UTS increases from 805 MPa at 0.1 s\(^{-1}\) to 850 MPa at 10 s\(^{-1}\)). In a study of 7.4Mn-1.55Al-0.14C-0.2Si (wt.%)(35 vol.% austenite) and a 10.1Mn-1.68Al-0.14C-0.2Si (wt.%)(57 vol.% austenite) that were tested in the temperature range of -60 – 100 \(^\circ\)C and in the strain rate range 0.002 – 0.2 s\(^{-1}\), the UTS decreased with increasing temperature at all strain rates and negative UTS strain rate sensitivity was observed for most test temperatures (100 \(^{\circ}\)C resulted in the lowest strain rate sensitivity) [165].

The retained austenite contents as a function of strain at 0.0005 and 10 s\(^{-1}\) for the three test temperatures are shown in Figure 4.20d. At all test temperatures and strain rates, the austenite fraction decreases with increasing strain. At all three test temperatures, the austenite is more stable at 10 s\(^{-1}\) than at 0.0005 s\(^{-1}\) due to adiabatic heating. The austenite stabilization at 10 s\(^{-1}\) is greatest at 70 \(^{\circ}\)C. For example, at 20% strain the approximate difference in austenite content between the low and high strain rates is 7 vol.% for 70 \(^{\circ}\)C and 3 vol.% for -10 and 22 \(^{\circ}\)C. Stabilization of austenite due to adiabatic heating at high strain rates contributes to negative UTS strain rate sensitivity at all three test temperatures. A slight softening effect due to specimen self-heating likely makes a small contribution to decreasing the flow stress at high strain rates for all three test temperatures. The 22 \(^{\circ}\)C and 70 \(^{\circ}\)C conditions exhibit significant serrated plastic flow at 0.0005 s\(^{-1}\), and the serrations diminish substantially at 0.1 s\(^{-1}\) for both test temperatures. The decrease in the DSA effect with increasing strain rate may contribute to the higher negative UTS strain rate sensitivity at 22 and 70 \(^{\circ}\)C compared to -10 \(^{\circ}\)C in the low strain rate regime. The greater negative UTS strain rate sensitivity at 70 \(^{\circ}\)C than 22 \(^{\circ}\)C between 0.0005 and 0.1 s\(^{-1}\) is interpreted to be due to the slightly greater austenite stabilization at high strain rates at 70 \(^{\circ}\)C and a stronger DSA effect at 70 \(^{\circ}\)C at 0.0005 s\(^{-1}\) (meaning that increasing strain rate will cause a larger decrease in flow stress). The positive UTS strain rate sensitivity in the high strain rate regime at 70 \(^{\circ}\)C may be related to the differences in microstructure between 70 \(^{\circ}\)C and the lower test temperatures at 10 s\(^{-1}\). For 70 \(^{\circ}\)C at 10 s\(^{-1}\), approximately one third of the austenite transforms to martensite near UTS, but for 22 \(^{\circ}\)C at 10 s\(^{-1}\), more than two thirds of the austenite transforms to martensite near UTS. The 25 vol.% stable austenite in the microstructure at 10 s\(^{-1}\) near UTS for the 70 \(^{\circ}\)C condition may be responsible for the positive strain rate sensitivity.

Figure 4.21a shows the average engineering UE and Figure 4.21b shows the average engineering TE at 0.0005, 0.1, and 10 s\(^{-1}\) from tensile tests performed at -10, 22, and 70 \(^{\circ}\)C.
Figure 4.20  Average (a) upper YS, (b) lower YS, and (c) UTS of TRIP7Mn versus strain rate for three different test temperatures. The error bars represent one standard deviation for three tests. (d) Retained austenite as a function of plastic engineering strain at a low and a high strain rate for three different test temperatures. The retained austenite measurements were performed on specimens pulled beyond the YPE regime.

The trends in UE and TE with respect to strain rate are also dependent on test temperature. There is almost no change in elongation with increasing strain rate at -10 °C. For the -10 °C data, the average UE decreases by a total of 2% strain and the average TE decreases by a total of 1% strain between 0.0005 and 10 s⁻¹, but the error bars overlap for all three strain rates for both UE and TE. At 22 °C, elongation decreases with increasing strain rate in the
low strain rate regime, and there is no statistically significant change in elongation in the high strain rate regime. At 70 °C, there is a decline of over 20% strain between 0.0005 and 0.1 s⁻¹, and there is no further change in elongation in the high strain rate regime. The trends in elongation between 0.0005 and 0.1 s⁻¹ for the three test temperatures are very similar to the trends in UTS, e.g. the UE, TE, and UTS have little sensitivity to strain rate at -10 °C but there is a substantial decrease in UE, TE, and UTS with increasing strain rate at 70 °C. The stabilization of austenite due to adiabatic heating at 0.1 s⁻¹ may contribute to the decrease in UE, but austenite as a function of strain was not measured at 0.1 s⁻¹ for the 70 °C condition. The major decrease in elongation for 70 °C compared to 22 °C in the low strain rate regime may be due to a stronger DSA effect at 70 °C than at 22 °C. As discussed in Section 2.3, Dastur and Leslie found that the DSA effect due to C-Mn couples in austenite in Hadfield steel leads to an increase in elongation with increasing temperature in the DSA regime [83]. Based on the interpretation that there is a stronger DSA effect at 70 °C than at 22 °C, the increase in flow stress and increase in elongation due to DSA are greater at 0.0005 s⁻¹ for 70 °C than 22 °C. Increasing strain rate diminishes the DSA effect for both 22 and 70 °C, but when the DSA effect diminishes at 70 °C, it causes a greater decrease in elongation. For all three test temperatures, the UE and TE do not change significantly between 0.1 and 10 s⁻¹ likely due to the austenite transformation rate and specimen heating being very similar at 0.1 and 10 s⁻¹.

![Figure 4.21](image.png)

Figure 4.21 Average (a) UE and (b) TE of TRIP7Mn versus strain rate for three different test temperatures. The error bars represent one standard deviation for three tests.
In summary, for the TRIP7Mn, the upper and lower YS strain rate sensitivities are similar for tests at -10, 22, and 70 °C. However, test temperature leads to different trends in UTS and elongation with respect to strain rate in the range of 0.0005 – 0.1 s\(^{-1}\). In the low strain rate regime, stabilization of austenite due to adiabatic heating at high strain rates contributes to negative UTS strain rate sensitivity at all test temperatures. Austenite stabilization due to adiabatic heating at high strain rate is slightly greater at 70 °C compared to -10 or 22 °C. Serrations in the stress-strain curve diminish significantly between 0.0005 and 0.1 s\(^{-1}\) at both 22 and 70 °C, which suggests that increasing strain rate reduces the DSA effect. A reduction in the DSA effect also contributes to lower UTS and elongation at 0.1 s\(^{-1}\). The DSA is likely stronger at 70 °C, which leads to significantly greater decreases in UTS and elongation with increasing strain rate compared to 22 °C. Between 0.1 and 10 s\(^{-1}\), the UTS strain rate sensitivity is lower at -10 and 22 °C compared to the strain rate sensitivity between 0.0005 and 0.1 s\(^{-1}\). The UTS strain rate sensitivity is positive between 0.1 and 10 s\(^{-1}\) at 70 °C. For all three test temperatures, the UE and TE does not change significantly between 0.1 and 10 s\(^{-1}\).

4.3 DP980 and QP980 Results

The DP980 and QP980 study was conducted to provide tensile mechanical property data at strain rates ranging 0.0001 – 100 s\(^{-1}\) for comparison to the high strain rate tensile deformation behavior of QP3Mn and TRIP7Mn. Additionally, the effect of deformation history on high strain rate tensile behavior was examined with three different pre-strain conditions, because the effect of prior deformation on high strain rate mechanical behavior is relevant to the performance of formed parts in an automotive crash. The as-received DP980 and QP980 microstructures are shown in Figure 4.22. The SEM micrograph in Figure 4.22a of DP980 consists of martensite that has lightly etched features and regions of smooth, featureless ferrite. The SEM micrograph in Figure 4.22b of QP980 shows smooth, featureless ferrite grains, regions containing martensite with finely etched features due to film austenite, and austenite also present as islands in the ferrite and at ferrite boundaries. In the LOM micrographs in Figures 4.22c and 4.22d, the ferrite appears white and the martensite or martensite/austenite regions appear black. Figures 4.22c and 4.22d show that the QP980 microstructure has distinct banding of the martensite and ferrite regions but the DP980 microstructure did not exhibit banding. The DP980 contains 37 ± 6 vol.% ferrite and 63 ± 6 vol.% martensite. The QP980 contains 36 ± 6 vol.% ferrite, 53 ± 6 vol.% martensite, and 11 ± 1 vol.% retained austenite measured by XRD.
4.3.1 DP980 and QP980 Tensile Properties at Various Strain Rates

Figure 4.23 shows representative engineering stress-strain curves at five strain rates for both as-received steels. The flow curves for the highest strain rates display load ringing. The DP980 exhibits continuous yielding with high work hardening rates particularly below 2% strain. The QP980 yields at a higher stress and the work hardening rate (referring the the slope of the engineering stress-strain curve, ds/de) rapidly decreases after 1% strain. Strain rate does not have a significant effect on work hardening behavior for the QP980.
Work hardening rates beyond 2% strain are slightly higher at strain rates above 10 s\(^{-1}\) for the DP980. Figure 4.24 shows mechanical property from all of the tests. The 0.2% offset YS and UTS versus strain rate are plotted in Figure 4.24a. The calculated energy absorbed up to 5% strain versus strain rate is plotted in Figure 4.24b. The UE and TE versus strain rate are plotted for DP980 in Figure 4.24c and for QP980 in Figure 4.24d.

Figure 4.23 Representative engineering stress-strain curves at five strain rates for (a) DP980 and (b) QP980. The tensile tests were performed in air at room temperature (23 °C).
The 0.2% offset YS and UTS are shown as a function of strain rate for the DP980 and QP980 in Figure 4.24a. The uncertainty of the flow stress measurement is estimated to be 7 MPa. The flow stress data exhibit two distinct regimes of behavior: one in the strain rate range of 0.0001 to approximately 1 s\(^{-1}\) and the second in the strain rate range of approximately 1 to 100 s\(^{-1}\). The change in both YS and UTS over the low strain rate range is less than 50 MPa. The low strain rate sensitivities are attributed to athermal and time insensitive mechanisms, i.e. long-range dislocation interactions, controlling the flow behavior [133, 138]. Both the YS and UTS increase by more than 100 MPa in the high strain rate range. In the high strain rate regime, the time for dislocations to overcome short-range barriers is limited, and athermal dislocation interactions make significantly larger contributions to flow behavior [133, 138]. The 0.2% YS strain rate sensitivities are very low in the low strain rate regime for both steels compared to the steels with primarily ferritic or austenitic microstructures; the low YS strain rate sensitivities are due to the large fraction of martensite in the DP980 and QP980 microstructures [9, 134, 138, 139]. The QP980 has a higher 0.2% YS strain rate sensitivity than the DP980 in the low and high strain rate regimes, but the UTS strain rate sensitivities in both regimes are very close. The QP980 initially contains 53 vol.% martensite compared to the 63 vol.% of martensite in the DP980. Based on the greater YS strain rate sensitivities of ferrite and austenite compared to martensite reported in literature, the slightly higher YS strain rate sensitivity of the QP980 occurs due to the higher content of relatively soft and strain rate sensitive ferrite and austenite in the microstructure [9, 134, 139–141, 143]. Figure 4.25a shows the fraction of retained austenite as a function of strain in the QP980 at various strain rates. At all strain rates, a majority of the austenite transformed to martensite at strains less than the uniform strain. Approximately 4 vol.% retained austenite remains untransformed in the QP980 at UTS meaning that the martensite content is 60 vol.% at UTS. The 60 vol.% martensite in the QP980 at UTS is close to the 63 vol.% martensite in the DP980 meaning that both steels have similar amounts of a hard constituent with low strain rate sensitivity (although the martensite is different between the DP980 and QP980 due to different chemistry and processing). The fresh martensite in the QP980 at UTS may contribute to the similar UTS strain rate sensitivities between the DP980 and QP980. The UTS for both steels has a negative strain rate sensitivity in the low strain rate regime. Adiabatic heating of the tensile specimen during deformation has a softening effect on flow stress. If the softening effect is greater than the increase in flow stress due to the increase in strain rate, then the flow stress will decrease with increasing strain rate. Figure 4.25b shows the temperature measured at UTS for QP980 at strain rates up to 0.1 s\(^{-1}\). The heating of the tensile specimen increases rapidly with increasing strain rate in the range of 0.0001 – 0.1 s\(^{-1}\), which may have an observable softening effect.
on the flow stress. At strain rates of 1 s\(^{-1}\) and higher, the increase in flow stress due to short-range dislocation barriers is greater than the softening effect of adiabatic heating. The opposing effects of adiabatic heating and increasing strain rate on flow stress will be explored further in Section 5.1.2. The specimen temperature increase plotted as a function of strain with the corresponding stress-strain curve for select low strain rate conditions is provided in Appendix B. Thermal camera images from tests at 0.1 and 10 s\(^{-1}\) are provided in Appendix B. The thermal camera images show the relatively uniform heating (to temperatures below approximately 60 °C) across the entire gage length during uniform deformation followed by localized heating during necking (more than 30 °C higher temperature in the necking region than the uniformly deformed regions) and a large temperature spike at the fracture surface (temperature exceeds 160 °C).

The UE and TE plotted versus strain rate are shown in Figure 4.24c for DP980 and Figure 4.24d for QP980. The uncertainty of the elongation measurement is estimated to be 0.1%. The elongation data has significant scatter, especially in TE, that makes it challenging to discern trends. The QP980 has higher UE and TE than the DP980 at all strain rates due to the retained austenite in the QP980 microstructure (although the martensite is different between the DP980 and QP980 and may affect elongation differently in each steel). For both steels, there is a slight decrease in UE (on the order of 2 to 3% strain) and no significant trend in TE at strain rates up to 10 s\(^{-1}\). The decrease in UE up to 0.1 s\(^{-1}\) coincides with the decrease in UTS. Between 10 and 100 s\(^{-1}\), the UE and TE increase by approximately 3 to 6% strain. During high strain rate tensile tests (above 1 s\(^{-1}\)), strain rate hardening causes a resistance to necking and leads to an increase in uniform elongation [14, 162].

Energy absorption is a property that can be used to predict the crash performance of a sheet steel component. The energy absorbed during deformation was calculated as the area under the engineering stress-strain curve up to 5% strain for the DP980 and QP980, because the lowest uniform elongation for the DP980 was 5.9%. The plot of energy absorbed versus strain rate in Figure 4.24b shows that the amount of energy absorbed is very similar for both materials despite the difference in work hardening behavior below 2% strain. The energy absorbed initially decreases with increasing strain rate, then begins to increase around 0.1 s\(^{-1}\) and increases significantly for strain rates of 10 s\(^{-1}\) and greater. This trend is similar to the trend in UTS with respect to strain rate. The increase in energy absorption with increasing strain rate above 0.1 s\(^{-1}\) is due to the higher positive flow stress strain rate sensitivity in the high strain rate regime (1 – 100 s\(^{-1}\)) compared to the low strain rate regime (0.0001 – 1 s\(^{-1}\)).

Figure 4.25a shows retained austenite for the QP980 steel as a function of plastic strain at 0.001, 0.1, 10, and 100 s\(^{-1}\). The uncertainty of the retained austenite measurement is
Figure 4.24  (a) Engineering UTS and 0.2% offset YS versus strain rate for DP980 and QP980. (b) Energy absorbed up to 5% strain versus strain rate. Engineering UE and TE versus strain rate for (c) DP980 and (d) QP980.

estimated to be 1 vol.% austenite based on repeated XRD measurements of austenite in as-received specimens. For all strain rates, the austenite fraction decreases with increasing strain. There is no substantial difference in the retained austenite transformation behavior due to strain rate. Figure 4.25b shows that adiabatic heating in excess of 25 °C was measured for QP980 at UTS at 0.1 s⁻¹. Adiabatic heating at UTS is expected to be the same or greater at higher strain rates (since the areas under the stress-strain curves are the same or greater for strain rates above 0.1 s⁻¹). Although adiabatic heating occurs at 0.1, 10, and 100 s⁻¹,
the austenite transformation rate is nearly the same at all four strain rates. Similarly, the austenite to martensite transformation is unaffected by strain rate \((0.0005 \text{ – } 100 \text{ s}^{-1})\) in the QP3Mn. The lack of an effect of strain rate on the austenite to martensite transformation supports the similar trends in UE and TE with respect to strain rate for DP980 and QP980. If the austenite transformation rate was significantly affected by strain rate, then the trend in elongation with respect to strain rate would be expected to differ between QP980 and DP980. An example of how the X-ray peak intensities change, particularly the decrease in the austenite peak intensities, with increasing strain for the \(0.001 \text{ s}^{-1}\) condition is provided in Appendix D.

![Figure 4.25](image)

Figure 4.25  (a) Retained austenite as a function of plastic engineering strain for QP980 at various strain rates. (b) Specimen temperature at UTS measured with a K-type thermocouple.

In summary, the DP980 and QP980 exhibit two distinct regimes of flow stress strain rate sensitivity with a transition around \(1 \text{ s}^{-1}\) between low strain rate sensitivity and higher, positive strain rate sensitivity, which is consistent with other AHSS. The YS strain rate sensitivity is close to zero for the DP980 and slightly positive for the QP980 up to \(1 \text{ s}^{-1}\). Above \(1 \text{ s}^{-1}\), the YS strain rate sensitivity is positive for both steels. A slight negative UTS strain rate sensitivity for DP980 and QP980 in the range of \(0.0001 \text{ – } 0.1 \text{ s}^{-1}\) is likely due to a softening effect from adiabatic heating. At \(1 \text{ s}^{-1}\) and greater strain rates, the UTS strain rate sensitivity is positive for DP980 and QP980. There is a lot of scatter in the UE and TE data for the DP980 and QP980, and strain rate has little effect on ductility. While the retained austenite does result in greater elongation for the QP980 than the DP980, it may only lead
to a slightly higher YS strain rate sensitivity in the QP980 while not affecting trends in UTS or ductility with respect to strain rate. Strain rate does not affect the austenite to martensite transformation in the QP980.

4.3.2 DP980 and QP980 Tensile Properties at Various Strain Rates for Three Pre-Strain Conditions

The majority of the high strain rate tensile testing in this work is from specimens in an as-processed condition without any deformation prior to tensile testing. However, stress in a tensile specimen is a function of strain, strain rate, temperature, and deformation history [64]. Deformation history is an important consideration for studying AHSS for automotive applications, because AHSS sheets are formed into automotive body members, which results in a certain level of pre-strain in the members [147]. For example, DP600 and TRIP600 tensile specimens pre-strained up to 10% strain prior to testing resulted in a small increase of the yield and tensile strength strain rate sensitivities with increasing amount of pre-strain [147]. The deformation history of a material can affect flow stress due to the specific deformation substructure that was formed. Plastic deformation is a path dependent process meaning that different deformation substructures can form depending on strain rate, temperature and stress state [64]. To gain a perspective on how prior deformation history may affect subsequent high strain rate mechanical behavior, three pre-strain conditions were tested at four strain rates for the DP980 and QP980 steels. Pre-straining was performed at 0.0008 s\(^{-1}\), and the pre-strained conditions were 5% strain at room temperature (5% PS), 5% strain at room temperature followed by tempering at 180 °C for 20 min (5% PS + 180 °C, 20 min), and 5% strain at 80 °C (5% PS at 80 °C). The 5% strain level was selected to be representative of severe plastic deformation during forming, and the heat treatment at 180 °C for 20 min was selected to represent a high temperature, long duration bake hardening heat treatment.

Example engineering stress-strain curves from the as-received and three pre-strain conditions for DP980 and QP980 are shown in Figure 4.26. The DP980 flow curves for the four different conditions are shown in Figure 4.26a for tests at 0.001 s\(^{-1}\) and in Figure 4.26b for tests at 10 s\(^{-1}\). The QP980 flow curves for the four different conditions are shown in Figure 4.26c for tests at 0.001 s\(^{-1}\) and in Figure 4.26d for tests at 10 s\(^{-1}\). The flow curves for the 5% PS condition correspond closely to the as-received curves for both steels at 0.001 and 10 s\(^{-1}\). The 5% PS + 180 °C, 20 min curves exhibit increased flow stress due to static strain aging and necking very soon after yielding for both steels at 0.001 and 10 s\(^{-1}\). The 5% PS at 80 °C curves have a similar static strain aging response as the 5% PS + 180 °C, 20 min condition for the DP980 at both strain rates, but the QP980 5% PS at 80 °C curves exhibit yield
point elongation for 5% strain followed by work hardening with no significant difference in flow stress from the as-received curves at both strain rates. Mechanical properties from all of the tests in the pre-strain study are shown in Figure 4.27. The 5.5% offset flow stresses are plotted versus strain rate for DP980 in Figure 4.27a and for QP980 in Figure 4.27b. The total elongations (pre-strain plus strain to fracture) are plotted versus strain rate for DP980 in Figure 4.27c and for QP980 in Figure 4.27d.

Figure 4.26 Engineering stress-strain curves for as-received and pre-strain conditions of DP980 at (a) 0.001 s\(^{-1}\) and (b) 10 s\(^{-1}\) and QP980 at (c) 0.001 s\(^{-1}\) and (d) 10 s\(^{-1}\).

Figures 4.27a and 4.27b show the 5.5% offset flow stresses versus strain rate, which were measured to compare the flow stress response of the three different pre-strain conditions relative to the as-received sample. The 5.5% strain refers to the total accumulated strain in each sample. Table 4.1 contains a summary of the difference in the average 5.5% offset flow stress between the as-received and pre-strain conditions for both steels and all three pre-strain
Table 4.1 – Difference in the average 5.5% offset flow stress between as-received and various pre-strain conditions from tests at four different strain rates for DP980 and QP980 (a positive value indicates a higher flow stress for the pre-strain condition). The temper was performed at 180 °C for 20 min.

<table>
<thead>
<tr>
<th>Condition</th>
<th>DP980 0.001 s⁻¹</th>
<th>DP980 1 s⁻¹</th>
<th>DP980 10 s⁻¹</th>
<th>DP980 100 s⁻¹</th>
<th>QP980 0.001 s⁻¹</th>
<th>QP980 1 s⁻¹</th>
<th>QP980 10 s⁻¹</th>
<th>QP980 100 s⁻¹</th>
</tr>
</thead>
<tbody>
<tr>
<td>5% PS</td>
<td>22</td>
<td>58</td>
<td>29</td>
<td>31</td>
<td>9</td>
<td>50</td>
<td>6</td>
<td>6</td>
</tr>
<tr>
<td>5% PS at 80 °C</td>
<td>59</td>
<td>83</td>
<td>50</td>
<td>34</td>
<td>0</td>
<td>22</td>
<td>-12</td>
<td>-53</td>
</tr>
<tr>
<td>5% PS + Temper</td>
<td>87</td>
<td>144</td>
<td>125</td>
<td>110</td>
<td>75</td>
<td>107</td>
<td>81</td>
<td>105</td>
</tr>
</tbody>
</table>

conditions at all four strain rates. Most of the values are positive, which means that the 5.5% offset flow stress is greater for the pre-strain condition than the as-received condition. For all strain rates and pre-strain conditions, the difference between the average 5.5% offset flow stress for the as-received samples and the pre-strain samples is greater for the DP980 than the QP980 indicating that the static strain aging response is stronger for all three pre-strain conditions for the DP980 than the QP980. Static strain aging occurs due to carbon and/or nitrogen pinning dislocations that were generated during prior plastic deformation of the steel [64]. The mobile dislocation density prior to aging and the availability of carbon and nitrogen in solution to pin dislocations are both factors that could be involved in the different aging responses of DP980 and QP980. A possible explanation for the lower susceptibility of QP980 to static strain aging is that the retained austenite in microstructure acts as a carbon sink for the carbon in solution. The 5% PS + 180 °C, 20 min condition exhibits the strongest aging effect for both steels. Across the entire strain rate range, the increase in the average 5.5% offset flow stress from the as-received condition to the 5% PS + 180 °C, 20 min condition ranges 87 to 144 MPa for the DP980 and 75 to 107 MPa for the QP980. In a study of first generation DP and TRIP steels, the DP steel exhibited a greater increase in flow stress after bake hardening (80 MPa versus 60 MPa for the TRIP) and a greater reduction in ductility (UE lower by 9% versus 6% for the TRIP) after 5% pre-straining and bake hardening at 175 °C for 30 min [167]. Atom probe tomography of the DP steel showed that during bake hardening carbon diffused to dislocations in the ferrite and carbon clusters and rod-like carbides formed in the martensite. For the TRIP steel, TEM and APT showed that the pre-straining and bake hardening led to the formation of dislocation cells in ferrite although the average carbon content of the ferrite was much lower than in the DP steel (0.02 at.% C in the TRIP versus 0.04 at.% C in the DP), and there was also segregation of carbon to micro-twin boundaries in the austenite. For both steels, the strain rate sensitivity between 0.001 and 1 s⁻¹ is positive for the pre-strain conditions and is near zero for the as-received condition. The positive strain rate sensitivity is higher for the DP980, which also exhibits a
stronger strain aging effect. However, static strain aging is usually associated with low strain rate sensitivity [64]. Between 10 and 100 s$^{-1}$, the strain rate sensitivity for all conditions is positive and higher than the strain rate sensitivity at the lower strain rates, and pre-strain does not have a clear effect on the magnitude of the strain rate sensitivity.

Figure 4.27  Engineering 5.5% offset flow stress from as-received and pre-strain conditions versus strain rate for (a) DP980 and (b) QP980. Total elongation versus strain rate for as-received and pre-strain conditions of (c) DP980 and (d) QP980.
Figures 4.27c and 4.27d show the TE versus strain rate for the three pre-strain conditions and the as-received condition. The TE refers to the total accumulated strain in the sample, e.g. the 5% pre-strain plus the additional strain until fracture upon reloading. Figure 4.27c shows that the TE for the 5% PS at 80 °C and 5% PS + 180 °C, 20 min DP980 conditions are consistently lower than the as-received and 5% PS conditions. Static strain aging decreases ductility in addition to increasing yield stress [64]. The two pre-strain conditions with lower TE also had higher 5.5% offset flow stresses. There is no significant difference in TE in Figure 4.27d between the as-received, 5% PS and 5% PS at 80 °C at any of the strain rates for the QP980. However, the 5% PS + 180 °C, 20 min specimens have average TE values that are 5 to 8% less than the average TE for the other three conditions. Again, the pre-strain condition with the lowest TE also has the highest 5.5% offset flow stress.

In summary, flow stress strain rate sensitivity is greater for the pre-strain conditions than the as-received condition in the range 0.001 – 1 s⁻¹, but pre-straining does not significantly affect strain rate sensitivity above 1 s⁻¹. The trend in elongation with respect to strain rate is similar between the as-received and pre-strain conditions. The DP980 is more susceptible to static strain aging for all of the pre-strain conditions. For both steels, the pre-straining followed by tempering at 180 °C for 20 min (to simulate bake hardening following forming) results in a strong static strain aging response with substantially reduced ductility. Static strain aging has a greater effect on ductility than strain rate.
CHAPTER 5
DIFFERENTIATING THE EFFECT OF ADIABATIC HEATING FROM
STRAIN RATE ON DEFORMATION BEHAVIOR

This chapter focuses on differentiating the effect of adiabatic heating from strain rate on flow stress and retained austenite stability in third generation AHSS. Adiabatic heating can have a significant softening effect on flow stress and a stabilizing effect on retained austenite under certain conditions for these new high strength steels. First, the behavior of YS and UTS as a function of strain rate is compared for QP980, DP980, QP3Mn, and TRIP7Mn. Then, the effect of adiabatic heating on flow stress and retained austenite stability is examined more in depth with QP3Mn and TRIP7Mn. A simple thermal model is used to predict the specimen temperature due to heating from plastic work under adiabatic heating conditions. These results are compared to experimental measurements at lower strain rates. The role of adiabatic heating in high strain rate tensile deformation behavior is examined more closely with strain rate change and interrupted tensile tests. Finally, the Olson-Cohen model is applied to the retained austenite measurements from different test temperatures to model the effect of temperature on the austenite to martensite transformation kinetics.

5.1 High Strain Rate Tensile Deformation Behavior of AHSS

To compare the effect of strain rate on flow stress for all four steels, the average true YS and UTS (true stress calculated from engineering UTS) data versus strain rate for DP980, QP980, QP3Mn, and TRIP7Mn are shown together in Figures 5.1a and 5.1b. Data from Figures 4.5a, 4.15a, and 4.24a were re-plotted in terms of true stress for Figure 5.1. Figure 5.1a shows that for all strain rates, the steels ranked in order from lowest to highest average yield strength are DP980, TRIP7Mn (lower YS), QP980, and QP3Mn. For a majority of the data, yield strength increases with increasing strain rate for all four steels (an example of an exception is the decrease in YS for QP3Mn between 10 and 30 s\(^{-1}\)). Between 0.0001 s\(^{-1}\) and 100 s\(^{-1}\), the YS increases by 99 MPa for DP980, by 142 MPa for QP980, by 91 MPa for TRIP7Mn, and by 189 MPa for QP3Mn. Figure 5.1b shows that the average UTS of DP980 and QP980 are the same or very close at all strain rates (maximum difference in average UTS is 37 MPa at 100 s\(^{-1}\)). The average UTS of the TRIP7Mn is higher than the average UTS values of DP980 and QP980 for most of the strain rates, but the difference between the UTS values decreases with increasing strain rate due to the negative strain rate sensitivity.
of the TRIP7Mn. At 30 and 100 s$^{-1}$ the TRIP7Mn, DP980, and QP980 average UTS values are the same or the error bars overlap. The QP3Mn has the highest UTS values at all of the strain rates. In contrast to the trend in average YS, the UTS is nearly constant for QP3Mn and decreases with increasing strain rate for DP980, QP980 and TRIP7Mn below 1 s$^{-1}$. Above 1 s$^{-1}$, the average UTS increases with increasing strain rate for DP980, QP980 and QP3Mn, and UTS is nearly constant for TRIP7Mn.

The calculation of the strain rate sensitivity parameter, $m$, across large strain rate ranges is complicated for QP3Mn and TRIP7Mn due to scatter in the data and the gradual, often non-linear change in flow stress with increasing strain rate. Dynamic-static strength ratios (flow stress at a high strain rate divided by flow stress at a static strain rate) can be used to compare the general strain rate sensitivities of different steels [134, 140, 142]. The concept of a dynamic-static strength ratio was used to calculate the high/low strain rate ratio of YS and UTS for two ranges of strain rates (instead of just comparing a high strain rate like 100 s$^{-1}$ to a quasi-static strain rate like 0.0001 s$^{-1}$). In the low strain rate regime, the average strength at 0.1 s$^{-1}$ was divided by the average strength at 0.0001 s$^{-1}$. In the high strain rate regime, the average strength at 100 s$^{-1}$ was divided by the average strength at 1 s$^{-1}$. The high/low strain rate ratios for YS and UTS plotted versus the quasi-static YS
and UTS are shown in Figure 5.2. A ratio greater than one indicates positive strain rate sensitivity, and a value less than one indicates negative strain rate sensitivity. The closer the ratio is to one the lower the strain rate sensitivity.

![Diagram showing high/low strain rate ratios for YS versus quasi-static true YS and UTS versus quasi-static true UTS for DP980, QP980, QP3Mn, and TRIP7Mn.](image)

Figure 5.2 High/low strain rate ratios for (a) YS versus quasi-static true YS and (b) UTS versus quasi-static true UTS (true stress calculated from engineering UTS) for DP980, QP980, QP3Mn, and TRIP7Mn. The open symbols are the average strength at 0.1 s$^{-1}$ divided by the average strength at 0.0001 s$^{-1}$, and the filled symbols are the average strength at 100 s$^{-1}$ divided by the average strength at 1 s$^{-1}$. The YS values are the lower yield point for the TRIP7Mn and the 0.2% offset YS for the other steels. Ratio greater than one indicates positive strain rate sensitivity, and a ratio below one indicates negative strain rate sensitivity.

Trends in YS strain rate sensitivity are similar for DP980, QP980, and TRIP7Mn, despite their different as-processed microstructures. This behavior is consistent with findings of Larour et al. in Figure 2.14b that YS strain rate sensitivity strongly correlates with the quasi-static YS for high strength steels [143]. The various strengthening mechanisms employed to achieve high strength steels usually increase the athermal component of flow stress, which becomes much greater in magnitude than the thermal component. The strong, dominant athermal component of flow stress leads to relatively low strain rate sensitivity compared to lower strength steels across the entire strain rate range regardless of microstructure. The relative amounts of different constituents in the microstructure may only lead to subtle differences in YS strain rate sensitivity as discussed for DP980 compared to QP980 in Section 4.3.1. Figure 5.2a shows that the YS strain rate sensitivity in the range 1 – 100 s$^{-1}$
is greater for DP980 and QP980 than for TRIP7Mn, which may be related to differences in the yielding behavior. The DP980 and QP980 exhibit continuous yielding; whereas, the TRIP7Mn has distinct upper and lower yield points with pronounced YPE. In Figure 5.1a, there is significant scatter in the QP3Mn YS data (e.g. one standard deviation of the 0.01 s\(^{-1}\) YS data is 93 MPa but is only 10 MPa for TRIP7Mn), which makes it difficult to identify clear trends. However, Figure 5.2a shows that between 0.0001 and 0.1 s\(^{-1}\) the QP3Mn YS strain rate sensitivity appears to be higher than the YS strain rate sensitivities of DP980, QP980 and TRIP7Mn (the high/low strain rate ratio is within ± 0.03 of 1 for DP980, QP980 and TRIP7Mn but is 1.12 for QP3Mn). This finding is unexpected, because the as-processed QP3Mn contains 86 vol.% martensite, which is higher than the amount of martensite in the as-processed DP980 and QP980. The high YS strain rate sensitivity of the QP3Mn also contradicts the findings of Larour \textit{et al.}, shown in Figure 2.14, that strain rate sensitivity is very low for AHSS containing a large quantity of a hard phase and having a high quasi-static YS \cite{143}. In Figure 5.2a, the DP980 and QP980 show greater YS strain rate sensitivity between 1 and 100 s\(^{-1}\) than TRIP7Mn and QP3Mn. Although the TRIP7Mn microstructure is 61 vol.% ferrite, the YS strain rate sensitivity at high strain rates is greater for the DP980 that contains 63 vol.% martensite and the QP980 that contains 53 vol.% martensite.

Figure 5.1b shows that the trends in UTS strain rate sensitivity differ significantly between TRIP7Mn and the other three steels. Figure 5.2b shows that the UTS strain rate sensitivities are negative for DP980, QP980 and TRIP7Mn in the low strain rate range with the strongest negative strain rate sensitivity for TRIP7Mn (ratio of 0.90). The UTS strain rate sensitivity for QP3Mn in the low strain rate regime is very low (ratio of 0.99). As discussed in Section 4.2.1, the large negative UTS strain rate sensitivity up to 1 s\(^{-1}\) for the TRIP7Mn is due to a combination of multiple factors, including austenite stabilization due to adiabatic heating and a decrease in DSA with increasing strain rate. The slight negative UTS strain rate sensitivity of DP980 and QP980 and the lack of UTS strain rate sensitivity for QP3Mn at strain rates up to 0.1 s\(^{-1}\) may be due to a softening effect from adiabatic heating, which is explored in Section 5.1.2. The lack of UTS strain rate sensitivity for QP3Mn at strain rates up to 0.1 s\(^{-1}\) may be related in part to adiabatic heating, although quasi-static test temperatures up to 85 °C did not affect the UTS. At 1 s\(^{-1}\) and above, Figure 5.2b shows that there is greater positive UTS strain rate sensitivity for DP980 and QP980 than QP3Mn and that the TRIP7Mn UTS has almost no strain rate sensitivity. The lower strain rate sensitivity for QP3Mn is consistent with the mostly martensitic microstructure of the QP3Mn, with the difference between the UTS and YS behavior being that a large amount of the austenite has transformed by the UTS, compared to the large quantities of strain rate sensitive ferrite in the DP980 and QP980 microstructures (37 and 36 vol.% ferrite
respectively). The lower strain rate sensitivity of the much higher strength QP3Mn is also consistent with the findings of Larour et al. for AHSS [143].

5.1.1 Thermal Modeling

The change in temperature at UTS as a function of strain rate is useful for understanding the contribution of adiabatic heating to a softening effect on flow stress and a stabilizing effect on the retained austenite. The change in temperature at UTS was measured experimentally up to 10 s\(^{-1}\) for the QP3Mn and TRIP7Mn. Above 10 s\(^{-1}\), the data acquisition rates of the thermocouple and infrared camera were too slow to record temperature during the tensile test. Therefore, an effort was made to use the temperature model given by Equation 2.7 to predict the change in temperature at UTS at 100 s\(^{-1}\).

The values of \(\beta\) equal to 0.95, \(\rho\) equal to 7800 kg/m\(^3\), and \(C_p\) equal to 450 J/kg·°C were chosen based on TRIP steel temperature modeling reported in literature [168, 169]. It was assumed that the material constants are similar for both steels and that the effects of austenite on the thermal conductivities and heat capacities are small due to the majority of the microstructure being martensite and/or ferrite. Differences in thermal conductivities and heat capacities between ferritic and austenitic microstructures are discussed in Section 2.5.4. The differences in chemistry may lead to slight differences in the values of \(\rho\) and \(C_p\), but small differences in these values have a negligible effect on the predicted \(\Delta T\). The fraction of plastic work transformed to heat during plastic deformation, \(\beta\), is approximated as 0.9 for most metals [133]. The specific value of \(\beta\) depends on the operative deformation mechanisms [133]. The value of \(\beta\) can also change as a function of strain (e.g. changes in dislocation density can affect \(\beta\)). For steel that deforms primarily by dislocation glide, \(\beta\) increases rapidly following yielding and becomes approximately constant above 0.04 plastic strain [169]. During deformation of steels exhibiting austenite to martensite transformation, heat is generated from plastic deformation due to dislocation interactions and the phase transformation. The latent heat of the phase transformation is assumed to be a small fraction of the total heat generated during plastic deformation. Zaera et al. suggested using a \(\beta\) higher than 0.9 to estimate the thermal contribution of the phase transformation under adiabatic heating conditions during high strain rate deformation [170]. For the \(\Delta T\) calculations in this work, a \(\beta\) of 0.95 is assumed to account for all of the heat generated during plastic deformation. Examples of how different values of \(\beta\), \(\rho\), and \(C_p\) affect the calculated \(\Delta T\) are provided in Appendix F.
The results from the model and the specimen temperature measurements at UTS for QP3Mn and TRIP7Mn are shown in Figure 5.3. At the lower strain rates, the model calculation is based on the area under the stress-strain curve for the same tensile test as the temperature measurement with the thermocouple spot welded to the center of the gage length. There is a large difference between the model prediction and the experimental temperature measurement at strain rates below 0.01 s\(^{-1}\) for the QP3Mn and below 0.1 s\(^{-1}\) for the TRIP7Mn, because there is time for thermal diffusion and heat dissipation during tensile tests at the low strain rates. At the higher strain rates, the modeling was performed for three tensile tests and the average change in temperature is reported. Figure 5.3a shows the change in specimen temperature at UTS from thermocouple measurements and calculated by Equation 2.7 for QP3Mn. The experimentally measured change in specimen temperature at UTS increases with increasing strain rate up to 33 °C at 0.01 s\(^{-1}\), and between 0.01 and 10 s\(^{-1}\) the change in specimen temperature is between 33 and 38 °C. The model prediction decreases slightly with increasing strain rate to a minimum near 27 °C at 1 s\(^{-1}\), and the model prediction increases with increasing strain rate (1 – 100 s\(^{-1}\)) to over 40 °C at 100 s\(^{-1}\). Figure 5.3b shows the change in specimen temperature at UTS from thermocouple measurements and calculated by Equation 2.7 for TRIP7Mn. The experimentally measured change in specimen temperature at UTS increases with increasing strain rate up to 74 °C at 0.1 s\(^{-1}\), and between 0.1 and 10 s\(^{-1}\) the change in specimen temperature is between 73 and 82 °C. The model prediction decreases from 82 to 69 °C between 0.0003 and 0.001 s\(^{-1}\), and above 0.001 s\(^{-1}\) the model prediction gradually increases from 69 to 74 °C with increasing strain rate. Equation 2.7 provides the change in temperature of the specimen under adiabatic conditions. At high strain rates, a near adiabatic condition can be reached when the test is too short for any significant thermal diffusion and heat loss. In Figure 5.3, the model prediction converges with the experimental temperature measurement at 0.01 s\(^{-1}\) for QP3Mn and at 0.1 s\(^{-1}\) for the TRIP7Mn. However, for the QP3Mn, the experimental temperature measurements suggests that a complete adiabatic heating condition is reached around 0.1 s\(^{-1}\) rather than at 0.01 s\(^{-1}\). The difference in the experimental temperature measurements between 0.1 and 1 s\(^{-1}\) is 2 °C compared to a 5 °C difference between 0.01 and 0.1 s\(^{-1}\).

For the QP3Mn, the model slightly underpredicts the change in specimen temperature at UTS for 0.1, 1, and 10 s\(^{-1}\). At 100 s\(^{-1}\), the model predicts that the average temperature change at UTS is 8 °C higher than the temperature change at 10 s\(^{-1}\). Since the QP3Mn UTS is fairly insensitive to temperature in the range of 22 to 85 °C, the slightly higher predicted heating at 100 s\(^{-1}\) is not expected to have a significantly greater softening effect on flow stress compared to the heating at 10 s\(^{-1}\). There is slight positive UTS strain rate sensitivity between 10 and 100 s\(^{-1}\) in Figure 4.5a despite the higher predicted heating at
Figure 5.3  Experiment measurements and model predictions for the adiabatic change in temperature at UTS for (a) QP3Mn and (b) TRIP7Mn. At low strain rates, the model prediction (black circle) and the experimental temperature measurement (filled grey circle) are from the same tensile test. At higher strain rates, the average model prediction from three different tensile tests (triangle) is plotted with error bars representing one standard deviation. Temperature measurements were made with a K-type thermocouple.

100 s\(^{-1}\). Additionally, the austenite to martensite transformation rate is similar between 10 and 100 s\(^{-1}\) for the QP3Mn as shown in Figure 4.6a, which means that the slightly higher predicted heating at 100 s\(^{-1}\) also does not significantly affect the austenite stability.

The model agrees well with the TRIP7Mn specimen temperature measurements at 0.1 and 10 s\(^{-1}\). For the TRIP7Mn, the predicted change in temperature at UTS at 100 s\(^{-1}\) is very close to the change in temperature predicted and experimentally measured for 10 s\(^{-1}\). Therefore, the effect of specimen self-heating on the flow stress and the austenite transformation rate is expected to be approximately the same between 10 and 100 s\(^{-1}\). The austenite to martensite transformation rate is the same at 10 and 100 s\(^{-1}\) for the TRIP7Mn as shown in Figure 4.13, which is consistent with similar amounts of adiabatic heating stabilizing the austenite. The average UTS at 10 and 100 s\(^{-1}\) for the TRIP7Mn in Figure 4.15a are the same. The lack of positive UTS strain rate sensitivity is unexpected, because DSA, adiabatic heating, and austenite stability are not expected to negatively impact strain rate sensitivity between 10 and 100 s\(^{-1}\). Based on the serrations in the flow curve and thermocouple measurements, the DSA effect seems to disappear between 0.1 and 1 s\(^{-1}\). The austenite to martensite transformation rate is the same between 10 and 100 s\(^{-1}\). Adiabatic heating
is predicted to be similar between tests at 10 and 100 s\(^{-1}\). The TRIP7Mn UTS strain rate sensitivity is low at high strain rates regardless of adiabatic heating, austenite stability, and DSA effects.

### 5.1.2 Role of Adiabatic Heating in High Strain Rate Deformation Behavior

The extent of the softening effect from adiabatic heating on flow stress was investigated with strain rate change and interrupted tensile tests. An engineering stress-strain curve from the strain rate change test and the temperature measured by a thermocouple at the center of the specimen are shown in Figure 5.4 for QP3Mn. During deformation at 0.1 s\(^{-1}\) between 1% and 5% strain, the specimen temperature increases by 15°C. When the strain rate is changed to 0.0005 s\(^{-1}\), the specimen cools by over 10°C before the next increase in strain rate. When the strain rate is increased between 0.0005 s\(^{-1}\) and 0.1 s\(^{-1}\) near 6% strain, there is a minimal amount of heating of the specimen and the effect of strain rate on flow stress can be observed without the softening effect of adiabatic heating. There is little heating of the specimen recorded by the thermocouple after the second increase in strain rate, because necking occurs close to 6% strain and the thermocouple was located away from the necked region. There is an initial abrupt increase in the flow stress (approximately 50 MPa) when the strain rate is increased from 0.0005 s\(^{-1}\) to 0.1 s\(^{-1}\) due to an increase in dislocation velocity necessary to match the higher imposed displacement rate. When the mobile dislocation density increases, the dislocation velocity decreases and the stress decreases [64]. Beyond the brief initial increase and decrease in flow stress, the flow stress at 0.1 s\(^{-1}\) is very similar to flow stress at 0.0005 s\(^{-1}\) (i.e. an extrapolation of the flow curve at 0.0005 s\(^{-1}\) would align closely with the flow curve at 0.1 s\(^{-1}\)). Thus, the flow stress is insensitive to strain rate between 0.0005 and 0.1 s\(^{-1}\).

The engineering stress-strain curves from the interrupted tests at 0.0005 s\(^{-1}\) and 0.1 s\(^{-1}\) for QP3Mn are shown in Figure 5.5. The change in temperature indicated on the plots is the difference in the specimen temperature at the very beginning of the test before any loading and the specimen temperature prior to unloading at either 4% or 6% strain. After unloading, the specimen was allowed to cool to room temperature prior to deformation to failure. Beyond the initial brief increase and decrease in stress that occurs upon reloading, the flow stress after the interruption is slightly higher (on the order of 20 MPa) than the flow stress prior to unloading for both the 4 and 6% strain tests at 0.0005 s\(^{-1}\). In contrast, there is a noticeable increase in flow stress of approximately 50 MPa upon reloading for both the 4 and 6% strain tests at 0.1 s\(^{-1}\). There is potentially a small amount of static strain aging that contributes to the higher flow stress upon reloading at both strain rates, but the change in

135
Figure 5.4  Engineering stress-strain curve with three changes in strain rate between 0.0005 s⁻¹ and 0.1 s⁻¹ for QP3Mn. The specimen temperature was measured with a K-type thermocouple spot welded to the center of the tensile specimen gage length.

flow stress is significantly higher for the 0.1 s⁻¹ tests. The greater changes in flow stress upon reloading for the 0.1 s⁻¹ tests compared to the 0.0005 s⁻¹ tests suggest that adiabatic heating at 0.1 s⁻¹ may have a slight softening effect on the flow stress (i.e. specimen self-heating had a softening effect on flow stress immediately prior to the interruption, then the specimen was allowed to cool, and upon reloading the flow stress was higher by approximately 50 MPa partly due to the lower specimen temperature). The UTS data from the quasi-static tensile tests at elevated temperature in Figure 4.8a show little sensitivity to temperature. The UTS data are a measure of flow stress at slightly different amounts of strain, but the difference in the average UE between 22 °C and 55 °C is only 1% strain. The discrepancy in the apparent temperature sensitivity of QP3Mn between these tests may be due in part to the test methodology. The interrupted tensile tests were performed on a single tensile specimen meaning that the microstructure, processing, and deformation history were identical when the specimen was unloaded and reloaded. The UTS data for the tests at different temperatures came from different specimens each with a unique initial microstructure (the initial microstructure between tensile specimens with the same processing will not be exactly identical), which contributes to scatter in the data. For example, the difference in the UTS data between the two tensile tests at the various elevated temperatures (40 – 85 °C) ranged from 4 to 19 MPa. A small sensitivity to temperature may be obscured by scatter in the data.
Figure 5.5  Engineering stress-strain curves for QP3Mn tensile tests interrupted at either 4% or 6% strain at two strain rates: (a) 4% strain, 0.0005 s\(^{-1}\), (b) 6% strain, 0.0005 s\(^{-1}\), (c) 4% strain, 0.1 s\(^{-1}\), and (d) 6% strain, 0.1 s\(^{-1}\).

The QP3Mn strain rate change data suggests that in the absence of adiabatic heating, flow stress near UTS has very low sensitivity to strain rate. The QP3Mn interrupted tensile test data at 0.1 s\(^{-1}\) suggests that adiabatic heating may have a slight softening effect on flow stress. The results of these two different tests suggest that it may be possible for adiabatic heating at 0.1 s\(^{-1}\) to cause a softening effect that would result in negative UTS strain rate sensitivity between 0.0005 and 0.1 s\(^{-1}\). However, the QP3Mn UTS versus strain rate data in Figure 5.1b show very little sensitivity to strain rate. A softening effect of adiabatic heating is more obvious in the DP980 and QP980 UTS data in Figure 5.1b. There is very little scatter in the UTS data for DP980 and QP980, and there is a slight but distinct negative strain rate sensitivity between 0.0001 s\(^{-1}\) and 0.1 s\(^{-1}\). A slight negative UTS strain rate sensitivity at low strain rates has also occurred for other high strength TRIP and martensitic steels in literature [140, 141, 144].
The relative contributions of the athermal and thermal components of flow stress and the opposing effects of strain rate and adiabatic heating on the thermal component of flow stress are depicted schematically for AHSS with different quasi-static strength levels in Figure 5.6. The effects of strain rate and specimen temperature on YS and UTS are depicted for a low strength steel, such as mild steel, in Figures 5.6a and 5.6d, for a medium strength steel in Figures 5.6b and 5.6e, and for a high strength steel, such as DP980 or QP980, in Figures 5.6c and 5.6f. The observed flow stress is comprised of an athermal component and

![Figure 5.6](image-url)

**Figure 5.6** Schematics depicting how the thermal and athermal contributions to flow stress may lead to the observed trends in YS and UTS with respect to strain rate based on the relative strength of the steel. The effects of temperature and strain rate on the thermal component of flow stress are separated into $\sigma_{\text{thermal}}(T)$ and $\sigma_{\text{thermal}}(\dot{\varepsilon})$. The UTS as a function of strain rate is shown schematically for (a) low strength, (b) medium strength, and (c) high strength steels. The YS as a function of strain rate is shown schematically for (d) low strength, (e) medium strength, and (f) high strength steels.
a thermal component, which is affected by temperature and strain rate. In the schematics, the thermal component is divided into two parts, $\sigma_{\text{thermal}}(T)$ and $\sigma_{\text{thermal}}(\dot{\varepsilon})$, to differentiate the effects of strain rate and temperature due to the specimen self-heating on the thermal component of flow stress. The athermal component of flow stress, $\sigma_{\text{athermal}}$, is insensitive to strain rate and temperature, and it is constant across the strain rate range. The $\sigma_{\text{athermal}}$ increases as the strength of the steel increases due to various strengthening mechanisms. For example, solid solution strengthening and composite strengthening with high strength bainite and/or martensite in multi-constituent microstructures lead to a decrease in strain rate sensitivity in AHSS [143]. Near yielding, the heating is negligible and there is no change in $\sigma_{\text{thermal}}(T)$ as a function of strain rate due to a constant specimen temperature regardless of strength level. However, the $\sigma_{\text{thermal}}$ is affected by strain rate, represented by the increase in $\sigma_{\text{thermal}}(\dot{\varepsilon})$, leading to positive strain rate sensitivity of the YS for all strength levels as shown in Figures 5.6d, 5.6e, and 5.6f. Near the UTS, the specimen temperature is elevated due to self-heating at strain rates above quasi-static rates. To simplify the schematics, it is assumed that the area under the stress-strain curve up to the UTS is approximately constant with respect to strain rate resulting in approximately the same amount of heat generated at all strain rates. Therefore, the specimen temperature at UTS increases with increasing strain rate until a complete adiabatic condition is reached (assumed to be near 0.1 s$^{-1}$). This rise in specimen temperature with increasing strain rate in the low strain rate regime causes a softening effect represented by the decrease in $\sigma_{\text{thermal}}(T)$ in Figures 5.6a, 5.6b, and 5.6c, and the heating would oppose the increase in flow stress due to increasing strain rate. Since the amount of heating is proportional to the area under the stress-strain curve, the self-heating of the tensile specimen may be higher for the high strength, high ductility AHSS compared to lower strength steels. This case would lead to a larger decrease in $\sigma_{\text{thermal}}(T)$ for the high strength steel depicted in Figure 5.6c compared to the medium and low strength steels depicted in Figures 5.6b and 5.6a. At low strain rates, the softening effect due to the heating may be greater than the increase in flow stress due to increasing strain rate for the high strength steel. The net effect would be a slight negative strain rate sensitivity of the UTS in the low strain rate regime (depicted in Figure 5.6c). At higher strain rates, the specimen temperature is approximately constant and the increase in flow stress due to increasing strain rate results in a positive strain rate sensitivity.

The roles of adiabatic heating and strain rate are more complicated for the TRIP7Mn due to the serrated flow behavior and the larger quantity of retained austenite in the microstructure that continually transforms to martensite up to 30% strain. The engineering stress-strain curve from the strain rate change test for the TRIP7Mn and the temperature measured by a thermocouple at the center of the specimen are shown in Figure 5.7. The
strain rate was increased from 0.0005 to 0.1 s\(^{-1}\) three times during the course of the test. When the strain rate increases in Figure 5.7, the flow stress is either slightly higher (by approximately 20 MPa) or close to the flow stress at the lower strain rate (again, considering flow stress beyond the brief increase and decrease in stress that immediately follows the instantaneous change in strain rate). The serrations in the flow curve also diminish when the strain rate is increased; the effect is most obvious from the first increase in strain rate near 14% strain. In Figure 4.15a, there is a decrease in the average UTS by 75 MPa between constant strain rate tensile tests at 0.0005 and 0.1 s\(^{-1}\). Yet, the increases in strain rate near 24% and 31% strain in Figure 5.7 do not produce a negative strain rate sensitivity. The lack of a negative strain rate sensitivity during the strain rate change test suggests that a combination of factors, not just the DSA alone, may be responsible for the TRIP7Mn negative UTS strain rate sensitivity in Figure 4.15a. The heating is low during the strain rate change test (specimen reaches less than 55 °C during the 0.1 s\(^{-1}\) portions of the test), and the specimen cools to below 30 °C during the 0.0005 s\(^{-1}\) portions of the test. Therefore, there is little effect of adiabatic heating on austenite stability during the strain rate change test compared to a test with a constant strain rate of 0.1 s\(^{-1}\). The stabilization of austenite due to adiabatic heating during tests at constant strain rates is a contributing factor for the observed negative UTS strain rate sensitivity for tests at various, constant strain rates.

Figure 5.7  Engineering stress-strain curve with six changes in strain rate between 0.0005 s\(^{-1}\) and 0.1 s\(^{-1}\) for TRIP7Mn. The specimen temperature was measured with a K-type thermocouple spot welded to the center of the tensile specimen gage length.
The engineering stress-strain curves from the interrupted tests at 0.0005 s\(^{-1}\) and 0.1 s\(^{-1}\) for the TRIP7Mn steel are shown in Figure 5.8. The change in temperature indicated on the plots is the difference in the specimen temperature at zero load when the test was initiated and the specimen temperature prior to unloading. After unloading, the specimen was allowed to cool to room temperature prior to deformation to failure. In the 0.0005 s\(^{-1}\) tests in Figures 5.8a, 5.8b, and 5.8c, there are no significant changes in flow stress upon reloading that can be discerned from the already present serrations in the flow curve. In the 0.1 s\(^{-1}\) tests in Figures 5.8d, 5.8e, and 5.8f, the flow stress increases and there is YPE upon reloading. The YPE appears to increase as the interruption strain increases. The increase in flow stress and the appearance of YPE upon reloading is indicative of a static strain aging response. It is difficult to differentiate the effect of the specimen temperature change during the interruption on the flow stress from the increase in the flow stress due to the static strain aging. Static strain aging likely occurred in the 0.0005 s\(^{-1}\) tests as well, but an upper yield point and YPE upon reloading are very challenging to distinguish from the serrations in the flow curve prior to unloading.

The retained austenite fractions as a function of plastic strain are compared at a quasi-static strain rate and 0.1 s\(^{-1}\) for QP980, QP3Mn, and TRIP7Mn in Figure 5.9a. The austenite is slightly more stable in the QP980 than the QP3Mn. For example, the austenite fraction decreases from an initial 11 vol.% to 5 vol.% at 11% strain for QP980, and the austenite fraction decreases from an initial 14 vol.% to 4 vol.% at 9.5% strain for QP3Mn. For both steels, there is no significant difference in transformation rate between the two strain rates. The TRIP7Mn retained austenite measurements were made beyond the YPE starting at 8 – 9% strain, which corresponds to when the austenite transformation rate diminishes in the QP980 and QP3Mn. The austenite transformation rate is lower at 0.1 s\(^{-1}\) for the TRIP7Mn. For the TRIP7Mn, the austenite fraction at 8% strain for the strain rate 0.0005 s\(^{-1}\) is 22 vol.% and the austenite fraction at 9% strain for the strain rate 0.1 s\(^{-1}\) is 21 vol.%. With increasing strain the difference in retained austenite content between the two strain rates ranges from 5 vol.% at 15% strain to 7 vol.% at 30% strain. The specimen temperatures as a function of strain during the test up to UTS for all three steels at 0.1 s\(^{-1}\) are shown in Figure 5.9b. The TRIP7Mn specimen temperature up to UTS far exceeds the QP3Mn and QP980. The specimen temperature at UTS is 50 °C for the QP980, 61 °C for the QP3Mn, and 97 °C for the TRIP7Mn. For the QP980 and QP3Mn, the specimen self-heating does not alter the transformation rate. The strain-induced austenite to martensite transformation in the TRIP7Mn occurs at much greater strain, and the specimen heats up to 50 °C before the austenite transforms post-YPE. The higher adiabatic heating can stabilize retained austenite
Figure 5.8  Engineering stress-strain curves for TRIP7Mn tensile tests interrupted at either 15%, 20%, or 25% strain at two strain rates: (a) 15% strain, 0.0005 s\(^{-1}\), (b) 20% strain, 0.0005 s\(^{-1}\), (c) 25% strain, 0.0005 s\(^{-1}\), (d) 15% strain, 0.1 s\(^{-1}\), (e) 20% strain, 0.1 s\(^{-1}\), and (f) 25% strain, 0.1 s\(^{-1}\).
and reduce the austenite transformation rate above approximately 10% strain. The role of specimen temperature in stabilizing austenite during high strain rate tensile tests is further examined through Olson-Cohen modeling of the austenite transformation kinetics in Section 5.2.

![Graphs showing retained austenite and temperature as a function of strain](image)

Figure 5.9  (a) Retained austenite as a function of plastic engineering strain for QP980, QP3Mn, and TRIP7Mn at a quasi-static and a high strain rate. (b) Temperature as a function of engineering strain for QP980, QP3Mn, and TRIP7Mn at 0.1 s\(^{-1}\). Temperature was measured by a K-type thermocouple spot welded to the center of the gage length.

### 5.2 Modeling the Austenite to Martensite Transformation

Recently there have been efforts to develop constitutive models for TRIP steels that incorporate the effect of strain rate [168, 170, 171]. These models incorporate micromechanical and kinetics models for the austenite to martensite transformation into the constitutive model for flow stress, and the computation of the models requires finite element analysis. The goal in this work is to utilize the Olson-Cohen model for the kinetics of the austenite to martensite transformation to gain an understanding of the extent of stabilization of the retained austenite due to specimen temperature. The Olson-Cohen model was applied to the QP3Mn and TRIP7Mn retained austenite data from the tensile tests at different elevated temperatures and a quasi-static strain rate to identify how the \( \alpha \) and \( \beta \) parameters change with temperature. An attempt was made to incorporate changes in \( \alpha \) and \( \beta \) due to the increase in specimen temperature during a high strain rate tensile test into the Olson-Cohen
model to predict the effects of adiabatic heating on austenite stability at high strain rates. The modified Olson-Cohen model is presented in Appendix G, because the model does not completely agree with experimental data. Appendix G contains equations for specimen temperature as a function of strain and for $\alpha$ and $\beta$ as a function of temperature. The equation for the modified Olson-Cohen model and results from the model at 0.0005 and 0.1 s$^{-1}$ for both steels are also included. The modified Olson-Cohen model shows some agreement with experimental observations for the QP3Mn and for data between approximately 8% and 10% strain for TRIP7Mn (8% strain is at the end of the YPE region) but does not agree with experimental observations at high strain levels for TRIP7Mn.

Nonlinear least squares regression was used to fit the Olson-Cohen model for strain-induced austenite transformation, given in Equation 2.4, to the experimental data from the quasi-static, elevated temperature tensile tests for both steels. The experimental data and model curves for TRIP7Mn are shown in Figure 5.10a. The closest fits were found for $n$ equal to 2, and the model curves agree well with the experiment data. The value $n$ equal to 2 is the lower bound for the parameter $n$ based on the assumption that the austenite grains are equiaxed and that the shear bands are randomly oriented with a length equal to the austenite grain size [99]. The model parameters $\alpha$ and $\beta$ are plotted as a function of temperature in Figure 5.10b. The trends in the $\alpha$ and $\beta$ parameters with respect to temperature are similar.
to the trends found for 304 stainless steel by Olson and Cohen [99]. Further, the values of $\alpha$ and $\beta$ over the investigated temperature range are close to the values shown in Figure 2.12b obtained by Samek et al. for a TRIP alloy with lower Mn and higher Al [107]. For the 7.1Mn-0.1C-0.12Si (wt.%) steel room temperature, quasi-static tensile test data in Figure 2.6b, the application of the Olson-Cohen model to the austenite data produced an $\alpha$ of 5.75 and a $\beta$ of 2.5 for $n$ equal to 3 for the 600 °C condition and an $\alpha$ of 18.6 and a $\beta$ of 1.86 for $n$ equal to 2 for the 625 °C condition [48]. For comparison, the TRIP7Mn has an $\alpha$ of 9.2 and a $\beta$ of 2.3 for $n$ equal to 2 for the test performed at 22 °C. The trend in $\alpha$ with respect to temperature suggests that the rate of shear band formation decreases rapidly with increasing temperature up to approximately 70 °C after which the rate declines slowly. For example, $\alpha$ decreases by 4.1 between 22 and 70 °C and by 0.6 between 70 and 115 °C. The parameter $\beta$, which is proportional to the probability of martensite forming at a shear band intersection, is nearly constant up to 70 °C and then decreases rapidly with increasing temperature. For example, $\beta$ decreases by 0.5 between 22 and 70 °C and by 1.2 between 70 and 115 °C. These trends indicate that for TRIP7Mn between 22 and 70 °C the rate of shear band formation decreases rapidly while the probability of forming martensite at shear band intersections is constant. Above 70 °C, the rate of shear band formation is lower and continues to decline slowly with increasing temperature, but the probability of forming martensite at shear band intersections decreases quickly. By 115 °C, both the rate of shear band formation and the probability of forming martensite are low compared to the lower temperature conditions.

The Olson-Cohen model fitting to the QP3Mn experimental data from the quasi-static, elevated temperature tensile tests resulted in poorer fits to the data compared to the TRIP7Mn. A value of $n$ equal to 2 provided the closest fit between the Olson-Cohen model and the experimental data (among $n$ values greater than or equal to 2), and the results are shown in Figure 5.11a. The model curve fits agree with most of the data for temperatures up to 70 °C. The curves fits are poor at 1% strain for all of the temperatures with the fits being the poorest for 70 and 85 °C. The poor fit of the Olson-Cohen model for most of the 85 °C data is due to the following trend in the austenite transformation rate as a function of strain: the fraction of martensite is constant between 1 and 4% strain and then sharply increases from 17 to 37 vol.% transformed martensite between 4% and 6% plastic strain. The Olson-Cohen model cannot predict the curvature necessary to fit that trend in the data. It is possible that a portion of the austenite has very different stability than the rest of the austenite. For example, the difference in morphology between blocky and lath-like austenite could be responsible for disparities in stability that become apparent with increased temperature. The $\alpha$ and $\beta$ parameters for $n$ equal to 2 are plotted as a function of test temperature in Figure 5.11b. The $\alpha$ values are similar between 22 and 55 °C and decrease between 55 and
Figure 5.11 Fraction of transformed martensite as a function of plastic true strain at various test temperatures for QP3Mn for (a) n equal to 2 and (c) n equal to 1. The data points are from the experiments and the curves are the Olson-Cohen model. (b) The Olson-Cohen model parameters $\alpha$ and $\beta$ as a function of temperature for (b) n equal to 2 and (d) n equal to 1. The curves in (b) are polynomial fits. In (d), the axis for $\beta$ is divided to show the extremely high $\beta$ values from the model fits for 70 and 85 °C. A linear fit for $\beta$ and a line showing the average for $\alpha$ are shown in (d) for 20 – 60 °C.

For example, $\alpha$ is nearly constant with an average of 47 between 22 and 55 °C and decreases by 25 between 55 and 85 °C. The $\beta$ values decrease with increasing temperature up to 55 °C and increase slightly between 55 and 85 °C. For example, $\beta$ decreases by 0.5 between 22 and 55 °C and increases by 0.1 between 55 and 85 °C. De Moor et al. calculated
\(\alpha\) and \(\beta\) parameters from room temperature, quasi-static tensile tests for a 0.17C-1.65Mn-0.38Si-1.11Al-0.08P (wt.\%) steel that was subjected to Q&P heat treatments with different partitioning temperatures (PT) to produce different amounts of austenite in the final microstructures [43]. The highest PT (450 °C) condition with initially 15 vol.% austenite had an \(\alpha\) equal to 21 and a \(\beta\) equal to 1.02. The lowest PT (350 °C) condition with initially 8 vol.% austenite had an \(\alpha\) equal to 71 and a \(\beta\) equal to 0.46. For comparison, QP3Mn has an \(\alpha\) equal to 49 and a \(\beta\) equal to 1.1 at 22 °C.

The curve fits for \(n\) equal to 1 are shown in Figure 5.11c, because they produced a slightly lower sum of differences squared between the experimental data and the model curve fits. A value of \(n\) equal to 1 would mean that the number of shear band intersections is directly proportional to the number of shear bands, which is not likely with increasing strain. However, a value of \(n\) equal to 1 can provide a reasonable fit to the experiment data. The \(\alpha\) and \(\beta\) parameters for \(n\) equal to 1 are shown in Figure 5.11d. Between 22 and 55 °C, the average value of \(\alpha\) is 18 and \(\beta\) decreases from 1.4 to 0.8. For 70 and 85 °C, the values of \(\alpha\) and \(\beta\) are completely different from the lower temperatures; the \(\alpha\) values are less than 0.15, and the \(\beta\) values are 55 and 64.

5.3 Discussion Summary

The YS strain rate sensitivities are similar between DP980, QP980, and TRIP7Mn, especially at strain rates below 10 s\(^{-1}\). The response of YS to strain rate for these three steels seems to be more closely linked to their high quasi-static YS and associated large athermal components of flow stress than differences in microstructure. Microstructure may only be responsible for the small differences in strain rate sensitivity between the three steels. The QP3Mn has unexpectedly high positive YS strain rate sensitivity at strain rates below 10 s\(^{-1}\) and very low strain rate sensitivity at high strain rates. There are similarities in the trend of UTS with respect to strain rate between the DP980, QP980, and QP3Mn. The UTS strain rate sensitivity is low below 10 s\(^{-1}\), and there is a low positive UTS strain rate sensitivity at higher strain rates. The DP980 and QP980 exhibit slight negative strain rate sensitivity at low strain rates likely due to adiabatic heating; whereas, the QP3Mn has almost no strain rate sensitivity at low strain rates. In contrast to the DP980, QP980, and QP3Mn, strain rate has a major effect on UTS for the TRIP7Mn. The TRIP7Mn exhibits significant negative strain rate sensitivity from 0.0001 – 1 s\(^{-1}\) due to a combination of austenite stabilization from adiabatic heating, a diminishing effect of DSA, and a small softening effect on flow stress due to adiabatic heating.
Modeling of the change in specimen temperature at UTS combined with \textit{in situ} specimen temperature measurements show that complete adiabatic heating conditions are reached at approximately 0.1 s\(^{-1}\) for sheet steel with thicknesses in the range of 1.1 – 1.4 mm. For high strength steel, adiabatic heating is approximately constant over several orders of magnitude of strain rate due to very similar areas under the stress-strain curve. For the TRIP7Mn, the approximately constant amount of heating at high strain rates leads to austenite to martensite transformation rates that are similar to each other over the strain rate range 0.1 – 100 s\(^{-1}\) (i.e. adiabatic heating causes a similar degree of austenite stabilization at strain rates from 0.1 to 100 s\(^{-1}\)). The temperature model predicts that the heating for the QP3Mn is slightly higher at 100 s\(^{-1}\) compared to lower strain rates, but the effect on flow stress is expected to be minimal since the QP3Mn flow stress is not very sensitive to temperature.

The QP3Mn strain rate change test indicates that in the absence of adiabatic heating, flow stress near UTS is insensitive to strain rate, and the QP3Mn interrupted tensile test data at 0.1 s\(^{-1}\) suggests that adiabatic heating may have a slight softening effect on flow stress. These results suggest that it may be possible for adiabatic heating at 0.1 s\(^{-1}\) to cause a softening effect that would result in negative UTS strain rate sensitivity between 0.0005 and 0.1 s\(^{-1}\). A negative UTS strain rate sensitivity was not observed for the QP3Mn UTS data from tests strain to fracture at constant strain rate, but the results help to the explain the slight negative UTS strain rate sensitivity observed for DP980 and QP980 in the range 0.0005 – 0.1 s\(^{-1}\). The low flow stress strain rate sensitivity observed in the TRIP7Mn strain rate change test indicates that the stabilization of austenite due to adiabatic heating at high strain rates is a contributor to the negative UTS strain rate sensitivity observed for tensile tests at constant strain rate. Additionally, the serrations in the flow curve diminish when the strain rate increases from 0.0005 to 0.1 s\(^{-1}\) in the strain rate change test and in the constant strain rate tests, which indicate that strain rate has a strong effect on DSA. A strong strain aging effect was observed for the TRIP7Mn interrupted tensile tests.

The Olson-Cohen model was used to fit the austenite to martensite transformation kinetics of QP3Mn and TRIP7Mn from the elevated temperature quasi-static tensile tests. The model shows good agreement with the TRIP7Mn experimental data beyond the YPE region. The Olson-Cohen model shows good agreement with the QP3Mn experimental data above approximately 2% strain. The \(\alpha\) and \(\beta\) parameters generally decrease with increasing test temperature.
The effect of strain rate on the deformation behavior of two experimental third generation AHSS grades, QP3Mn and TRIP7Mn, was examined through tensile tests in the range of 0.0001 – 200 s\(^{-1}\) and retained austenite measurements from interrupted tensile tests at various strain rates. The effect of adiabatic heating on flow stress and retained austenite stability was studied through specimen temperature measurements, temperature modeling, quasi-static tensile tests at elevated temperature, retained austenite measurements from interrupted tensile tests at elevated temperature, and modeling of the austenite to martensite transformation. The combined effects of temperature and strain rate on flow stress and the austenite to martensite transformation were also examined. The effects of strain rate on two commercial steels, DP980 and QP980, were studied for comparison to the results of the experimental third generation AHSS grades. A pre-strain study was also conducted with the DP980 and QP980 steels to understand how prior deformation can affect high strain rate tensile behavior.

The conclusions drawn from this work are:

- During high strain rate tensile tests (up to 100 s\(^{-1}\)) of advanced high strength sheet steel, the austenite to martensite transformation rate can be reduced if the transformation occurs at high levels of strain (greater than approximately 10% strain), because heat builds up in the specimen and stabilizes the austenite. If most of the austenite transformation occurs at very low strain (below approximately 5% strain), then the self-heating of the tensile specimen is insufficient to have an observable stabilizing effect on the austenite.

- Austenite stability, when modified by test temperature (-10 to 70 ºC), has a small impact for TRIP7Mn and no impact for QP3Mn on the effect of strain rate on the austenite to martensite transformation. For TRIP7Mn, adiabatic heating causes an increase in austenite stability during high strain rate tests at all test temperatures, but the austenite stabilization is greatest at 70 ºC; whereas, for QP3Mn, the austenite to martensite transformation rate is the same at low and high strain rates for all test temperatures.

- There are major differences in the effect of strain rate on the tensile mechanical properties of a medium Mn TRIP steel like TRIP7Mn compared to most other AHSS grades.
due to the disappearance of discontinuous serrated plastic flow and the stabilization of austenite at high strain rates. Austenite stabilization due to adiabatic heating at high strain rates is the major cause for strong negative UTS strain rate sensitivity and reductions in work hardening rate and ductility with increasing strain rate up to 10 s\(^{-1}\).

- Retained austenite may contribute to positive YS strain rate sensitivity of Q&P steels containing only austenite and martensite as demonstrated by the strong positive strain rate sensitivity of QP3Mn. At UTS when most of the austenite has transformed to martensite, the UTS strain rate sensitivity is low and reflects the mostly martensitic microstructure.

- For the DP980 and QP980 equivalent strength steels, a small amount of retained austenite in the QP980 microstructure leads to slightly higher YS strain rate sensitivity but does not affect trends in elongation and UTS with respect to strain rate. The transformation of most of the austenite to martensite during plastic deformation and the insensitivity of the transformation to strain rate both contribute to the similar UTS strain rate sensitivities between DP980 and QP980.

- In high strength steel with sufficient ductility, it is possible for specimen self-heating to produce a softening effect greater than the strengthening effect of increasing strain rate on UTS at low strain rates (below approximately 0.1 s\(^{-1}\)). The softening effect leads to zero or slightly negative strain rate sensitivity and can occur in AHSS regardless of the presence of austenite.
CHAPTER 7

FUTURE WORK

The potential areas for further experimental work on the high strain rate behavior of third generation AHSS include: advanced characterization of deformed microstructures from tensile tests at various strain rates and investigating the effect of stress state and higher strain rates on the austenite to martensite transformation. There are also several ways to continue modeling efforts of the high strain rate tensile deformation behavior of third generation AHSS including more complex temperature modeling, further developing the Olson-Cohen model for the effect of strain rate on the austenite to martensite transformation, and constitutive modeling of flow stress. Finally, the effect of strain rate on ductility and necking behavior are challenging to study with strain gages that detach during the high strain rate tensile test; therefore, the use of a high speed stereo digital image correlation (DIC) system is recommended for a more in-depth investigation of the effect of strain rate on ductility of AHSS.

Further microstructural characterization may provide more insight into the mechanisms behind the observed mechanical behavior of QP3Mn and TRIP7Mn with respect to strain rate and temperature. TEM of the deformation structures, particularly in the retained austenite, may provide experimental evidence for how strain rate and temperature affect the rate of shear band formation, the number of shear band intersections, and the type of shear band intersections that lead to \( \alpha \) martensite formation in these steels. For the TEM analysis, potentially a large number of different views of deformation structures from the same sample and phase would need to be examined. It may be challenging to find untransformed austenite in the microstructure, since the untransformed austenite comprises a small fraction of the microstructure in interrupted tensile test specimens.

The effect of strain rate on the austenite to martensite transformation could be examined at strain rates above 100 s\(^{-1}\) and for conditions with more complex stress states. At higher strain rates, specimen heating may be greater and the mechanical driving force could potentially increase (e.g. through the formation of more shear band intersections). The effect of strain rate on the order of 1000 s\(^{-1}\) on the austenite to martensite transformation is relevant to the performance of the steel in crash scenarios. Additionally, the stress state during deformation can affect the austenite to martensite transformation [105, 172]. The effect of stress state on the transformation is relevant to forming operations.
There are several ways to continue modeling efforts for the high strain rate deformation behavior of third generation AHSS. The specimen temperature model used in this work assumes uniform deformation of the gage length, but uniform deformation does not always occur during a tensile test (e.g. discontinuous plastic deformation of TRIP7Mn at low strain rates or the potential to form multiple necks at high strain rates). It may be useful to develop the specimen temperature model further to account for localized plastic deformation that causes a temperature gradient in the tensile specimen. The modified Olson-Cohen model that incorporates adiabatic heating in the prediction of the fraction of martensite as a function of strain at high strain rates could be further refined to account for how prior deformation in the austenite affects the instantaneous transformation kinetics during a high strain rate tensile test. A constitutive model of the flow stress that incorporates the effect of strain rate could be developed for the various grades of third generation AHSS. The application of a mechanism-based constitutive model for dynamic mechanical behavior, such as the Zerilli and Armstrong model, for third generation AHSS may lead to a better understanding of the magnitude of the athermal and thermal contributions to the flow stress and the contributions of individual micro-constituents to the flow stress strain rate sensitivity \cite{173, 174}. An existing constitutive model would need to be further developed to incorporate the composite-like behavior of the multi-constituent microstructure, to account for the austenite to martensite transformation, and to account for the mechanism of serrated plastic flow that has been observed for various grades of third generation AHSS. A successful constitutive model could be used to predict how other potential third generation AHSS microstructures may behave with respect to strain rate.

Finally, the use of a high speed stereo DIC system would be highly beneficial for measuring strain during high strain rate tensile tests, because strain gages detach during the tensile test and cannot record strain during necking. The current method of estimating strain from the detached strain gage introduces a source of error to the UE and TE values. A DIC system with two high speed cameras would allow for complete strain mapping of the gage length throughout the high strain rate tensile test. This system has several major advantages over strain gages: strain can be measured throughout the entire test, necking behavior can be studied accurately including the formation of multiple necks, and strain can be accurately mapped and studied for materials that exhibit yield point elongation and/or serrated plastic flow.
REFERENCES CITED


Quenching and partitioning heat treatment experiments were performed on tensile specimens of the cold-rolled QP3Mn steel. The goal of the heat treatment experiments was to select a heat treatment that produced 10% or more retained austenite and resulted in greater than 10% total elongation. Four heat treatments were chosen based on the results of De Moor et al. on an alloy with similar amounts of carbon, manganese, and silicon [28]. The heat treatment profile is given in Figure A.1. Salt pots were used for all three processing steps. All specimens were austenitized at 830 °C for 120 s and then were held at the quench temperature for 10 s to allow for thermal equilibration. The specimens were then heated to the partitioning temperature and held for a specific amount of time. Each specimen was heat treated with a different combination of quench temperature (QT), partitioning temperature (PT), and partitioning time (Pt). The four different heat treatments listed by QT/PT/Pt are 165 °C/400 °C/100 s, 180 °C/400 °C/10 s, 180 °C/400 °C/150 s, and 180 °C/430 °C/100 s.

The engineering stress-strain curves from the four heat treatments are shown in Figure A.2. All of the stress-strain curves exhibit continuous yielding, very low work hardening beyond 2% strain, UTS near or above 1450 MPa, TE near or above 13% strain, and approximately 3% strain or greater post-uniform elongation. Tensile specimens from the three heat treatments with partitioning times greater than or equal to 100 s have flow curves with more defined yield points followed by little work hardening. The tensile curve from the heat
treatment with the 10 s partitioning time exhibits more gradual continuous yielding with greater work hardening below 5% strain compared to the tensile curves from the heat treatments with 100 s partitioning times. The greater work hardening of the 10 s partitioning time specimen contributes to the highest tensile strength of the four heat treatments.

![Engineering Stress-Strain Curves](image)

**Figure A.2** Engineering stress-strain curves for four different salt pot heat treatments of the cold-rolled 0.3C-3Mn-1.6Si (wt.%) steel. All samples were austenitized at 830 °C for 120 s prior to quenching and partitioning. QT is quench temperature, PT is partitioning temperature, and Pt is partitioning time.

The mechanical properties and retained austenite measurements from the tensile tests for the four heat treatments are summarized in Table A.1. The four heat treatments produce retained austenite amounts in the range of 10 to 14 vol.%. The retained austenite content is the highest at 14 vol.% for the $165 °C/400 °C/100$ s and $180 °C/400 °C/10$ s heat treatments, which also have higher tensile strengths than the other two heat treatments. All four heat treatments result in total elongations greater than 10% strain. The $180 °C/400 °C/10$ s heat treatment was selected for this work, because it produces a high amount of retained austenite at 14 vol.%, a high UTS at 1580 MPa, and total elongation greater than 10% strain. The $180 °C/400 °C/10$ s heat treatment was selected over the $165 °C/400 °C/100$ s, because the quenching salt pot begins to solidify at $165 °C$ making the heat treatment challenging to perform and the short 10 s partitioning time minimizes precipitation of carbides during partitioning [35, 39].
Table A.1 – Mechanical properties and retained austenite produced by Q&P processing of cold-rolled 0.3C-3Mn-1.6Si (wt.%) steel.

<table>
<thead>
<tr>
<th>QT (°C)</th>
<th>PT (°C)</th>
<th>Pt (s)</th>
<th>0.2% YS (MPa)</th>
<th>UTS (MPa)</th>
<th>UE (%)</th>
<th>TE (%)</th>
<th>RA (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>165</td>
<td>400</td>
<td>100</td>
<td>1304</td>
<td>1504</td>
<td>9.7</td>
<td>14.8</td>
<td>14</td>
</tr>
<tr>
<td>180</td>
<td>400</td>
<td>10</td>
<td>1173</td>
<td>1580</td>
<td>9.4</td>
<td>12.8</td>
<td>14</td>
</tr>
<tr>
<td>180</td>
<td>400</td>
<td>150</td>
<td>1342</td>
<td>1497</td>
<td>9.7</td>
<td>13.6</td>
<td>12</td>
</tr>
<tr>
<td>180</td>
<td>430</td>
<td>100</td>
<td>1340</td>
<td>1454</td>
<td>8.9</td>
<td>13.2</td>
<td>10</td>
</tr>
</tbody>
</table>
APPENDIX B
TEMPERATURE DATA

The specimen temperature was measured \emph{in situ} with a K-type thermocouple and with a high speed infrared camera during tensile tests at various strain rates up to 10 s\(^{-1}\) for QP980, QP3Mn, and TRIP7Mn. The specimen temperature, measured by the thermocouple, is plotted as a function of strain with the corresponding stress-strain curve for select low strain rate conditions for QP3Mn and TRIP7Mn. Beyond UTS at the low strain rates there was sometimes a decrease in temperature if necking occurred away from the thermocouple. A temperature spike occurred when the specimen fractured, but the magnitude of the recorded temperature spike depended on how close the thermocouple was to the fracture site. Example thermal camera images for all three steels are also provided, and the thermal camera images show a temperature map of the gage length during the tensile test and post-fracture (only one of specimen halves can be viewed post-fracture).

Prof. Greg Walker from Vanderbilt University is gratefully acknowledged for providing the thermal camera. Jake Benzing from Vanderbilt University is gratefully acknowledged for collaborating on the thermal camera experiments.

B.1 QP3Mn Temperature Measurements

The specimen temperature measured by thermocouple as a function of strain and the corresponding engineering stress-strain curve for QP3Mn tensile tests at 0.0005, 0.001, 0.01, and 0.1 s\(^{-1}\) are shown in Figure B.1. Thermal camera images taken during a tensile test of QP3Mn at 1 s\(^{-1}\) are shown in Figure B.2.

B.2 TRIP7Mn Temperature Measurements

The specimen temperature measured by thermocouple as a function of strain and the corresponding engineering stress-strain curve for TRIP7Mn tensile tests at 0.0005, 0.001, 0.01, and 0.1 s\(^{-1}\) are shown in Figure B.3. Thermal camera images taken during a tensile test of TRIP7Mn at 1 s\(^{-1}\) are shown in Figure B.4.
Figure B.1 The engineering stress-strain curve and specimen temperature as a function of strain for QP3Mn tensile tests at (a) 0.0005 s\(^{-1}\), (b) 0.001 s\(^{-1}\), (c) 0.01 s\(^{-1}\), and (d) 0.1 s\(^{-1}\). The stepwise behavior of the temperature measurement at 0.01 s\(^{-1}\) and 0.1 s\(^{-1}\) is due to the limited DAQ rate of the thermocouple reader. Temperature was measured with a K-type thermocouple in the center of the gage length.

B.3 QP980 Temperature Measurements

Thermal camera images taken during tensile tests of QP980 at 0.1 s\(^{-1}\) and 10 s\(^{-1}\) are shown in Figure B.5.
Figure B.2 Thermal camera images from a QP3Mn test at 1 s$^{-1}$. The maximum of the calibrated thermal camera range was 150 °C and the maximum temperature that could be recorded by the software was 160.2 °C. The temperature reading reached the maximum at the fracture surface. The bar to the right of the images shows the temperature color scale in degrees Celsius. (Color image - see PDF version)
Figure B.3 The engineering stress-strain curve and specimen temperature as a function of strain for TRIP7Mn tensile tests at (a) 0.0005 s\(^{-1}\), (b) 0.001 s\(^{-1}\), (c) 0.01 s\(^{-1}\), and (d) 0.1 s\(^{-1}\). Temperature was measured with a K-type thermocouple in the center of the gage length.
Figure B.4 Thermal camera images from a TRIP7Mn test at 1 s\(^{-1}\). The image at 0.110 s occurs near the end of YPE. The maximum of the calibrated thermal camera range was 150 °C and the maximum temperature that could be recorded by the software was 160.2 °C. The temperature reading reached the maximum at the fracture surface. The bar to the right of the images shows the temperature color scale in degrees Celsius. (*Color image - see PDF version*)
Figure B.5  Thermal camera images of QP980 tensile tests at (a) 0.1 s\(^{-1}\) and (b) 100 s\(^{-1}\). The maximum of the calibrated thermal camera range was 150 °C and the maximum temperature that could be recorded by the software was 160.2 °C. The temperature reading reached the maximum at the fracture surface. The bars to the right of the images show the temperature color scale in degrees Celsius. (Color image - see PDF version)
APPENDIX C

STRESS-STRAIN DATA

Mechanical properties from true stress-strain curves for all of the tensile tests at room temperature and various strain rates for the QP3Mn and TRIP7Mn are summarized. The average true 0.2% offset YS, UTS, and UE for QP3Mn are reported in Table C.1. The average true upper YS, lower YS, UTS, and UE for TRIP7Mn are reported in Table C.2. The lower YS for the TRIP7Mn is the minimum in flow stress immediately following the upper yield point. The values reported in the tables are averaged from three tests per strain rate, and one standard deviation of the data is also reported for each average value. The uncertainty of the flow stress measurement is 7 MPa. The uncertainty of the elongation measurement is 0.1%.

Table C.1 – Average true 0.2% offset YS, UTS, and UE for QP3Mn at various strain rates. Three tests were performed for each strain rate, and one standard deviation is reported for each average value.

<table>
<thead>
<tr>
<th>Strain Rate (s⁻¹)</th>
<th>0.2% Offset YS (MPa)</th>
<th>UTS (MPa)</th>
<th>UE (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.0001</td>
<td>1096 ± 4</td>
<td>1712 ± 17</td>
<td>9.4 ± 0.2</td>
</tr>
<tr>
<td>0.0005</td>
<td>1214 ± 52</td>
<td>1703 ± 23</td>
<td>8.7 ± 0.2</td>
</tr>
<tr>
<td>0.001</td>
<td>1145 ± 56</td>
<td>1739 ± 15</td>
<td>8.8 ± 0.4</td>
</tr>
<tr>
<td>0.01</td>
<td>1167 ± 93</td>
<td>1693 ± 20</td>
<td>8.2 ± 0.7</td>
</tr>
<tr>
<td>0.1</td>
<td>1231 ± 28</td>
<td>1689 ± 5</td>
<td>8.1 ± 0.3</td>
</tr>
<tr>
<td>1</td>
<td>1314 ± 29</td>
<td>1675 ± 9</td>
<td>6.9 ± 0.6</td>
</tr>
<tr>
<td>10</td>
<td>1327 ± 27</td>
<td>1707 ± 13</td>
<td>8.0 ± 0.7</td>
</tr>
<tr>
<td>50</td>
<td>1247 ± 64</td>
<td>1779 ± 16</td>
<td>9.1 ± 0.2</td>
</tr>
<tr>
<td>100</td>
<td>1284 ± 29</td>
<td>1790 ± 28</td>
<td>9.6 ± 0.7</td>
</tr>
<tr>
<td>200</td>
<td>1328 ± 63</td>
<td>1775 ± 16</td>
<td>8.2 ± 0.8</td>
</tr>
</tbody>
</table>
Table C.2 – Average true upper YS, lower YS, UTS, and UE for TRIP7Mn at various strain rates. Three tests were performed for each strain rate, and one standard deviation is reported for each average value.

<table>
<thead>
<tr>
<th>Strain Rate (s(^{-1}))</th>
<th>Upper YS (MPa)</th>
<th>Lower YS (MPa)</th>
<th>UTS (MPa)</th>
<th>UE (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.0001</td>
<td>775 ± 5</td>
<td>727 ± 6</td>
<td>1420 ± 4</td>
<td>30.4 ± 1.0</td>
</tr>
<tr>
<td>0.0005</td>
<td>809 ± 4</td>
<td>722 ± 29</td>
<td>1424 ± 43</td>
<td>30.6 ± 0.5</td>
</tr>
<tr>
<td>0.001</td>
<td>806 ± 7</td>
<td>743 ± 3</td>
<td>1399 ± 6</td>
<td>31.8 ± 0.5</td>
</tr>
<tr>
<td>0.01</td>
<td>800 ± 20</td>
<td>743 ± 10</td>
<td>1333 ± 17</td>
<td>28.9 ± 2.5</td>
</tr>
<tr>
<td>0.1</td>
<td>802 ± 8</td>
<td>748 ± 2</td>
<td>1283 ± 25</td>
<td>27.6 ± 0.9</td>
</tr>
<tr>
<td>1</td>
<td>853 ± 35</td>
<td>783 ± 6</td>
<td>1254 ± 16</td>
<td>27.3 ± 0.2</td>
</tr>
<tr>
<td>10</td>
<td>878 ± 8</td>
<td>782 ± 16</td>
<td>1249 ± 29</td>
<td>27.1 ± 0.4</td>
</tr>
<tr>
<td>50</td>
<td>866 ± 27</td>
<td>796 ± 11</td>
<td>1268 ± 12</td>
<td>27.5 ± 0.3</td>
</tr>
<tr>
<td>100</td>
<td>927 ± 9</td>
<td>818 ± 17</td>
<td>1264 ± 5</td>
<td>27.8 ± 0.1</td>
</tr>
<tr>
<td>200</td>
<td>969 ± 30</td>
<td>848 ± 16</td>
<td>1265 ± 14</td>
<td>25.2 ± 0.8</td>
</tr>
</tbody>
</table>
Examples of the relative X-ray intensity versus 2-theta from an as-processed condition and after various amounts of strain from quasi-static tensile tests are provided for QP3Mn in Figure D.1, TRIP7Mn in Figure D.2, and QP980 in Figure D.3. Four ferrite peaks, \{110\}, \{200\}, \{211\}, \{220\}, and four austenite peaks, \{111\}, \{200\}, \{220\}, \{311\}, are labeled in each plot. The intensity of the austenite peaks decreases with increasing amount of strain. The ratio of the ferrite peak intensities for the TRIP7Mn is very different from the QP980 and QP3Mn, which indicates a significant texture difference between the steels. Repeated measurements of retained austenite in as-processed samples of QP980 fall within 1 vol.\% of each other. Uncertainty of the XRD measurement of retained austenite is estimated to be 1 vol.\% based on these repeated measurements. Austenite measurements from four as-processed QP3Mn samples resulted in a 1 vol.\% standard deviation, and austenite measurements from four as-processed TRIP7Mn samples in a 2 vol.\% standard deviation.

Figure D.1 QP3Mn X-ray peak intensities from the as-processed condition and after various amounts of strain from tensile tests at 0.0005 s\(^{-1}\). Samples were scanned over a 2-theta range of 40\(^\circ\)– 105\(^\circ\) with a 200 s dwell time and 0.05\(^\circ\) step size using Cu K\(_{\alpha}\) radiation. Austenite (A) and ferrite (F) peaks are labeled.
Figure D.2  TRIP7Mn X-ray peak intensities from the as-processed condition and after various amounts of strain from tensile tests at 0.0005 s\(^{-1}\). Samples were scanned over a 2-theta range of 40°–105° with a 200 s dwell time and 0.05° step size using Cu K\(_\alpha\) radiation. Austenite (A) and ferrite (F) peaks are labeled.

Figure D.3  QP980 X-ray peak intensities from the as-processed condition and after various amounts of strain from tensile tests at 0.0005 s\(^{-1}\). Samples were scanned over a 2-theta range of 40°–105° with a 200 s dwell time and 0.05° step size using Cu K\(_\alpha\) radiation. Austenite (A) and ferrite (F) peaks are labeled.
APPENDIX E

TRIP7MN EBSD DATA

The TRIP7Mn as-received microstructure is shown in the image quality map, inverse pole figure map, and the phase map from an EBSD scan in Figure E.1. EBSD was also performed for select conditions from interrupted tensile tests to determine whether EBSD could provide information on the deformation mechanisms of the TRIP7Mn at various strain rates and test temperatures. The image quality map, inverse pole figure map, and the phase map from an EBSD scan of a tensile test interrupted at 20% plastic strain are shown for: 22 °C and 10 s\(^{-1}\) in Figure E.2, 70 °C and 0.0005 s\(^{-1}\) in Figure E.3, and 115 °C and 0.0005 s\(^{-1}\) in Figure E.4. While the retained austenite content is lower in the EBSD scans from the interrupted tests compared to the as-processed condition, the EBSD scans do not show obvious differences in the deformation mechanisms (e.g. twinning in the austenite).

Ana Araujo is gratefully acknowledged for her assistance with sample preparation and running the EBSD scans.
Figure E.1  EBSD scan of the TRIP7Mn as-processed microstructure: (a) image quality map, (b) inverse pole figure map, and (c) phase map. (Color image - see PDF version)
Figure E.2 EBSD scan of the TRIP7Mn microstructure from a 22 °C, 10 s⁻¹ tensile test that was interrupted at 20% plastic strain: (a) image quality map, (b) inverse pole figure map, and (c) phase map. (Color image - see PDF version)

Figure E.3 EBSD scan of the TRIP7Mn microstructure from a 70 °C, 0.0005 s⁻¹ tensile test that was interrupted at 20% plastic strain: (a) image quality map, (b) inverse pole figure map, and (c) phase map. (Color image - see PDF version)
Figure E.4  EBSD scan of the TRIP7Mn microstructure from a 115 °C, 0.0005 s\(^{-1}\) tensile test that was interrupted at 20% plastic strain: (a) image quality map, (b) inverse pole figure map, and (c) phase map. (Color image - see PDF version)
APPENDIX F
SPECIMEN TEMPERATURE MODELING

The values of $\beta$ equal to 0.95, $\rho$ equal to 7800 kg/m$^3$, and $C_p$ equal to 450 J/kg·°C were used in Equation 2.7 to model the specimen temperature for both QP3Mn and TRIP7Mn in Figure 5.3. The minimum and maximum values for $\beta$, $\rho$, and $C_p$ used in literature to model the increase in specimen temperature during high strain rate tensile tests of AHSS are summarized in Table F.1 [9, 134, 139, 168–170]. Examples of how the minimum and maximum values of the constants affect the calculated change in specimen temperature at UTS are shown in Figure F.1 for QP3Mn and in Figure F.2 for TRIP7Mn. In both Figures F.1 and F.2, the model results for are shown for the following combinations of the minimum and maximum values of the material constants: (a) $\beta = 0.9$, $\rho = 7800\text{kg/m}^3$, $C_p = 445\text{J/kg·°C}$, (b) $\beta = 0.9$, $\rho = 7870\text{kg/m}^3$, $C_p = 445\text{J/kg·°C}$, (c) $\beta = 0.9$, $\rho = 7800\text{kg/m}^3$, $C_p = 500\text{J/kg·°C}$, and (d) (a) $\beta = 1$, $\rho = 7800\text{kg/m}^3$, $C_p = 445\text{J/kg·°C}$. There is a large difference between the model prediction and the experimental temperature measurement at strain rates below 0.1 s$^{-1}$, because there is time for thermal diffusion and heat dissipation during tensile tests at the low strain rates.

Table F.1 – Minimum and maximum values of material constants for modeling the change in specimen temperature due to heat generation from plastic deformation.

<table>
<thead>
<tr>
<th>Constant</th>
<th>Minimum</th>
<th>Maximum</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\beta$</td>
<td>0.9</td>
<td>1</td>
</tr>
<tr>
<td>$\rho$ (kg/m$^3$)</td>
<td>7800</td>
<td>7870</td>
</tr>
<tr>
<td>$C_p$ (J/kg·°C)</td>
<td>445</td>
<td>500</td>
</tr>
</tbody>
</table>

Figures F.1a and F.1b for QP3Mn and Figures F.2a and F.2b for TRIP7Mn show that a small change in $\rho$ does not have a significant effect on the predicted $\Delta T$. A low value of $C_p$ results in better agreement between the model and the experimental measurements as shown by Figures F.1a and F.1c for QP3Mn and Figures F.2a and F.2c for TRIP7Mn. A $\beta$ equal to 1 results in better agreement between the model and the experimental measurements than $\beta$ equal to 0.9 when $\rho$ and $C_p$ are the minimum values as shown by Figures F.1a and F.1d for QP3Mn and Figures F.2a and F.2d for TRIP7Mn. A $\beta$ equal to 1 means that all of the work to plastically deform the specimen is converted to heat and none remains stored in the dislocation structure. A $\beta$ equal to 0.95 was used for the modeling presented in Figure 5.3,
since a $\beta$ equal to 1 is theoretically unreasonable for a plastically deformed metal and the latent heat of the austenite to martensite transformation is expected to be very small (not enough extra heating to increase $\beta$ all the way to 1).
Figure F.1  Experiment measurements and model predictions for the change in temperature at UTS for QP3Mn. The model results for different combinations of the minimum and maximum values of the material constants are shown for (a) $\beta = 0.9$, $\rho = 7800 \text{kg/m}^3$, $C_p = 445 \text{J/kg} \cdot ^\circ \text{C}$; (b) $\beta = 0.9$, $\rho = 7870 \text{kg/m}^3$, $C_p = 445 \text{J/kg} \cdot ^\circ \text{C}$; (c) $\beta = 0.9$, $\rho = 7800 \text{kg/m}^3$, $C_p = 500 \text{J/kg} \cdot ^\circ \text{C}$, and (d) (a) $\beta = 1$, $\rho = 7800 \text{kg/m}^3$, $C_p = 445 \text{J/kg} \cdot ^\circ \text{C}$. At low strain rates, the model prediction (black circle) and the experimental temperature measurement (filled grey circle) are from the same tensile test. At higher strain rates, the average model prediction from three different tensile tests (triangle) is plotted with error bars representing one standard deviation. Temperature measurements were made with a K-type thermocouple.
Figure F.2  Experiment measurements and model predictions for the change in temperature at UTS for TRIP7Mn. The model results for different combinations of the minimum and maximum values of the material constants are shown for (a) $\beta = 0.9$, $\rho = 7800\, \text{kg/m}^3$, $C_p = 445\, \text{J/kg}\cdot{\degree}\text{C}$, (b) $\beta = 0.9$, $\rho = 7870\, \text{kg/m}^3$, $C_p = 445\, \text{J/kg}\cdot{\degree}\text{C}$, (c) $\beta = 0.9$, $\rho = 7800\, \text{kg/m}^3$, $C_p = 500\, \text{J/kg}\cdot{\degree}\text{C}$, and (d) $\beta = 1$, $\rho = 7800\, \text{kg/m}^3$, $C_p = 445\, \text{J/kg}\cdot{\degree}\text{C}$. At low strain rates, the model prediction (black circle) and the experimental temperature measurement (filled grey circle) are from the same tensile test. At higher strain rates, the average model prediction from three different tensile tests (triangle) is plotted with error bars representing one standard deviation. Temperature measurements were made with a K-type thermocouple.
A modification to the Olson-Cohen model is proposed that accounts for the temperature dependence of the model parameters based on specimen heating during high strain rate tensile tests. To account for the effect of adiabatic heating on the austenite transformation kinetics, the change in the specimen temperature as a function of strain must be known and the rate of change of the kinetics from the Olson-Cohen model as a function of strain and specimen temperature must be calculated. The derivative of Equation 2.4 expresses the rate of the austenite transformation and is given by Equation G.1. The temperature dependence of the specimen temperature on strain is designated as the function $T_\varepsilon$, and the $\alpha$ and $\beta$ parameters are a function of $T_\varepsilon$. The $\alpha$ and $\beta$ in Equation G.1 are replaced by $\alpha(T_\varepsilon)$ and $\beta(T_\varepsilon)$ to give an expression for the rate of change of the transformation kinetics as a function of specimen temperature, which is dependent on strain. The calculation of $f_{\alpha'}$ for a tensile specimen that changes temperature as a function of strain requires the integration given by Equation G.2. During high strain rate tensile tests, the specimen temperature changes a function of strain, which is represented by $T_\varepsilon$. In Equation G.2, the $\alpha$ and $\beta$ parameters are a function of $T_\varepsilon$. The derivative of Equation 2.4 and the integration of Equation G.2 for QP3Mn and TRIP7Mn for 0.0005 and 0.1 s$^{-1}$ test conditions were performed using MATLAB® software.

\[
\frac{df_{\alpha'}}{d\varepsilon} = \alpha \cdot \beta \cdot n \cdot \exp\left(-\beta \left(1 - \exp\left(-\alpha \cdot \varepsilon\right)\right)^n\right) \cdot \exp\left(-\alpha \cdot \varepsilon\right) \cdot \left(1 - \exp\left(-\alpha \cdot \varepsilon\right)\right)^{n-1} \tag{G.1}
\]

\[
f_{\alpha'} = \int_{0}^{\varepsilon_U} (\alpha(T_\varepsilon)) \cdot (\beta(T_\varepsilon)) \cdot n \cdot \exp\left(-\left(\beta(T_\varepsilon)\right) \cdot \left(1 - \exp\left(-\left(\alpha(T_\varepsilon)\right) \cdot \varepsilon\right)\right)^n\right) \cdot \exp\left(-\left(\alpha(T_\varepsilon)\right) \cdot \varepsilon\right) \cdot \left(1 - \exp\left(-\left(\alpha(T_\varepsilon)\right) \cdot \varepsilon\right)\right)^{n-1} \frac{df_{\alpha'}}{d\varepsilon} \text{ where } T_\varepsilon = f(\varepsilon)|_{\varepsilon} \tag{G.2}
\]

The expressions for $\alpha(T_\varepsilon)$ and $\beta(T_\varepsilon)$ for QP3Mn and TRIP7Mn are provided in Table G.1. These functions come from the curve fits shown in Figures 5.10b and 5.11d. For the TRIP7Mn, third order polynomials were reasonable fits to the $\alpha$ and $\beta$ data. However, the expressions for $\alpha(T_\varepsilon)$ and $\beta(T_\varepsilon)$ for QP3Mn were more difficult to define due to the challenges in fitting the Olson-Cohen model to the experimental data. The $\alpha$ and $\beta$ parameters for the model fitting with $n$ equal to 1 were used, because sum of differences squared were lower for the Olson-Cohen model fits with $n$ equal to 1 compared to $n$ equal to 2. The $\alpha$ and $\beta$ parameters from the 70 and 85 °C data were neglected, since those data sets were the
most difficult to fit with the Olson-Cohen model. Based on the data in Figure 5.11d, $\alpha(T_\varepsilon)$ is assumed to be a constant equal to the average of the three $\alpha$ values, and $\beta(T_\varepsilon)$ is described by a linear trend line. The QP3Mn specimen temperature at UTS is in the approximate range of 60 – 65 °C for 0.1 – 100 s$^{-1}$ strain rates; therefore, the expressions for $\alpha$ and $\beta$ as a function of temperature based on the 22 – 55 °C tests are representative of most of the specimen temperature range during the high strain rate tensile tests.

Table G.1 – Expressions for the Olson-Cohen $\alpha$ and $\beta$ parameters as a function of temperature based on fits to experimental data. The data for TRIP7Mn came from tensile tests at 0.0005 s$^{-1}$ in the temperature range 22 – 115 °C. The data for QP3Mn came from tensile tests at 0.0005 s$^{-1}$ in the temperature range 22 – 55 °C.

<table>
<thead>
<tr>
<th>Steel</th>
<th>Parameter Function</th>
<th>$R^2$</th>
</tr>
</thead>
<tbody>
<tr>
<td>TRIP7Mn</td>
<td>$\alpha(T_\varepsilon) = \left( -1.9 \times 10^{-9} \right) \cdot T_\varepsilon^3 + 0.00085 \cdot T_\varepsilon^2 - 0.17 \cdot T_\varepsilon + 13$</td>
<td>0.99</td>
</tr>
<tr>
<td></td>
<td>$\beta(T_\varepsilon) = \left( 8.2 \times 10^{-7} \right) \cdot T_\varepsilon^3 - 0.00035 \cdot T_\varepsilon^2 + 0.017 \cdot T_\varepsilon + 2.0$</td>
<td>0.96</td>
</tr>
<tr>
<td>QP3Mn</td>
<td>$\alpha(T_\varepsilon) = 18$</td>
<td>–</td>
</tr>
<tr>
<td>QP3Mn</td>
<td>$\beta(T_\varepsilon) = -0.017 \cdot T_\varepsilon + 1.9$</td>
<td>0.73</td>
</tr>
</tbody>
</table>

Since the austenite to martensite transformation rate observed experimentally is similar at 0.1, 10, and 100 s$^{-1}$ for both the QP3Mn and TRIP7Mn, the 0.1 s$^{-1}$ condition will be used as a representative example for modeling. The modeling was also performed for 0.0005 s$^{-1}$ to check that the model could make accurate predictions for a condition with little or no self heating. Curve fits were applied to the specimen temperature measurements as a function of plastic true strain at 0.0005 s$^{-1}$ and 0.1 s$^{-1}$. For the TRIP7Mn, the fit was assumed to be linear at 0.1 s$^{-1}$. At 0.0005 s$^{-1}$ for the TRIP7Mn, the specimen temperature oscillates within ± 0.5 °C of 27 °C due to repeated cycles of heating and cooling due to passage of PLC bands along the gage length. Therefore, the temperature is assumed to be a constant 27 °C for the model at 0.0005 s$^{-1}$. For the QP3Mn, the temperature was measured over a much smaller range of strain, and second-order polynomial fits were more reasonable than linear fits. The expressions for $T_\varepsilon$ are summarized in Table G.2.

Table G.2 – Expressions for the specimen temperature as a function of strain, $T_\varepsilon$, from tensile tests at 0.0005 and 0.1 s$^{-1}$ based on fits to experimental data.

<table>
<thead>
<tr>
<th>Steel</th>
<th>Strain Rate (s$^{-1}$)</th>
<th>$T_\varepsilon$ (°C)</th>
<th>$R^2$</th>
</tr>
</thead>
<tbody>
<tr>
<td>TRIP7Mn</td>
<td>0.0005</td>
<td>$T_\varepsilon = 27$</td>
<td>–</td>
</tr>
<tr>
<td>TRIP7Mn</td>
<td>0.1</td>
<td>$T_\varepsilon = 271 \cdot \varepsilon + 24$</td>
<td>0.98</td>
</tr>
<tr>
<td>QP3Mn</td>
<td>0.0005</td>
<td>$T_\varepsilon = -326 \cdot \varepsilon^2 + 69 \cdot \varepsilon + 25$</td>
<td>0.99</td>
</tr>
<tr>
<td>QP3Mn</td>
<td>0.1</td>
<td>$T_\varepsilon = 2106 \cdot \varepsilon^2 + 273 \cdot \varepsilon + 29$</td>
<td>0.99</td>
</tr>
</tbody>
</table>
The predicted martensite fraction as a function of strain from the modified Olson-Cohen model in Equation G.2 and the experimental results at 0.0005 and 0.1 s\(^{-1}\) for TRIP7Mn and QP3Mn are shown in Figure G.1. Figures G.1b and G.1d show the calculated values for the specimen temperature as a function of strain from the equations in Table G.2 and the calculated values for the \(\alpha\) and \(\beta\) parameters as a function of strain from the equations in Table G.1. For the TRIP7Mn, the specimen self-heating at 0.1 s\(^{-1}\) leads to large decreases in the predicted \(\alpha\) and \(\beta\) with increasing strain. In Figure G.1a, the model at 0.0005 s\(^{-1}\) is close to the experimental data. For 0.1 s\(^{-1}\), the model and the experiment data agree up to approximately 14% strain. Above 14% strain, the model predicts a greater transformation rate than the transformation rate shown by the experiment data. The model prediction for 0.1 s\(^{-1}\) is unreasonable above 20% strain, because it predicts that more austenite transformation will occur at 0.1 s\(^{-1}\) than at 0.0005 s\(^{-1}\).

For the QP3Mn, the modified Olson-Cohen model prediction for the fraction of transformed martensite as a function of strain at 0.0005 and 0.1 s\(^{-1}\) is very close to the experimental values, as shown in Figure G.1c. Based on the quasi-static elevated temperature tests, the change in the specimen temperature during a high strain rate test is not expected to be high enough to affect \(\alpha\) but \(\beta\) is expected to decrease with increasing strain, as shown in Figure G.1d. The slight decrease in \(\beta\) with increasing strain at 0.1 s\(^{-1}\) is responsible for slightly reducing the transformation kinetics in the model at high strains and causing the small difference between the 0.0005 and 0.1 s\(^{-1}\) curves at high strains.
Figure G.1 The experimental data (points) for the fraction of transformed martensite as a function of strain and the modified Olson-Cohen model prediction (curve) with the effect of specimen temperature on α and β at 0.0005 s\(^{-1}\) and 0.1 s\(^{-1}\) for (a) TRIP7Mn and (c) QP3Mn. The specimen temperature, α, and β as a function of strain are shown for 0.0005 s\(^{-1}\) and 0.1 s\(^{-1}\) for (b) TRIP7Mn and (d) QP3Mn.