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The focus of the research project was to systematically assess the strain rate dependence of strengthening mechanisms in new advanced high strength sheet steels. Data were obtained on specially designed and produced Dual Phase and TRIP steels and compared to properties of automotive sheet steels currently in use.

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Report Documentation Page

Title and Subtitle: AISI/DOE Technology Roadmap Program
Constitutive Behavior of High Strength Multiphase
Sheet Steels Under High Strain Rate Deformation
TRP 9904

Author(s): David Matlock, Co-Principal Investigator
John Speer, Co-Principal Investigator

Performing Organization Advanced Steel Processing and Products Research Center
Department of Metallurgical and Materials Engineering
Colorado School of Mines
Golden, CO 80401

Abstract:

The focus of this research program was to systematically assess the strain rate dependence of strengthening mechanisms (e.g. ferrite grain size, cold work, solid solution strengthening, low-temperature aging, martensite properties and volume fraction, and amount and stability of retained austenite) in new advanced high-strength sheet steels. Data were obtained on specially designed and produced Dual-Phase and TRIP steels and compared to properties of current automotive sheet steels (e.g. IF, HSLA, AKDQ, etc.). This project was initiated in December 1999 with the following four primary objectives:

- Establish a test capability for measuring mechanical properties at elevated strain rates
- Quantify the effectiveness of different metallurgical strengthening mechanisms in the automotive high strain rate regime
- Design an alloying/processing approach for producing an optimized multiphase sheet steel with controlled microstructural variations, and assess high strain-rate properties.
- Provide data on constitutive material behavior for use in improved crash simulation models.

The identified project deliverable(s) were:

- Experimental data on the effects of strain rate on the strengthening efficiency of the various strengthening mechanisms applicable to advanced high strength multi-phase sheet steels.
- State-of-the-art review of the metallurgy of advanced high strength multi-phase sheet steels.
- Test methodologies for the evaluation of sheet steels at high strain rates.
- Assessment of constitutive equations for multi-phase sheet steels at high strain rates.

Final Report

AISI/DOE Project 9904

“Constitutive Behavior of High Strength Multiphase Sheet Steels Under High Strain Rate Deformation Conditions”

Submitted to

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March 31, 2005

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Executive Summary

A laboratory facility was established for tensile testing sheet steels up to displacement rates of approximately 10 m/s and correspondingly at strain rates up to 500 s^{-1} . The system was used to evaluate a series of Dual-Phase and TRIP steels processed to produce materials with controlled volume fractions and properties of the second phase. The microstructures were developed based on the results of a comprehensive literature survey entitled “The Mechanical and Physical Metallurgy of Low-Carbon Sheet Steels Produced by Dual-Phase and TRIP Steel Processing.” Process methodologies were developed and the resulting austenite carbon contents and volume fractions correlated to desired values and to predictions based on Thermo-Calc. Data were also obtained on IF steels with different ferrite grain sizes and solid solution strengthening additions, HSLA steels, and one steel processed with a fully martensitic microstructure. Analysis of the interrelationships between imposed strain rate during tensile testing and microstructure were developed. The results showed that the strain rate sensitivity of the advanced multiphase steels was primarily controlled by the strain rate dependence of the ferrite and the strength could be expressed in terms of a rule of mixtures calculation.

1.0 Introduction

Low-carbon automotive sheet steels have been traditionally processed to produce microstructures that consist primarily of ferrite and, depending on carbon content, small amounts of pearlite in hot rolled strip or fine dispersions of spheroidized carbide particles in cold-rolled and annealed sheet. The ductilities and formabilities of such steels are high but the strengths are low. Higher strength HSLA steels that incorporate microalloying and process modifications are also produced. The requirements of weight reduction and greater safety in automobiles have driven the development of higher strength automotive sheet steels (e.g. Dual-Phase and TRIP steels) with good combinations of strength and ductility.

Multiphase sheet steels including dual-phase and TRIP steels have generated increasing interest in the past few years. TRIP steels typically involve ferrite/bainite/austenite microstructures containing significant levels of retained austenite, which transform to martensite during deformation (via “Transformation Induced Plasticity”). While improved formability associated with the TRIP mechanism has been reported widely, the on-vehicle application is also driven by weight reduction potential resulting from improvements in crash-energy management associated with unique high strain rate deformation behavior of these steels. To date, most mechanical property studies on sheet steels have concentrated on results obtained under quasi-static (i.e. low strain rate) deformation conditions. To develop safer vehicles, a better understanding is required of high strain rate properties, and the associated physical metallurgical principles and microstructure/property relationships, in new advanced high strength sheet steels.

In response to the needs for improved understanding of the deformation behavior of advanced high strength sheet steels (AHSS), this program was initiated by the Advanced Steel Processing and Products Research Center (ASPPRC) at the Colorado School of Mines (CSM) in response to the American Iron and Steel Institute Research Solicitation in cooperation with the US Department of Energy. The program was responsive to several priority research needs of the industry defined by the Product Development chapter of the *Steel Industry Technology Roadmap* and by the specific projects identified for attention in the 1999 proposal solicitation, most notably the need for expansion of constitutive equations to predict the behavior of automotive steels under high impact loading conditions.

The focus of this research program was to systematically assess the strain rate dependence of strengthening mechanisms (e.g. ferrite grain size, cold work, solid solution strengthening, low-temperature aging, martensite properties and volume fraction, and amount and stability of retained austenite) in new advanced high-strength sheet steels. Data were obtained on specially designed and produced Dual-Phase and TRIP steels and compared to properties of current automotive sheet steels (e.g. IF, HSLA, AKDQ, etc.). This project was initiated in December 1999 with the following four primary objectives:

- Establish a test capability for measuring mechanical properties at elevated strain rates
- Quantify the effectiveness of different metallurgical strengthening mechanisms in the automotive high strain rate regime
- Design an alloying/processing approach for producing an optimized multiphase sheet steel with controlled microstructural variations, and assess high strain-rate properties.
- Provide data on constitutive material behavior for use in improved crash simulation models.

The identified project deliverable(s) were:

- Experimental data on the effects of strain rate on the strengthening efficiency of the various strengthening mechanisms applicable to advanced high strength multi-phase sheet steels.
- State-of-the-art review of the metallurgy of advanced high strength multi-phase sheet steels.
- Test methodologies for the evaluation of sheet steels at high strain rates.
- Assessment of constitutive equations for multi-phase sheet steels at high strain rates.

A research team of faculty and students was assembled to work on the project. The team members who participated on various aspects of the project are:

Principal Investigators:

Dr. John G. Speer, CSM (jspeer@mines.edu)
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Ryan S. Kircher, M.S. Candidate – Thesis to be completed April 2005

Undergraduate Research Assistants:

Matthew Ruggiero
Michael P. Morimoto
Ty Colman
Collin Donohoue

Several corporations were partners in this program, and due to restructuring of the industry, several changes occurred in the participant list during the project. The participating companies were:

AK Steel (first year only)
Bethlehem Steel Corporation (now International Steel Group)
Ispat Inland Inc.
National Steel Corporation (now part of US Steel)
Rouge Steel Company (now Severstal N.A.)
US Steel Company
MTS Systems Corporation
DaimlerChrysler Corporation

A comprehensive literature survey entitled “The Mechanical and Physical Metallurgy of Low-Carbon Sheet Steels Produced by Dual-Phase and TRIP Steel Processing” on the properties and processing of DP and TRIP steels was completed and distributed to participants on April 16, 2002. A copy of that report is presented here as Appendix A. Specific material parameters were identified for further investigation and four experimental steel compositions were selected. US Steel produced these steels for CSM. Process methodologies, summarized in Section 2.0 of this report, were developed and the resulting austenite carbon contents and volume fractions correlated to desired values and to predictions based on Thermo-Calc. Tensile data on samples with controlled microstructural variations were obtained following the procedures

outlined in Section 3.0 and summarized in more detail in the thesis of Denise Bruce presented in Appendix B. The specific experimental microstructures are summarized in Section 4.0 and analyses of the interrelationships between imposed strain rate during tensile testing and microstructure were developed, along with an assessment of the effects of strain rate on bake hardening response. These results, summarized in Section 5.0, are detailed in the theses of Bruce [Appendix B] and Ryan Kircher (A draft version is presented in Appendix C; a final version will be distributed to participants on completion, anticipated for April 2005) and in the copies of the technical papers presented in Appendices D to K. Correlations between experimentally measured strain rate dependent mechanical property data and constitutive equation formulations are presented. In addition to the information summarized here, CSM also participated in a round-robin testing program on high strain rate testing of sheet steels organized by the International Iron and Steel Institute, and a separate report will be issued by that organization.

2.0 Microstructural Design Concepts and Material Processing

Families of Advanced High Strength Steels (AHSS) are being developed and are being used more extensively for automotive applications. Several classes of these steels are considered in the present study to evaluate the effects of strain rate on mechanical properties. Specific material classes of interest include:

- (a) Solid-solution strengthened Interstitial-Free Steel (IF)
- (b) Bake Hardenable Ultra Low Carbon Steel (BH)
- (c) High Strength Low Alloy Steels (HSLA)
- (d) Dual Phase Steel (DP)
- (e) Transformation-Induced Plasticity Steel (TRIP)

Interestingly, the major microstructural constituent of these steels is primarily ferrite and the second phase particles are engineered through alloy design and processing to induce extra strengthening in the ferrite either through grain refinement, dislocation hardening, precipitate strengthening or by solid solution strengthening. An analysis of the microstructures of the AHSS steels indicated above is summarized as follows:

IF steel	Ferrite + Solid Solution Strengtheners,
BH Steel	Ferrite + Solid Solution Strengtheners + Interstitial C,
HSLA Steel	Ferrite + Grain Refinement + second phase particles,
Dual Phase	Ferrite + Martensite
TRIP Steel	Ferrite + Bainite + Austenite (10-15%) + Martensite

Since ferrite is the major microstructural constituent in all of these steels, it is expected that during deformation, most of the plastic strain will be accommodated by the ferrite phase itself. Strain rate dependence of the flow stress of these materials will therefore be largely dictated by the ferrite properties. The second phase particles may

exert significant influence on the ferrite properties and therefore modify the strain rate sensitivity of ferrite and hence the microstructure in totality. It was therefore, decided that the strain rate sensitivity of all the above-mentioned microstructures will be evaluated to establish a correlation between microstructures and to be able to meaningfully interpret high strain rate deformation behaviors.

Dual-Phase (DP) steels are referred to the steels with microstructures consisting largely of ferrite and martensite. TRIP steels are the latest addition to the family of high strength steels with dispersed phases in a matrix of ferrite initiated by Matsumura *et al.* in 1987 [1,2]. The approach consisted of intercritical annealing and subsequent isothermal transformation to produce bainite and retained austenite dispersed in a matrix of ferrite. Retention of a significant amount of retained austenite (~10%) is a major microstructural design objective in TRIP steels. The ability of retained austenite to transform to martensite during deformation has led to the use of the term TRIP steels to describe intercritically annealed/isothermally transformed high strength steel. High strength, but with greater ductility than produced in dual phase steels, was obtained by the isothermal transformation processing.

The present study was aimed at designing microstructural variations in these high strength steels and assessing the quasi-static and high strain rate deformation behavior influenced by such microstructural variations. The elements of microstructural control (through processing and chemistry design) in DP and TRIP steels along with a consideration of microstructural effects on mechanical properties, with emphasis on imposed high strain rate deformation, has been discussed in detail in the literature review in Appendix A. Based on the literature review, the following microstructural variations in DP and TRIP steels were identified for examining the influence on high strain rate deformation on the stress-strain behavior.

2.1 Microstructural Design

2.1.1 Dual-phase microstructure

Dual-phase microstructures with a fine grained ferrite matrix and a uniform distribution of fine martensite islands was identified as the desirable material for this study. The following variations in the microstructure were sought in the DP materials for the present study:

- i. Vary the volume fraction of martensite in three different DP steels without changing the martensite C-content.
- ii. Vary the martensite C-content in two different DP steels without changing the martensite volume fraction.

2.1.2 TRIP microstructure

The following variations in the microstructures consisting of ferrite, bainite and retained austenite were identified for assessment:

- i. Vary the volume fraction of retained austenite in two steels without changing the austenite C-contents.
- ii. Vary the austenite C-content without changing the austenite volume fraction.

2.2 Chemical Compositions of Experimental Steels

It was desired that the same steels be used to produce both DP and TRIP microstructures so that the influence of alloying elements on the starting austenite and ferrite constituents remained common to both families of steels. A Si-bearing and an Al-bearing steel composition were proposed to compare the influence of Al and Si on mechanical properties. Replacement of Si by Al is of specific interest to the TRIP steels, to examine the influence on stability of the retained austenite. The partial elimination of Si could result in a reduced volume fraction of austenite which, in turn, may affect the tensile stress and the uniform elongation [3,4]. Therefore, an increased total initial C-content for the Si-Al bearing steel was selected to expand the γ -region [5] and provide the same amount of retained austenite in the final microstructure in the Si-bearing and Al-bearing steels for comparison.

A thermodynamic model was used to guide alloy design and processing. Four experimental steels were made at the US Steel Research Laboratory in Monroeville, PA in the third week of November 2002 and the compositions are given in Table 2.1. Steels A, B and C are Si-bearing steels and vary only in their total carbon content. These steels are designated in the report as Low-C-Si, Med-C-Si and High-C-Si according to the level of total C-contents. Steel D is Al-bearing steel and is designated as High-C-Al steel. The hot rolled sheets with an initial sheet thickness of 2.5 mm were cold rolled to approximately 1.0 mm in about 14 passes.

Table 2.1: Chemical compositions of experimental steels (in wt. pct.)

Steels	Identification	C	Mn	Si	Al	P	S
A	Low-C-Si	0.10	1.52	1.48	0.032	0.009	0.0038
B	Med-C-Si	0.14	1.48	1.46	0.032	0.009	0.0040
C	High-C-Si	0.21	1.46	1.47	0.031	0.009	0.0041
D	High-C-Al	0.21	1.49	0.34	0.99	0.010	0.0050

2.3 Alloy Processing Design to Produce Experimental DP and TRIP Microstructures

2.3.1 DP Microstructure Development

All experimental microstructures were obtained by laboratory heat treating. Coupons from Low-C-Si, Med-C-Si and High-C-Si steels were annealed at 775, 800 and 825 °C for 5 min in salt bath furnaces and then water quenched to room temperature. Microstructures of these steels were characterized with respect to volume fraction of martensite, C-content in martensite, retained austenite content and ferrite grain size. Martensite volume fractions were obtained using both image analysis and point counting methods on micrographs taken at a magnification of 500X. The martensite C-content (C_{mar}) was determined using the following expression:

$$C_{total} \text{ (wt.\%)} = V_{\alpha}C_{\alpha} + V_{\gamma}C_{\gamma} + V_{mar}C_{mar} \quad [2.1]$$

Retained austenite fraction (V_{γ}) and austenite C-content (C_{γ}) were determined from XRD scans using direct comparison method. Ferrite C-content (C_{α}) was taken as the equilibrium C-content of ferrite at the annealing temperature indicated by Thermo-Calc analysis. Grain boundary segregation of carbon during quenching was not taken into consideration as the samples were water quenched from the annealing temperature and the grain boundary coverage at the annealing temperature is known to be negligible [6].

Figure 2.1: Evolution of martensite fraction and its C-content in steels A (Low-C-Si), B (Med-C-Si) and C (High-C-Si) resulting from different annealing treatments. The trend lines showing martensite fraction as a function of annealing treatment were derived by Thermo-Calc analysis. The alpha-numeric figures alongside the symbols indicate the martensite C-contents as calculated using Eq. (2.1).

Figure 2.1 shows the martensite volume fraction obtained in each heat treated sample along with those predicted by Thermo-Calc for different annealing temperatures used. The martensite C-content is also indicated in the figure (as C'XX') for each heat treatment. Results reveal an excellent correlation between the microstructure that obtained after each heat treatment and that predicted by thermo-calc calculations. An assessment of the microstructures obtained in the heat treated samples with respect to martensite volume fraction and its C-content, allows selection of processing parameters required to develop the specified DP microstructures. From the data in Figure 2.1, it is estimated that annealing of High-C-Si, Med-C-Si and Low-C-Si steels (Table 2.1) at 775 °C followed by water quenching will result in three different dual phase microstructures with different martensite volume fractions but with similar martensite C-contents. Again, annealing of High-C-Si at 775 °C and Low-C-Si steel at 825 °C followed by water quenching will result in two different DP microstructures with similar martensite volume fractions but with different martensite C-contents. Table 2.2 shows the heat treatment applied to produce the dual-phase microstructures based on the experimental and theoretical microstructure modeling.

Table 2.2: Optimized heat treatment schedule for producing aimed dual-phase microstructures.

Steel	Heat Treatment	Aimed Microstructure
Low-C-Si	775 °C – 5 min WQ	Similar martensite C-content but different volume fractions.
Med-C-Si		
High-C-Si		
Low-C-Si	825 °C – 5 min	Similar martensite volume fractions but different martensite C-content.
High-C-Si	775 °C – 5 min	

2.3.2 TRIP Microstructure Development

Consideration of the binary Fe-C equilibrium phase diagram reveals that a simple way to vary the austenite phase fraction (without changing its carbon content) is to vary the total initial carbon concentration in the steel [Appendix A]. However, the presence of alloying elements Mn, Si and Al may influence the equilibrium austenite fraction and composition since the binary Fe-C diagram does not strictly apply to multi-component alloys. Consequently, equilibrium conditions were computed using Thermo-Calc software to indicate austenite fraction and respective austenite C-contents as a function of different annealing temperatures. The final retained austenite content and austenite C-content in the microstructure is dependent on the isothermal transformation characteristics of the steel at intermediate transformation temperatures subsequent to intercritical annealing. The transformation temperature was selected to be in the range where austenite transforms to bainitic ferrite and carbon-enriched austenite (assuming that Fe₃C precipitation is suitably inhibited by Si and/or Al). A high transformation temperature accelerates the austenite transformation to bainite and a low temperature makes the transformation too sluggish. Thus, in view of varying transformation rates, an isothermal transformation temperature of 400-425 °C was selected to ensure control

of austenite transformation kinetics within practical time limits. Representative samples from steels mentioned in Table 2.1 were heat treated at intercritical temperatures of 775 – 825 °C for times ranging from 20 s to 500 s to establish austenite transformation behavior and accordingly choose processing parameters for developing aimed TRIP microstructures.

Figure 2.2 shows typical austenite transformation behavior of the steels for an annealing temperature of 775 °C and isothermal holding temperature of 400 °C. It is revealed that the Low-C-Si and High-C-Si steels when annealed at 775 °C and isothermally held at 400 °C for about 200 s will produce two different TRIP microstructures with different austenite fractions but with similar retained austenite C-contents.

Figure 2.2: Isothermal transformation behavior of austenite with time at 400 °C for the steels in Table 2.1. The steel samples were intercritically annealed at 775 °C for 300 s prior to holding isothermally at 400 °C.

Similarly, Low-C-Si samples were given two different heat treatments as described in Table 2.3 to obtain two different TRIP microstructures with similar austenite volume fractions but with different austenite C-contents. Figure 2.3 shows the austenite transformation behavior of the steels for an annealing temperature of 780 °C and isothermal holding temperature of 420 °C. Comparing the data of Figures 2.2 and 2.3, it is revealed that the Al-bearing steel has more carbon in retained austenite at all isothermal holding temperatures. The bainite reaction seems to be sluggish in Al-

bearing TRIP steel due to the initial higher retained austenite C-content. A detailed explanation to this effect is given in Appendix I.

A comparison of Figures 2.2 and 2.3 reveals that a possible way for development of two TRIP microstructures with similar austenite volume fractions but with different austenite carbon-contents will be as indicated in Figure 2.4 and is summarized as follows:

- intercritical annealing of Med-C-Si at 775 °C for 300 s and then isothermal holding at 400 °C for 200 s;
- intercritical annealing of High-C-Al steel at 780 °C and isothermally holding at 420 °C for 200 s.

Table 2.3 summarizes the heat treatment followed for various TRIP steels used to develop the TRIP microstructures for high rate deformation studies.

Figure 2.3: Isothermal transformation behavior of austenite with time at 420 °C for the steels described in Table 1. The steel samples were intercritically annealed at 780 °C for 300 s prior to holding at 420 °C.

Figure 2.4: Isothermal transformation behavior of austenite in Med-C-Si steel and High-C-Al steels. The Med-C-Si steel was intercritically annealed at 775 °C and isothermally held at 400 °C for different length of time as indicated. The High-C-Al steel was intercritically annealed at 780 °C and isothermally held at 420 °C for different length of time.

Table 2.3 Processing Parameters and Steel Designations for TRIP Steels.

Steels	Designation	Processing	Aimed Microstructure	Remarks
Low-C-Si	Low Austenite-LC (Low Austenite Low Carbon)	775 °C 300s 400 °C 200s	Low Austenite with Low C-content	Effect of Austenite Stability
Low-C-Si	Low Austenite-HC (Low Austenite High Carbon)	780 °C 300s 420 °C 300s	Low Austenite with High Austenite C-content	
Med-C-Si	Med-Austenite-LC (Medium Austenite Low Carbon)	775 °C 300s 400 °C 200s	Medium Austenite with Low austenite C-content	Effect of Austenite Volume Fraction
High-C-Si	High-Austenite-LC (High Austenite Low Carbon)	775 °C 300s 400 °C 200s	High Austenite with Low austenite C-content	
High-C-Al	Med-Austenite-HC (Medium Austenite High Carbon)	780 °C 300s 420 °C 200s	Med Austenite with High austenite C-content	Effect of Al

2.4 Selection of Other Steel Microstructures for High Strain Rate Testing

In addition to the high strength steel microstructures developed as mentioned above, the present study was also extended to assess high strain rate deformation behavior of other important steel microstructures such as interstitial-free steel, ultra-low carbon bake hardening steel, dual-phase steel (DP590 grade), fully martensitic steel and high-strength-low-alloy (HSLA) steels. Studies of high strain rate deformation in these materials not only provided valuable material property data correlating respective microstructures, but most of the results could also be significantly used in the interpretation of the high strain rate data of the experimental DP and TRIP microstructures because of excellent interrelationships between the microstructures. Table 2.4 lists the summary of chemical compositions of the steels evaluated by Bruce [Appendix B] in her assessment of the high strain rate deformation behavior of IF steels, HSLA steels, DP steels and TRIP steels. These steels were also considered in several of the publications in Appendices D to K. Table 2.5 lists the chemical compositions of steels that were processed and tested along with experimental DP and TRIP steels during the study by Kircher [Appendix C]. The DP 590 steel mentioned in Table 2.5 was part of the materials tested during a Round Robin program on high strain rate testing of high strength steel sheets coordinated by IISI. The martensitic steel was a plain C-Mn steel heat treated in the laboratory to develop a fully martensitic microstructure.

Table 2.4 Material designations and chemical composition (wt. pct.) for each steel used in the study by Bruce [Appendix B].

Steels	C	Mn	P	Si	Ti	Nb	Other
IF-1	0.0038	0.17	0.003	0.002	0.072	NR	0.0074S-0.06Al-0.0042N
IF-2	0.0024	0.10	0.006	0.01	0.024	0.059	0.013S-0.05Cr-0.026Al-<0.003V
IF-3	0.003	0.2	0.01	NR	0.05	NR	0.04Al
IF-4	0.004	0.9	0.09	NR	0.05	NR	0.04Al
IF-5	0.008	1.42	0.009	0.220	0.128	NR	0.006S-0.016Al-0.0030N
ULC-C	0.0056	0.520	0.032	0.221	0.002	0.012	0.008S-0.010Cr-0.081Al-0.0034N
ULC-D	0.0054	0.550	0.034	0.254	0.003	0.014	0.006S-0.050Cu-0.020Ni-0.030Cr-0.077Al-0.0032N
HSLA-1	0.050	0.390	0.011	0.016	0.006	0.029	0.008S-0.029Cr-0.039Al-0.001V
HSLA-2	0.09	1.46	0.020	0.08	NR	0.045	0.001S
HSLA-3	0.08	1.5	0.001	0.5	0.04	0.0017	0.04Al
DP	0.088	1.00	0.010	0.31	NR	NR	0.006S-0.020Cr-0.53Al
TRIP-1	0.18	1.5	0.009	1.9	NR	0.005	0.003S
TRIP-3	0.10	1.52	0.0016	1.48	NR	NR	0.51Cu-0.0036S-0.046Al
11000-H00 copper	--	--	--	--	--	--	99.99% pure copper

NR: Not reported

ULC: Ultra-low carbon steel

DP: Dual phase steel

IF: Interstitial free steel

HSLA: High strength low alloy steel

TRIP: Transformation induced plasticity steel.

Table 2.5: Chemical compositions of the materials (wt. %) used for high strain rate tensile testing in Kircher's thesis [Appendix C].

Steels	C	Mn	Si	Al	P	S	Others
IF	0.009	0.24		0.050	0.064		Nb- 0.025
BH	0.008	0.32		0.022	0.078		Nb – 0.005
DP590	0.15	1.95	0.45	0.037	0.013	0.007	
Martensitic	0.10	1.50					

3.0 Experimental Methodologies for High Strain Rate Testing of Sheet Steel Samples

3.1 Heat Treatment

Two salt baths, one medium temperature bath (510 – 900 °C) for intercritical annealing and one low temperature bath (150 – 590 °C) for isothermal bainitic transformation, were used for heat treatment to produce DP and TRIP microstructures. Small sheet samples (30 mm x 20 mm) were used during heat treatment optimization process. Once the specific heat treatment was identified to produce aimed DP and TRIP microstructures, actual tensile samples prepared for high strain rate testing were used. However, the size of the salt baths and consideration for temperature uniformity limited the number of tensile samples that could be used at one time. With this system a batch of 16 tensile samples could be heat treated at one time for microstructure development. To minimize potential effects of inhomogenities introduced by heat treating different sample batches, direct comparisons of high strain rate properties were limited to samples within a single heat treat set.

3.2 Microstructure Characterization

3.2.1 Metallography

Dual phase steels consist primarily of ferrite with dispersed martensite and a very small fraction of retained austenite (1-2%). While microstructure revelation through metallography and interpretation for dual-phase steel does not pose a challenge, the TRIP steel microstructure revelation was indeed a challenge. TRIP steels consist of ferrite, bainite, retained austenite and martensite-austenite constituents. In the absence of any current report on simple metallographic etching technique for multi phase microstructure revelation, a novel two-step etching procedure was developed in the laboratory to precisely reveal all the microstructural constituent in TRIP steels [7]. A two-stage etching procedure involving 4% picral etching followed by 10% aqueous sodium meta-bisulfate tinting was used. The details of the etching procedure and its application in multiphase microstructure revelation is given in Appendix D. The step etching reveals martensite in a bright straw tint, austenite in bright white, ferrite in off-white, and carbides black. Grain sizes were measured using the linear intercept

method on micrographs taken at 500X. Martensite volume fractions in dual phase steels were measured by point counting as per ASTM E562-02 as well as image analysis on micrographs taken at 500X.

3.2.2 Retained Austenite Characterization

The retained austenite volume fractions DP and TRIP steels were measured by X-ray diffraction line profile analysis with Cu K_α radiation using the direct comparison method [8]. The peak reflections, $(110)_\alpha$, $(200)_\alpha$, $(211)_\alpha$, $(220)_\alpha$ of ferrite and $(111)_\gamma$, $(220)_\gamma$, $(311)_\gamma$, $(311)_\gamma$ of austenite, were used for quantitative measurements. A Phillips X-pert diffractometer operating at 45 kV and 40 mA was used. The integrated area and position of each peak was determined using the Philips Pro-Fit version 1.0 peak fitting software. The following relationship incorporating the integrated intensities of the $\{111\}_\gamma$, $\{200\}_\gamma$, $\{220\}_\gamma$, and $\{311\}_\gamma$ austenite peaks and the $\{110\}_\alpha$, $\{200\}_\alpha$, $\{211\}_\alpha$ and $\{220\}_\alpha$ ferrite peaks, in addition to R-values for each peak was used to quantify austenite volume fraction:

$$V_g = \frac{\frac{1}{n_g} \sum_0^n \frac{I_g^{hkl}}{R_g^{hkl}}}{\frac{1}{n_a} \sum_0^n \frac{I_a^{hkl}}{R_a^{hkl}} + \frac{1}{n_g} \sum_0^n \frac{I_g^{hkl}}{R_g^{hkl}}} \quad [3.1]$$

where V_g is volume fraction austenite, I_g^{hkl} is measured integrated intensity of an austenite peak, I_a^{hkl} is measured integrated intensity of a ferrite peak, R_g^{hkl} and R_a^{hkl} corresponds to R-value of austenite or ferrite, and n_g and n_a are the numbers of hkl lines for which integrated intensities were measured.

This incorporation of several peaks for ferrite and austenite in the calculation method was intended to minimize texture effects on the volume fraction estimation of retained austenite. R-values are a function of θ , hkl , the crystal structure and composition of each phase, and are representative of calculated theoretical intensity values. The R-values used in retained austenite determination for this study were obtained using unit cell volumes calculated from experimentally measured lattice parameters, which inherently incorporate the carbon and alloy content. Additionally, when $K_{\alpha 1}$ and $K_{\alpha 2}$ peaks were present for ferrite, the integrated intensity incorporated only the $K_{\alpha 1}$ peak. The Lorentz-polarization factor was modified to account for the use of a graphite monochromator.

Carbon concentrations in retained austenite were assessed from measurements of austenite lattice parameters using the following relation [9]:

$$a_0 (nm) = 0.3578 + 0.0033C (wt. \%) \quad [3.2]$$

In an important study on the effect of alloying addition on the lattice parameter of austenite by Dyson *et al.* [9] it was revealed that aluminum significantly distorts the

lattice of austenite. This is largely due to the differences in atomic size of iron (0.124nm) and aluminum (0.143 nm). The lattice distortion is minimal for silicon in austenite as the atomic radii of iron (0.124 nm) and silicon (0.117 nm) are similar. It is reported that the individual elements increase the lattice parameter of austenite in the sequence as Al>Mn>Cu>Co>W>Mo>V>Si. It was proposed that the lattice parameter of austenite may be expressed by the relation mentioned above taking into account the effects due to Al, Mn and Si.

3.3 Mechanical Testing

A high strain rate testing system was purchased from the MTS Systems Corporation in Eden Prairie and the hardware was installed at the Colorado School of Mines in November 2000. The system consists of a 50 kN (11,000 lb) capacity actuator mounted in a 500 kN (110,000 lb) capacity frame. Conventional speeds are controlled through a 10 gpm servo valve supplied by a 10 gpm hydraulic pump. High rates are controlled through a 400 gpm servo valve supplied by a 5 gallon capacity oil accumulator. System control is provided through a MTS TestStar digital controller with associated computer system. The specified peak velocity for the system was 13.5 m/s at zero load and 10 m/s at 50% of the peak load capacity. At CSM the system was verified to meet or exceed these performance specifications. The system is equipped with tensile grips capable of testing standard and sub-sized ASTM E-8 flat tensile samples and a slack-adaptor is installed between the lower grip and the moving actuator to accommodate the required displacement to accelerate the actuator. Data acquisition at conventional rates is with the MTS TestStar system and at high rates with a National Instruments data acquisition board. The maximum sampling rate currently achieved by the system is 500 kHz, and can measure up to four (4) signals during a test.

Tensile tests were performed over a strain rate range of 0.001 s^{-1} to 200 s^{-1} . The mechanical properties of each material were characterized at quasi-static rates using a screw-driven testing frame at a strain rate of 10^{-3} s^{-1} . All other mechanical tests were conducted utilizing the high strain rate procedure. The materials were tested at strain rates of 0.001, 0.01, 0.1, 1, 10, 100 and 200 s^{-1} . Each material was to be tested in duplicate. Due to limitations in heat treatment capacity, only 16 specimens of each processing parameter could be produced at one time.

3.3.1 Specimen Size

In an effort to further standardize the testing methodology of high strain rate tensile testing, a single sample size was developed to encompass all of the materials tested. Figure 3.1 presents a schematic sample drawing to indicate sample dimensions used. As discussed below, the grip portion of the specimen was used as a load-measuring device. The dimensions of the grip section of the specimen are important to the successful and accurate measurement of load. The grip sections of the sample were precision machined to ensure uniformity. Since the grip section of the specimen was used as a load cell, it was necessary to insure that the grip section only experienced elastic deformation during the tests. This required a large grip-to-gauge width ratio. Due to the installation of a new gripping system [Appendix A] after this

program was initiated, sample grip widths were limited to 15.2 mm. All materials were machined into tensile specimens prior to heat-treating. All test samples were cut transverse to the rolling direction.

Notes:

1. The ends of the reduced section shall not differ in width by more than 0.001 in. Also, there may be a gradual decrease in width from the ends to the center, but the width at each end shall not be more than 0.003 in. larger than the width at the center.
2. The edges of the reduced section shall be machined parallel over the gage length of the sample within a tolerance of 0.0005 in.
3. The material surface may be left in the as-received condition.
4. The radii of the fillets shall be equal to each other within a tolerance of 0.05 in. and the centers of curvature of the two fillets at a particular end shall be located across from each other on a line perpendicular to the centerline within a tolerance of 0.10 in.
5. Grip section edges shall be machined parallel within a tolerance of 0.05 in. Final grip section width shall be at least 0.555.
6. The ends of the specimen shall be symmetrical in width with the centerline of the reduced section within 0.005 in.

Figure 3.1: Schematic drawing for experimental tensile sample used for high strain rate testing.

Figure 3.2: Schematic drawing of the experimental tensile sample illustrating locations for strain gages used for strain and load measurement.

3.3.2 High Strain Rate Tensile Testing

The tensile testing of all materials at strain rates 0.01 s^{-1} up to 200 s^{-1} were conducted using the high strain rate tensile testing methodology developed by Bruce [10] and Clarke [11]. As shown in Figure 3.2, two strain gages were mounted on each test sample, one for load measurement and one for strain measurement as discussed below.

3.3.3 Strain Measurement

Two different methods were used to measure strain during high-rate tensile tests: one through actuator displacement, and one with a strain gauge mounted on the gauge section of the specimen. Measuring strain by actuator displacement with the high-rate system is quite accurate at low strain rates due to the high rigidity of the 500 kN testing frame and sample dimensions. However, the data acquisition rate of the displacement measuring device is insufficient at high strain rates. A high elongation strain gauge (Vishay Measurements #EP-08-250BG-120) was used to measure strain at high strain rates. Special attention was given to ensure proper gauge attachment. The gauge, in ideal situations may measure strains up to 30%, however, as discussed by Clarke [11], the gauges rarely reached strains above 20%. A method of extrapolating the strain data to obtain a full stress strain curve has been developed, and was utilized in the mechanical testing of this test program [10-11].

3.3.4 Load Measurement

Two methods of acquiring load were used in testing the materials at high strain rates. A 40 kN piezoelectric load washer (Kistler Type 9071A) was placed in the upper grip assembly to measure load. The load washer provided adequate load data up to rates of 10 s^{-1} , however at high strain rates, load ringing appears in the load signal. Although the load ringing effect could be graphically diminished using a mathematical smoothing operation [10-12], important details of the yielding behavior are often masked by such simplification. A method for directly measuring the load using the grip section of the sample is evolved and is detailed in the thesis of Bruce [10]. By mounting an elastic strain gauge (Vishay # EA-06-125-BT-120) on the grip section of the specimen, and knowing the elastic modulus of the material and the specimen dimension at the grip section, it is possible to calculate the load. This method has proven to be effective as long as the grip section deforms elastically. Due to gripping limitations, and the low yield to tensile ratio of the dual-phase steels, the grip strain gage produced meaningful load measurements at low strain values, and for dual phase steels these data were supplemented with load measurements from the load washer following procedures outlined elsewhere [10-12].

3.3.5 Test Repeatability

In order to investigate the subtle differences between the mechanical behavior of TRIP and dual phase steels at a variety of strain rates, test results must be repeatable. To ensure that the testing procedure itself did not introduce additional uncertainty into the results of the high rate tests, engineering stress strain curves for a dual phase steel and a TRIP steel were obtained for two different strain rates of 100 s^{-1} and at 0.001 s^{-1} . Two samples were tested at each strain rate. Figure 3.3 represents the stress strain data for the samples tested and reveal that multiple testing at each strain rate yielded nearly identical data.

Figure 3.3: Engineering stress-strain curves of a) Dual-Phase Steel and b) High-C-Al TRIP steel tested at two different strain rates. At each strain rate, data on two tensile samples are shown superimposed. The strain axis for the high strain rate data has been off set by a strain of 0.40 for ease of presentation.

4.0 Microstructural Development in Dual Phase and TRIP Steels

4.1 Dual-phase steels

Table 2.2 summarizes the microstructure design for this study. Accordingly, tensile samples made from Low-C-Si, Med-C-Si and High-C-Si steels (Table 2.1 in Chapter 2) were annealed at 775 °C for 300 s and then quenched in water at room temperature. This heat treatment was designed to produce three different dual phase microstructures with similar martensite C-contents but with different martensite volume fractions. The resulting microstructures are shown in Figures 4.1-4.3. Table 4.1 presents the microstructure details of these three steels. It is revealed that three dual phase microstructures are characterized by similar martensite C-contents but exhibit different volume fractions. These three microstructures are designated henceforth as 'Low-Martensite', 'Med-Martensite' and 'High-Martensite' according to their martensite content. Table 4.1 also includes the microstructure details of the DP590 grade dual phase steel obtained from the IISI Round Robin program. Since this microstructure contains the lowest amount of martensite (~13%), it is designated as 'Extra-Low-Martensite' in this report. The microstructure of this steel is given in Figure 4.4. The microstructure of the fully martensitic steel produced in the lab is also shown in Figure 4.5. This microstructure was developed to assess the high strain rate deformation behavior of a fully martensitic steel.

Tensile samples from Low-C-Si were annealed at 825 °C for 300 s followed by water quenching. This heat treatment was designed to produce a high martensite content with a low martensite C-content. Figure 4.6 reveals the microstructure of Low-C-Si samples water quenched from 825 °C. This microstructure reveals similar martensite volume fraction (Table 4.1) to that of 'Med-Martensite' dual phase steel but has lower martensite C-content. This steel, together with the 'Medium-Martensite' steel (Table 4.1), will represent two different dual phase microstructures with similar martensite fractions but with different martensite C-contents. These two microstructures are designated in the study as 'LC-Martensite' (for low-carbon-martensite) and 'HC-Martensite' (for high-carbon martensite) according to their martensite carbon contents. High strain rate deformation studies of these microstructures will therefore, reveal the influence of martensite C-content on the strain rate dependence of flow behavior.

As revealed in the analysis presented in Table 4.1, a variation in martensite content is also associated with a change in ferrite grain size. However, the high strain rate deformation studies with fully ferritic steel (interstitial-free) by Bruce [Appendix B] revealed that grain size variations in ferrite apparently do not influence the strain rate dependent yield stress of ferrite, only the strength level. In this context, the high strain rate deformation behavior of the above dual phase steels having similar martensite C-contents but different martensite volume fractions will primarily depend only on the variation in martensite content.

Table 4.1: Microstructure details of dual phase steels produced.

Steel	Heat Treatment	Designation	Martensite volume fraction	Ferrite grain size, μm	Mart. C-content, wt. %	Micro-structure
Low-C-Si	775 °C WQ	Low-Martensite	~0.28	6	0.29	Similar martensite C-content but different volume fractions.
Med-C-Si		Med-Martensite	0.38 -0.40	5	0.32	
High-C-Si		High-Martensite	~0.49	3	0.35	
DP 590	NA	Extra-low-Martensite	0.13	7.6	NA	Low martensite content
Martensitic	880 °C WQ	Full-Martensite	~100%		0.10	Fully Martensitic
Low-C-Si	825 °C WQ	LC-Martensite	~0.39	7.3	0.21	Similar martensite volume fractions but different martensite C-content.
Med-C-Si	775 °C WQ	HC-Martensite	~0.39	5	0.32

Figure 4.1: Light optical micrograph of the 'Low-Martensite' dual phase containing about 28 volume percent of martensite. Picral - sodium meta bisulfate etched.

Figure 4.2: Light optical micrograph of the 'Med-Martensite' dual phase steel containing about 39 volume percent of martensite. Picral - sodium meta bisulfate etch.

Figure 4.3: Light optical micrograph of the 'High-Martensite' dual phase steel containing about 49 volume percent of martensite. Picral - sodium meta bisulfate etch.

Figure 4.4: Light optical micrograph of the 'Extra-Low-Martensite' dual-Phase steel containing 13 volume percent martensite. This material was obtained during the participation of ASPPRC in a round robin test program sponsored by IISI. on high strain rate testing of advanced high strength materials. Picral - sodium meta bisulfate etch.

Figure 4.5: Light optical micrograph of the fully martensitic steel prepared by heat treating a 0.10C-1.5Mn steel. Picral - sodium meta bisulfate etch.

Figure 4.6: Light optical micrograph of the 'Low-C-Mart' dual-Phase steel containing 39 volume percent martensite but with low martensite C-content (0.17). Picral - sodium meta bisulfate etch.

4.2 TRIP steels

As summarized in Table 2.3 of Chapter 2, tensile samples from Low-C-Si, Med-C-Si and High-C-Si were heat treated at 775 °C for 300 s and then isothermally held at 400°C for 200s followed by water quenching. This heat treatment produced three different microstructures with similar retained austenite C-contents but with different volume fractions of retained austenite. Figures 4.7-4.9 show three different TRIP microstructures produced by the heat treatment. The microstructures were revealed using the specialized etching technique as discussed in Appendix D. The detail microstructural analyses are given in Table 4.2. The microstructures in general reveal martensite in straw color, retained austenite in fine white particles with a bluish background, and ferrite as the white matrix as indicated in Figure 4.7. Light optical micrographs did not reveal carbide particles in the bainitic ferrite presumably due to the presence of Si which inhibits cementite precipitation. The first three microstructures represented in Table 4.2 are designated as Low-Austenite-LC (for Low-austenite fraction with Low Carbon Content), Med-Austenite-LC and High-Austenite-LC according to their austenite fraction.

The microstructures of these three Si-bearing TRIP steels reveal the presence of martensite or martensite-austenite constituents in the final microstructure. This is due to the low austenite carbon in the retained austenite particles immediately before water

quenching. Many isolated retained austenite particles may be seen in the microstructure of Low-Austenite-LC TRIP steel.

The Low-C-Si steel was annealed at 780 °C and then held at 420 °C for 300 s to produce a TRIP microstructure with high austenite C-content. Figure 4.10 shows the microstructure of this steel. This microstructure has an austenite volume fraction similar to the Low-Austenite-LC steel microstructure and reveals many fine isolated austenite particles. Austenite transformation behavior of this microstructure indicated higher stability of the retained austenite. This TRIP steel was designated as Low-Austenite-HC and is presented in Figure 4.11.

The High-C-Al steel (Table 2.3 in Chapter 2) tensile samples were heat treated as per the processing indicated in Table 2.3 in Chapter 2 to produce a TRIP microstructure with similar volume fraction to one of the Si-bearing TRIP steels to examine the effect of Al. Table 4.2 indicates that the Al-TRIP steel microstructure contains about 14% retained austenite but with higher austenite C-content than the Si-bearing Med-Austenite-LC steel microstructure. This microstructure is designated as Med-austenite-HC (for medium austenite fraction with high austenite carbon contents). High strain rate deformation results of Med-Austenite-LC and Med-Austenite-HC steels will reveal the influence of Al and Si on the strain rate sensitivity of these steels.

4.2.1 Effect of Alloying on TRIP Microstructure Evolution

From the microstructural examination it appears that Al influences the austenite transformation behavior significantly both at the intercritical annealing temperature and during isothermal transformation [Appendix I]. Al increases austenite carbon content. The specific characteristics of the microstructural features revealed in the Si-bearing and Al-bearing TRIP steels include:

- a) The ferrite grain size in the Si-bearing steel is finer than in the HC-Al steel.
- b) Si-bearing TRIP steels contain a larger fraction of M-A constituents than the Al-bearing TRIP steel.
- c) The bainite regions are greater in Si-bearing steels than in Al-bearing steel.
- d) The average grain size of the retained austenite particles is finer in the Al-bearing TRIP steel.

The bainite transformation was found to be significantly influenced by silicon and aluminum. The lower martensite content in the final Al-bearing TRIP steel indicates a carbon-enriched austenite at the onset of isothermal transformation. Higher austenite carbon contents retard transformation to martensite upon cooling to room temperature. Since the austenite carbon content is already high, further C-enrichment during isothermal holding stops the bainite transformation as the carbon content in the austenite reaches the T_0 -concentration (at T_0 , ferrite and austenite of identical compositions have the same free energy). Austenite enrichment to the level at T_0 , is believed to represent completion of the bainite transformation by a displacive mechanism) [13]. The microstructure of Med-austenite-HC steel (Figure 4.11) shows

very little bainite and supports this hypothesis.

Figure 4.12 shows the effect of alloying elements Si and Al on the T_0 temperature in the Fe-0.21C-1.5Mn system and the corresponding carbon concentration in austenite at the T_0 temperature derived by using Thermo-Calc. It is evident that silicon has limited influence on the T_0 line or the austenite C-content. Aluminum decreases the T_0 temperature or correspondingly increases the austenite carbon content at T_0 . It follows that alloying additions in TRIP steels which decrease the T_0 temperature will in fact decrease the driving force for $g \rightarrow a$ transformation and for bainite formation in particular. It is noted that the austenite carbon content in the Al-TRIP steel is higher at the start of the bainite transformation during isothermal holding. The limited extent of bainite transformation along with an increased carbon content in the retained austenite in the Al-TRIP steel may therefore be explained from the above discussion.

Table 4.2: Microstructural characterization of TRIP steels presented in Figures 4.7-4.9.

Steel	Heat Treatment	Designation	Retained Austenite Content	Retained Austenite C-content, wt.%	Ferrite Grain Size, μm	Micro-structure
Low-C-Si	775 – 300s 400 – 200s	Low-Austenite-LC	9.1%	1.06	9.3	Similar austenite C-content but with different austenite fractions
Med-C-Si		Med-Austenite-LC	14.0%	1.01	7	
High-C-Si		High-Austenite-LC	16.0%	1.09	6.35	
Low-C-Si	780-300s 420-300s	Low-Austenite-HC	10.4	1.0-1.2	9.2	
High-C-Al	780-300s 420-200s	Med-Austenite-HC	13.5	1.44	9.3

Figure 4.7 Light optical micrograph of 'Low-austenite-LC' TRIP steel with retained martensite particles revealed in straw color. Picral - sodium meta bisulfate etch. A: austenite, B: bainite, MA: martensite-austenite, M: martensite.

Figure 4.8: Light optical micrograph of 'Med-austenite-LC' TRIP steel showing more martensite particles and larger fraction of bainitic ferrite. Picral - sodium meta bisulfate etch.

Figure 4.9: Light optical micrograph of 'High-austenite-LC' TRIP steel with higher amount of retained austenite. Significant bainitic transformation. Microstructure reveals martensite and martensite-austenite constituents. Picral - sodium meta bisulfate etch.

Figure 4.10: Light optical micrograph of 'Low-austenite-HC' TRIP steel with fine isolated austenite particles. Limited bainitic transformation. Picral- sodium meta bisulfate etch.

Figure 4.11: Light optical micrograph of 'Med-austenite-HC' TRIP steel with retained austenite particles and limited Martensite. Picral - sodium meta bisulfate etch.

Figure 4.12: Effect of alloying additions on the allotropic phase boundary T_0 and corresponding distribution of carbon in austenite in the low alloy TRIP steels as computed by Thermo-Calc program.

5.0 Effects of Strain Rate on Mechanical Properties of Sheet Steels

5.1 High Strain Rate Deformation Behavior of Interstitial-Free Steels

The microstructural matrices of most advanced high strength steels (AHSS) are primarily ferrite (~75%). It is expected that properties of ferrite will control in a large way the dynamics of high strain rate deformation. In order that the high rate deformation behavior of dual phase and TRIP steels be properly understood, the strain rate sensitivity properties of single phase ferritic steels were examined. Interstitial-free (IF) steels with different engineered microstructural variations were tested at different strain rates to examine the influence of various microstructural parameters on the high strain rate deformation behavior. These steels are summarized in Table 5.1 and in the tables presented previously. The chemical compositions indicate that all the IF steels in Table 5.1 are stabilized, either with Ti and/or Nb. The influence of following microstructural variations of IF steels were studied:

1. Composition of base IF steel (IF -1, IF -2, IF-3 of Table 2.4 in Chapter 2) as modified by solid solution alloy additions (IF -4, IF-3 of Table 2.4 and IF -6 of Table 2.5 in Chapter 2).
2. Dislocation density as controlled by cold work (by tensile pre-straining of IF -2 steel of Table 2.4 in Chapter 2).
3. Ferrite grain size (IF -1 in Table 2.4 of Chapter 2) ,
4. Bake hardening response (IF -6 and BH steel in Table 2.5 of Chapter 2).

Table 5.1 Summary of Interstitial-free steels and the chemical composition in wt.% selected for high strain rate studies.

Designation	Reference		C	Mn	P	Si	Ti	Nb
IF-1	Appendix B	Base	0.0038	0.17	0.003	0.002	0.072	
IF-2	Appendix B	Base	0.0024	0.10	0.006	0.01	0.024	0.059
IF-3	Appendix B	Base	0.0030	0.20	0.010	NA	0.05	
IF-4	Appendix B	Mn+P	0.0040	0.90	0.090	NA	0.05	
IF-6	Appendix J	P	0.0090	0.24	0.064	NA	-	0.05

Figure 5.1 shows the stress strain behavior of IF-1, IF-3, IF -4 and IF -6 steels as a function of strain rate. The steels were tested at strain rates from 0.001 s^{-1} to 900 s^{-1} . All the IF steels reveal significant strain rate sensitivity. Both the yield and tensile stress increase with an increase in strain rate. Yielding is continuous at lower strain rates for all IF-3, IF-4 and IF-6 steel but discontinuous yielding appears to develop at higher strain rates. IF -1 steel reveals discontinuous yielding at all imposed strain rates and the yield point elongation becomes more pronounced at higher strain rates.

Figure 5.1 Engineering stress-strain data for (a) IF-1 steel, (b) IF-3 steel, (c) IF-4 steel and (d) IF-6 steel for different strain rate of testing.

5.1.1 Interstitial Free Steel- Influence of Solid Solution Strengthening

The engineering stress strain data of Figure 5.1 for IF -3, IF-4 and IF-6 steels may further be analyzed to reveal a possible influence of solid solution strengthening effect on strain rate sensitivity. From Table 5.1, it is clear that the interstitial contents in all the IF steels have been stabilized with either Ti or Nb additions or with a combination of both (IF-2). It may therefore, be inferred that a possible variation in the strain rate sensitivity of the IF steels will only be due to a difference in the amount of solid solution additions other than C or N. Solid solution additions Mn and P are highest in IF-4 steel. Figure 5.2 presents the yield stress of steels IF-3, IF-4 and IF-6 as a function of strain rate. The figure clearly suggests that the strain rate sensitivity of IF-4 steel, which is rich in solid solution elements Mn and P, is the lowest of the three IF steels plotted. IF-6

steel, having lower P content than IF-4 but higher than IF-3 steel, shows intermediate values of strength increase with strain rate increase. The logarithmic strain rate sensitivity (β) as explained in Section 4.5 in Appendix B has the highest value for IF-3 steel. It is however, evident from Figure 5.2 that the solid solution strengthening influence on strain rate sensitivity diminishes at higher strain rates similar to solid solution softening at low temperatures [14].

Figure 5.2: Yield stress values for three different IF steels as a function of strain rate.

5.1.2 Interstitial Free Steel- Influence of Grain Size

Grain size is reported to be a long range barrier to the dislocation motion in bcc lattice, and hence the effect of strain rate on strength, which is controlled by short range interactions and thermal fluctuations, may not be a significant function of strain rate. This supposition was further tested in the present study for IF steel. IF-1 steel indicated in Table 5.1 was heat treated to produce different grain size variations, and the details of the heat treatment and testing are described in Section 4.8.1 in Appendix B.

Figure 5.3 shows the engineering stress-strain data for four different grain sizes of IF-1 steel tested at two different strain rates of 0.001 s^{-1} and 35 s^{-1} . The stress increase due to an increase in strain rate from 0.001 s^{-1} to 35 s^{-1} for all the grain sizes is similar. This is more clearly evident in the results of Figure 5.4 summarizing the yield stress of the IF-1 steel for different grain sizes plotted according to the Hall-Petch equation:

$$\sigma_y = \sigma_o + k_y d^{-1/2} \quad [5.1]$$

where σ_y is the flow stress, usually taken at a strain of 0.002 for yielding, σ_o and k_y are the common Hall-Petch parameters (k_y is often referred to as the grain boundary

strength coefficient), and d is the grain size. From Figure 5.4, it is noted that the yield stress increment with increasing strain rate is independent of grain diameter, and the grain size effect on the strain rate sensitivity is almost absent for the IF steels evaluated here. These results further corroborate the fact that changes in grain size do not significantly influence the strain rate dependent stress increment in essentially single phase ferritic steels (IF) and that grain boundaries are long-range (athermal) obstacles to dislocation motion. The data in Figure 5.4 also show that s_0 increases with strain rate. This observation is consistent with previous interpretations (as discussed in Appendix F) which consider that s_0 is comprised of a thermal component (s^* , short range barrier) and an athermal component (s_a , long range barrier). Thus, the strain rate dependence of s_0 reflects primarily the effects of strain rate on s^* .

Figure 5.3: True stress-strain curves at two strain rates for IF-1 steel with four different grain sizes. All sample gage lengths were 25.4 mm.

Figure 5.4: Yield stress versus (grain diameter)^{-1/2} for IF-1 steel tested at five different strain rates. All sample gage lengths were 25.4 mm.

5.1.3 Interstitial Free Steel- Influence of Cold Work

Tensile samples from IF-2 steel were divided in five different batches and given 0%, 2%, 5%, 10% and 18% tensile prestrain to introduce different amounts of cold work (e.g. dislocation density) in the ferrite matrix and were used to examine the strain rate dependence of flow behavior. The details of the specimen preparation and testing are described in Section 4.6 in Appendix B. Prestraining was done at a tensile strain rate of $4 \times 10^{-4} \text{ s}^{-1}$, and the prestrained samples were further tested at strain rates from 0.001 to 500 s^{-1} .

Engineering stress-strain curves for the prestrained samples tested at a strain rate of 0.001 s^{-1} are shown in Figure 5.5a. Figure 5.5a shows the increasing yield strength and tensile strength, and decreasing strain hardening with increasing amount of pre-strain. Figure 5.5b presents the same data of Figure 5.5a but the stress-strain curves are offset by the amount of respective prestrain. The overlapping flow curves in Figure 5.5b suggests that the strength only reflects work hardening and not subsequent strain aging.

Figure 5.6a summarizes strain rate sensitivity for the IF-2 steel by plotting flow stress at a true strain of 0.02 as a function of strain rate for all five pre-strain levels. For all pre-strain levels, Figure 5.6a shows two distinct regions of strain rate sensitivity and a constant increase of flow stress for any level of prestrain at all strain rates of testing. Figure 5.6a further supports the observation that strain rate sensitivity does not vary significantly with different amounts of pre-strain. Figure 5.6b is a representation of some of the data contained in Figure 5.6a. This figure plots flow stress at 0.02 true strain versus the pre-strain level for three different strain rates. Figure 5.6b illustrates

that the strength increment from a low strain rate to a high strain rate is approximately constant regardless of the amount of quasi-static pre-strain.

Figure 5.5: Stress-strain curves at a strain rate of about 0.001 s^{-1} for IF-2 steel with five different pre-strain levels. a) Engineering stress-strain curves. b) True stress-strain curves with pre-strained results offset by the prestrain amount.

Figure 5.6: a) Flow stress at 0.02 true strain versus strain rate for an IF steel with five different amounts of pre-strain. b) Flow stress at 0.02 true strain versus pre-strain amount for an IF steel at three different strain rates.

5.2 High Strain Rate Deformation Behavior of Ultra Low Carbon - Bake Hardenable (BH) Steels

Bake hardenable steels are increasingly used in passenger cars for reduced car weight and increased crashworthiness. Chemistry and processing of these steels are designed to produce steels that have both high formability and sufficient strength. Bake hardenable steels are designed to have a small amount of interstitial carbon or nitrogen in solution. During paint baking (usually carried out at low temperature of 150°C) of formed automotive components, the interstitial solute atoms diffuse to the existing dislocations generated during forming and immobilizes them. It is believed [2] that new dislocations must be generated to cause plastic strain. The applied stress has to increase to generate and propagate new dislocations resulting in the observed increase in yield strength in bake hardenable steels.

With the increased application of bake hardenable steels in the automotive industry and the emphasis on predicting material behavior during a crash where strain rates may exceed 100 s⁻¹ [3], it is important to understand the mechanical behavior of BH steels over all strain rates anticipated in a crash situation. To date, the primary data available on BH steels were obtained at quasi-static strain rates on samples prestrained in tension a fixed amount of strain (e.g. 2 pct) and subjected to a short time age (typically 20 min. at 170°C) [4]. This study was designed to evaluate the strain rate sensitivity of a BH steel and compare the high strain rate deformation properties with that of a stabilized interstitial free steel, both in the as-received condition and after a prestrain and aging treatment, to assess the interrelationships between strength increases due to work hardening, solute pinning, and strain rate hardening. The chemical compositions of the stabilized IF and BH steels used here are given in Table 5.2. Since both the steels contain similar amounts of solid solution elements, it is expected that any variation in the strain rate sensitivity will be a manifestation of solute carbon content only.

Table 5.2 Chemical composition in wt.% for the steels used in the present study.

Steel	Reference	C	P	Mn	Nb	Al
IF-6	Appendix J	0.009	0.064	0.24	0.025	0.050
BH	Appendix J	0.008	0.078	0.32	0.005	0.022

Tensile samples from both the steels were tested over a wide range of strain rates, from 10⁻³ to approximately 900 s⁻¹. Both steels were tested in the as-received condition, and after a prestrain and bake treatment (samples were prestrained in tension at a quasi-static rate, followed by a 30 minute aging at 177 °C). Longitudinal tensile samples with a 25.4 mm reduced gauge length were machined with the as-received thickness. Figures 5.7a and 5.7b display the quasi-static stress-strain curves for the IF-6 steel and BH steel in the as-received condition and after bake hardening treatment (2 pct strained and 30 min. @ 177 °C). In each figure, the curve for the prestrained steel sample is offset by a strain of 0.02 (amount of prestrain). The

mechanical properties of the prestrained and baked IF-6 steel strongly match the properties of the as-received material. The yield strength of the prestrained and baked sample increased only by the work hardening increment (28 MPa) of the as-received material during prestraining. The IF-6 steel shows a continuous yielding behavior in the as-received condition and also after prestraining and baking. This suggests that the IF-6 steel is indeed stabilized and there is no pinning of dislocations even after baking heat treatment.

The plot in Figure 5.7b for the bake hardenable steel shows discontinuous yielding in the flow behavior for the as-received sample and the yield point elongation increases after straining and baking treatment. This is indicative of the presence of solute interstitials in the matrix of the BH steel in the as-received condition and the increased yield point elongation is suggestive of pinning of mobile dislocations (generated during straining) during baking. There is a distinct increase in yield strength after prestrain and baking. This increase in yield strength can be divided into two contributions: work hardening and bake hardening. The bake hardening contribution can be found by subtracting the flow stress after 2 pct. prestrain from the lower yield strength after baking, as seen in the inset in Figure 5.7b. This value (i.e. BH) was determined to be 47 MPa at quasi-static strain rates, and the work hardening contribution was found to be 26 MPa.

Figure 5.8a displays the yield strength for each test condition for the IF-6 steel plotted as a function of strain rate. In this figure, yield strength is taken as the 0.2 % offset yield strength for samples. There are many important strain rate effects shown in Figure 5.8a. First, for all test conditions, the yield strength increased with strain rate. The strength increase after prestraining and aging for the IF-6 steel remained approximately constant at all strain rates, at 28 MPa in comparison to the yield strengths of the as-received condition at all strain rates. This implies that the increase in dislocation density associated with prestrain had no effect on the strain rate sensitivity of the material. This is in conformity to the results presented in Section 5.1 for the high strain rate deformation behavior of IF steels. Cold work introduces long-range obstacles to dislocation movement, and contributions of these obstacles to flow stress are independent of temperature or strain rate.

The strain rate sensitivity of the yield strength also increased with strain rate. This phenomenon is better seen by examining Figure 5.8b, which plots β for the lower yield strength as a function of strain rate. β is a measure of strain rate sensitivity and is derived from the slope of a semi-logarithmic plot of strain rate versus flow stress as given below:

$$b = \frac{\partial s}{\partial (\log \dot{\epsilon})} \quad [5.2]$$

where σ = flow stress, $\dot{\epsilon}$ = strain rate, and β = the semi-logarithmic strain rate sensitivity coefficient. A mechanical property, in this case yield strength, is plotted as the function of the log of strain rate and the slope of this line is commonly referred to as the β value. A high β value correlates to high strain rate sensitivity. As shown, the sensitivity to strain rate in the IF-6 steel is insensitive to the aging treatment. These results are consistent with several prior studies [11,12,13].

Figure 5.7: Quasi-static true stress-strain curves for a) the IF-6 steel in the as-received condition and after prestrained and baked condition, and b) the BH steel in both conditions. The inlay demonstrates the measurement of WH and bake hardenability.

Figure 5.8: (a) Yield strength values of the IF-6 steel and (b) strain rate sensitivity value β for yield strength of the IF-6 steel, plotted as a function of strain rate.

Figure 5.9a displays the lower yield strength of the BH steel in the as-received condition and after a paint baking treatment (2 pct stained and 30 min. @ 177 °C) plotted as a function of strain rate. The results were similar to the results of the IF steel in that 1) the yield strength for each condition increased with strain rate and 2) the sensitivity of the yield strength increased with strain rate. However, there was one distinct difference in the behavior of the BH and IF steels. The strain rate sensitivity of the prestrained and baked BH steel was less than the as-received condition. Figure 5.9b is a plot of β values for yield strength for the BH steel. It clearly shows that for all rates studied the prestrained and baked condition was less sensitive to strain rate.

Figure 5.9: (a) Yield strength values of the BH steel as a function of strain rate and (b) strain rate sensitivity values β for yield strength of the BH steel plotted as a function of strain rate.

Figure 5.10 compares the effects of strain rate on the strength increase after prestrain and aging for the IF-6 and BH steels. At each strain, this figure plots the difference between the yield strength after prestrain and baking and the yield strength of the as-received materials presented in Figures 5.8a and 5.9a. As shown, the strength increase for the IF steel remains essentially constant at about 28 MPa, while the strength increase for the BH steel decreases from about 70 MPa at quasi-static strain rate to essentially the same value as the IF steel at high strain rates. At each strain rate the difference in the two curves shown in Figure 5.10 is the bake hardening increment, which decreases from approximately 47 MPa at low strain rates to essentially zero at high strain rates.

Figure 5.10: Difference between the lower yield strength of the prestrained and baked condition and the as received condition as a function of strain rate for the two steels studied.

Three major observations from these data on the response to bake hardening are: (a) at quasi-static strain rates, there is a significant increase in strength due to aging and a return of a well defined yield point on loading; (b) the strain rate sensitivity of the baked materials at the lower strain rates is much lower than the as received, unbaked material; and (c) the contribution of bake hardening decreases with an increase in strain rate. These observations can be rationalized by noting that the stress required to move dislocations in iron is comprised of thermal and athermal components. For the IF-6 steel and the un-aged material discussed here, the strain rate dependencies shown in Figures 5.8b and 5.9b are almost identical. The increase in the dependency of strength with strain rate suggests that there is an increase in the contribution of thermally activated dislocation processes, e.g. the formation of double-kink pairs, as dislocations move past Peierls barriers, to the overall stress required for dislocation movement such that at high strain rates the properties are dominated by the stress for individual dislocation movement.

In contrast, for the prestrained and baked BH steel at the lower strain rates, the high strengths and low sensitivity of strength to strain rate suggest that the mechanisms responsible for strengthening are different. If it is assumed that the process of aging produces a fully pinned dislocation structure, then on reloading the stress for deformation corresponds primarily to the stress to generate sufficient new dislocations as controlled by the long range stress fields associated with the dislocation structure and the stress required to operate dislocation sources. In this strain rate regime, the flow stress would exhibit a low and mild dependency on strain rate, as observed. However, with an increase in strain rate the inherent resistance to the movement of individual dislocations by the same processes operating for the unaged material dominate, and the effects of pinning due to aging are overshadowed by the inherent

processes of dislocation movement over short range barriers, such as Peierls barriers. Therefore, both materials in this study exhibit equivalent flow stresses and strain rate sensitivities at the high strain rates and these results are considered further in Appendix J.

5.3 High Strain Rate Deformation Behavior of High-Strength Low-Alloy (HSLA) Steel

High strength low alloy steels were included in the present research program to study high strain rate deformation behavior of ferritic-pearlitic microstructures. These steels are strengthened by a variety of mechanisms, including solid solution strengthening, precipitation strengthening, and grain size refinement. Since solid solution strengthening is expected to be the only short-range barrier in these steels, the high strain rate deformation behavior should reflect the same features as seen in IF steels.

Table 5.3 shows the chemical compositions of two HSLA steels chosen for the study of high strain rate deformation of steels with ferritic-pearlitic microstructures. Tensile samples were prepared from HSLA-1 and HSLA-2 samples and tested in tension at strain rates from 0.01 s^{-1} to 500 s^{-1} . These two steels differ greatly in chemical composition and HSLA-2 has greater degree of solid solution strengthening than HSLA-1.

Table 5.3 Chemical compositions of HSLA steels chosen for high strain rate deformation study (wt. pct.)

Designation	Reference	C	Mn	P	Si	Ti	Nb
HSLA-1	Appendix B Table 2.4 Chapter 2	0.050	0.390	0.011	0.016	0.006	0.029
HSLA-2		0.09	1.46	0.020	0.08	NR	0.045

Figure 5.11 shows engineering stress-strain curves for HSLA-1 and HSLA-2 steels. The HSLA-1 steel shows yield point elongation during elastic to plastic transition and the extent of yield point elongation increases with increasing strain rate. The yield and tensile strengths both increase with increasing strain rate. The strain hardening rate appears to decrease with increasing strain rate. In contrast to HSLA-1 steel, HSLA-2 steel reveals nearly continuous yielding at all strain rates. The yield and tensile strengths increase and also the strain hardening appears unchanged with increasing strain rate.

Figure 5.11: Engineering stress-strain data for (a) HSLA-1 and (b) HSLA-2 steel for different strain rate of testing.

Figure 5.12a shows the yield stress as a function of strain rate for HSLA-1 and HSLA-2 steels and the strain rate sensitivity of these two steels can be interpreted. The HSLA-1 steel shows higher strain rate sensitivity than the HSLA-2 steel. In order to examine the influence of only solid solution strengthening on the strain rate sensitivity of HSLA-steels, the yield strength of HSLA-2 steel, which is richer in Mn content (Table 5.3), was modified to take in to account only the strengthening due to increase solid solution content. The accounting procedure is described in detail in Section 6.2.2 of Appendix B. The modified yield strengths of HSLA-2 steel have been plotted versus strain rate in Figure 5.12b and compared with the yield strength of HSLA-1. It is interesting to note that the solid solution influence is very similar to that of IF steels shown in Figure 5.2 and the influence diminishes at higher strain rates.

In summary, it has been demonstrated through the above series of high strain rate deformation testing of microstructures ranging from pure ferritic (IF) to ferritic-pearlitic (HSLA) that strengthening due to cold work and grain refinement does not influence strain rate dependence of yield stress. Solid solution strengthening has a significant influence on the strain rate sensitivity of ferrite. The solid solution effect however, diminishes at high strain rates. The presence of interstitials decreases the strain rate sensitivity of ferritic matrix. Bake hardenability exerts a similar influence on the strain rate sensitivity of pure ferritic steel as solid solution strengthening.

Figure 5.12: (a) Yield strength versus strain rate for two HSLA steels with different solute content. (b) Modified yield strength of HSLA-2 steel (accounting for only solid solution strengthening) as compared to HSLA-1 steel as a function of strain rate.

5.4 Quasi Static Mechanical Properties of Dual Phase Steels

The quasi static ($10^{-3} s^{-1}$) engineering and true stress-strain plots of dual phase steels with different martensite volume fractions are given in Figure 5.13. A summary of the tensile properties along with microstructural details, including martensite volume fractions and calculated martensite carbon contents, is presented in Table 5.4. The processing details for these microstructures have been summarized in Table 4.1 in Chapter 4. Based on the martensite volume fraction, the steels are designated as Low-Martensite, Med-Martensite, High-Martensite and Extra-Low-Martensite as indicated in Table 5.1. It is evident from the mechanical property details (Table 5.4) that the yield and tensile strength of the dual phase steels increases with increase in martensite fraction. Yielding is continuous in all the dual phase steels as seen in Figure 5.13. The yield strength measurements for the DP steels in this study were calculated by applying the 0.2% strain offset method.

Table 5.4: Quasi-static mechanical properties and microstructural features for each DP steel studied.

Steel	Martensite Volume Percent (%)	Martensite Carbon Content	Quasi-Static Properties			
			YS $_{0.2\%}$, MPa	UTS, MPa	YS/UTS	Unif. El. %
Extra-Low-Martensite	13	N/A	370	620	0.60	18.9
Low-Martensite	28	0.29	446	871	0.51	13.2
Med-Martensite	39	0.32	564	1178	0.48	8.6
High-Martensite	49	0.35	654	1397	0.47	8

Figure 5.13 reveals that the yield, tensile stress and the work hardening of the dual phase steels increases with increasing martensite fraction. The strength and uniform elongation values have been plotted in Figure 5.14 as a function of martensite volume fraction. The uniform elongation systematically decreases with an increase in strength, i.e. an increase in martensite volume fraction. The essentially linear increase in tensile strength up to 49% martensite fraction may be attributed to the increase in martensite fraction in the microstructure. The chemical strength of the ferrite may be assumed to be same in all three dual phase microstructures, as the steels have been quenched from the same intercritical temperature. The grain size and dislocation density introduced by the different amounts of martensite in the ferrite will change the yield strength of the ferrite.

Figure 5.13: Engineering and true stress-strain plots for dual-phase steels at quasi-static strain rate ($\dot{\epsilon} = 0.001 \text{ s}^{-1}$).

Figure 5.14: Variation in mechanical properties ($\dot{\epsilon} = 0.001 \text{ s}^{-1}$) of experimental dual phase steels as a function of martensite content.

5.5 High Strain Rate Properties of DP Steels

The results of the high strain rate testing of each dual phase steel microstructure indicated in Table 5.4 are presented in this section. The figures present true stress strain curves that are offset by 5-10% strain for each subsequent strain rate. This allows comparison, in a single figure, such as in Figure 5.15, of the work hardening behavior for a particular material at each strain rate. Figures such as Figure 5.16 summarize the yield strength and UTS for each material as a function of strain rate on a semi-logarithmic scale.

Figure 5.15 shows the true stress-strain curves for Low-martensite dual phase steel (28% martensite) over a strain rate range of 0.001 s^{-1} to 75 s^{-1} . The figure indicates that the work hardening behavior does not significantly change over the range of strain rates examined in this study. The high strain rate curves show apparently serrated flow behavior. It is believed this is a result of load ringing during the test and that the material in reality exhibits continuous yielding behavior. The material has a low uniform elongation at all strain rates. This is a result of the martensite volume fraction (28%) present in the microstructure. Figure 5.16 plots the yield strength and the UTS for Low-martensite dual phase steel as a function of strain rate. The results indicate that the yield strength of the material is mildly sensitive to strain rate. The flow stress values of Low-martensite steel increase with strain rate over the entire range of deformation rates examined.

Figure 5.17 shows the true stress versus true strain curves for the Med-Martensite steel (38% martensite) and indicates a similar response to increased strain rate as Low-Martensite steel. The work hardening behavior of the Med-Martensite steel does not significantly change with strain rate. The slight sensitivity of the UTS to strain rate of Med-Martensite steel is demonstrated in Figure 5.18, which plots the yield and ultimate tensile strength as a function of strain rate. The yield strength of Med-Martensite steel is slightly more sensitive to strain rate than the UTS.

Figure 5.15: True stress – True strain plots of Low-Martensite Dual-phase steel (28% martensite) tested at different strain rates. Each stress-strain plot is offset by 0.05 true strain for clarity. Duplicate tensile tests are shown superimposed in this figure, indicating the excellent reproducibility of the results.

Figure 5.16: Variation in yield stress (at 0.2% engineering strain offset) and ultimate tensile strength of Low-Martensite dual-phase steel as a function of different strain rates.

Figure 5.17: True stress – True strain plots of Med-Martensite dual-phase steel tested at different strain rates. Each stress-strain plot is offset by 0.05 true strain for clarity. Duplicate tensile tests are shown superimposed in this figure, indicating the excellent reproducibility of the results.

Figure 5.18: Variation in yield stress (at 0.2% engineering strain offset) and ultimate tensile strength of Med-martensite Dual-phase steel as a function of different strain rates.

Figure 5.19 shows the true stress – true strain curves for High-martensite steel (49% martensite) over the strain rate range of 0.00118 to 64.6 s^{-1} . Similar to the Low-martensite and Med-martensite dual phase steels, the high-martensite dual phase steel does not exhibit a noteworthy change in work hardening behavior with an increase in strain rate. High-martensite steel is the highest strength material, with a quasi-static UTS of 1397 MPa, and has the highest volume fraction of martensite (49%). The high volume fraction of martensite results in the limited ductility demonstrated at all strain rates. Figure 5.20 plots the yield strength and UTS of high-martensite steel as a function of strain rate. The figure indicates that the YS and UTS of high-martensite steel are virtually insensitive to strain rate.

In comparison to all of the dual phase steels produced from the experimental steels prepared for this research, the Extra-low-martensite DP590 steel (13 % martensite) has the lowest strength and highest ductility. With increasing strain rate the work hardening increases as shown in Fig. 5.21. The yield strength and UTS of Extra-low-martensite steel increase with strain rate. This is more evident in Figure 5.22, which plots the yield strength and UTS of Extra-low-martensite steel as a function of strain rate. The yield strength of Extra-low-martensite steel is more sensitive to strain rate than the other dual phase steels.

Figure 5.19: True stress – True strain plots of High-martensite Dual-phase steel (49 % martensite) tested at different strain rates. Each stress-strain plot is offset by 0.05 true strain for clarity. Duplicate tensile tests are shown superimposed in this figure, indicating the excellent reproducibility of the results.

Figure 5.20: Variation in yield stress (at 0.2% engineering strain offset) and ultimate tensile strength of High-Martensite Dual-phase steel (49 % martensite) as a function of different strain rates.

Figure 5.21: True stress – True strain plots of Extra-Low-Martensite Dual-phase steel (13 % martensite) tested at different strain rates. Each stress-strain plot is offset by 0.10 true strain for clarity.

Figure 5.22: Variation in yield stress (at 0.2% engineering strain offset) and ultimate tensile strength of Extra-Low-Martensite Dual-phase steel (13 % martensite) as a function of different strain rates.

5.6 Summary of Strain Rate Sensitivity of Dual phase Microstructures

From the results of the strain rate dependence of the yield stress as a function of martensite volume fraction in different dual phase microstructures, it is worth examining the strain rate sensitivity of different dual phase microstructures. A simple way to examine this is by calculating the ratio of dynamic yield and tensile strengths to static yield and tensile strengths for each of the dual phase steels studied. The strength ratios were calculated by using the yield or tensile strength at a strain rate of 100 s^{-1} as the dynamic strength, and the yield or tensile strength at a strain rate of 0.00118 s^{-1} as the static strength. Table 5.5 shows the strength ratios calculated for the dual phase microstructures mentioned in Table 5.4. It is apparent that the yield strength of all the dual phase steels is more sensitive to strain rate than the tensile strength. It is also evident from Table 5.5 that as the martensite content increases in the dual phase microstructure, the strength ratio decreases, i.e. the material becomes less sensitive to strain rate changes. High-martensite steel which has the highest yield and tensile strength is least sensitive to strain rate increase.

Table 5.5: Dynamic/Static stress ratios of the yield and tensile strengths of all DP steels tested.

Steel	Martensite Volume Percent	$\sigma_{100 \text{ s}^{-1}} / \sigma_{0.001 \text{ s}^{-1}}$	
		Yield Strength	UTS
Extra-Low-Martensite	13 %	1.43	1.13
Low-Martensite	28 %	1.28	1.10
Med-Martensite	38 %	1.20	1.03
High-Martensite	49 %	1.11	1.04

The dynamic to static strength ratios with increasing martensite fraction have been plotted in Figure 5.23 and are compared with the corresponding values for the IF steel mentioned in Table 2.5 of Chapter 2. The influence of increasing martensite fraction on the decreasing strain rate sensitivity is clearly suggested. This result has also been reported in literature. It is worth mentioning that the martensite in three dual phase microstructures e.g., Low-Martensite, Med-Martensite and High-Martensite has essentially the same amount of carbon.

The relative insensitivity of yield and tensile strength of High-martensite dual phase steel to high strain rate testing was further ascertained by creating a fully martensitic microstructure and testing at various high strain rates. A plain C-Mn steel with nominal chemical composition of 0.10 wt.% C and 1.5 wt.% Mn was processed to a fully martensitic microstructure. Tensile samples made out of the steel were subjected to tensile testing at various strain rates. Figure 5.24 shows the engineering stress-strain curves for the fully martensitic material tested over a range of strain rates. As Figure 5.24 indicates, the material shows little response to changes in strain rates similar to that revealed for the High-Martensite dual phase steel shown in Figure 5.19.

Figures 5.25 and 5.26 summarize the yield strength and tensile strength data of all the dual phase steels tested at different strain rates together with that of the high strain rate data of IF steel mentioned in Table 2.5 of Chapter 2. The figure reveals the gradual decrease in strain rate sensitivity of ferritic microstructures with increase in martensite fraction.

Figure 5.23: Dynamic/Static stress ratios of the yield and tensile strengths of all DP steels tested and compared with that of an interstitial-free steel with a composition mentioned in Table 2.5 of Chapter 2.

Figure 5.24: Engineering stress-strain data for various strain rates for a fully martensitic steel.

Figure 5.25: Variation in yield stress of different dual phase steels (13, 28, 38, and 49 % martensite) and an IF steel as a function of strain rates.

Figure 5.26: Variation in tensile strength of dual phase microstructures as a function of strain rate. The tensile strength data of the extra-low-martensite steel and IF steel have also been plotted in the figure for comparison purposes.

5.7 High Strain Rate Deformation Behavior of TRIP (Transformation-induced Plasticity) Steels

Table 4.2 in Chapter 4 lists the microstructure details of the TRIP steels that were produced with different volume fractions of retained austenite, but with similar austenite carbon contents. Three TRIP microstructures with retained austenite contents of 9.1%, 14% and 16% were produced from three different Si-bearing TRIP steels. These microstructures are designated as Low-austenite-LC, Med-austenite-LC and High-austenite-LC TRIP according to their retained austenite contents. 'LC' designation is used to distinguish these microstructures from other TRIP microstructures having higher retained austenite carbon-contents.

The quasi-static stress-strain behavior of the three TRIP steels are shown in Figure 5.27. The Low-austenite-LC steel (9.1 % γ) shows yield point elongation during yielding. The strain hardening behavior in the other two steels are almost identical. The Med-austenite-LC TRIP steel (14 % γ) shows lower total elongation than the High-austenite-LC steel (16 % γ). The Med-austenite-LC and High-austenite-LC steels show similar austenite transformation behavior with deformation strain. The early necking in the Med-austenite-LC steel is interpreted to reflect the lower austenite fraction compared to the High-austenite-LC steel. The yield point elongation in Low-austenite-LC steel is due to the presence of low amount of martensite and also limited bainitic transformation. The High-austenite-LC TRIP steel shows almost continuous yielding behavior due to the presence of large amount of martensite and also a larger fraction of

bainitic ferrite. Bainite formation introduces mobile dislocations in the ferrite matrix, which are responsible for the increase in ferrite strength as well as continuous yielding.

Figure 5.27: Engineering stress-strain curves for the TRIP steels with different retained austenite volume fractions (9.1, 14, and 16 % γ) tested at a strain rate of 0.001s^{-1} . The inset figure is a blow out plot of the yielding stage for three TRIP steels.

Figure 5.28 shows the true stress-true strain plots for the Low-Austenite-Low C TRIP steel (9.1 % γ) for strain rate range of 0.001 s^{-1} to 84 s^{-1} . The steel shows a yield point behavior at low strain rate. At higher strain rates a pronounced yield drop is noticed. The work hardening behavior is similar at all rates of testing. The sensitivity of the yield and the tensile stresses to strain rate are similar as seen in Figure 5.29.

The true stress-strain curves for the Med-austenite-LC TRIP steel (14% γ) as a function of strain rate are shown in Figure 5.30. This steel shows a higher rate of work hardening than the Low-Austenite-LC TRIP steel at all strain rates. The steel also shows continuous yielding at all strain rates. The yield stress shows slightly higher strain rate sensitivity than the tensile strength as illustrated in Figure 5.31.

Figure 5.32 shows the true stress-strain curves for High-austenite-LC TRIP steel (14% γ). The steel shows continuous yielding at all strain rates. The strain rate sensitivity of this steel is illustrated in Figure 5.33 which shows dependencies similar to those observed for the other two TRIP steels.

Figure 5.28: True stress-true strain plots for Low-Austenite-LC TRIP steel (9.1% γ) at different strain rates. Each pair of duplicate curves is offset by 0.10 strain for clarity.

Figure 5.29: Variation in yield stress and ultimate tensile stress of Low-Austenite-LC TRIP steel (9.1% γ) with strain rates.

Figure 5.30: True stress-true strain plots for Med-Austenite-LC TRIP steel (14% γ) at different strain rates. Each curve pair is offset by 0.10 strain for clarity.

Figure 5.31: Variation in yield stress and ultimate tensile stress of Med-Austenite-LC TRIP steel (14% γ) with strain rates.

Figure 5.32: True stress-true strain plots for the High-austenite-LC TRIP steel for different strain rates of testing. The curves are offset by a strain of 0.10 for ease of understanding.

Figure 5.33: Variation in yield stress and ultimate tensile stress of High-Austenite-LC TRIP steel with strain rates.

Figure 5.34: Variation in yield stress of three Si-bearing TRIP steels with different retained austenite content as a function strain rate.

Figure 5.35: Variation in ultimate tensile stress of three Si-bearing TRIP steels with different retained austenite content as a function strain rate.

Figures 5.34 and 5.35 summarize the yield stress and ultimate tensile values as a function of strain rate for the three Si-bearing TRIP steels with microstructures having different retained austenite contents. It is clear from the figures that the Med-austenite-

LC and High-austenite-LC steels show identical strain rate sensitivity. The steel with a low austenite fraction shows a lesser degree of strain rate sensitivity. Interestingly, the Low-austenite-LC steel also revealed yield point elongation during testing at all strain rates. The strain rate sensitivity of the yield stress for the Low-austenite-LC steel resembles that of the BH steel. The ultimate tensile strength for any of the TRIP steels does not show any strain rate sensitivity till at least a strain rate of 10s^{-1} . As discussed in Appendix C the heat treatments used to produce the different austenite volume fractions also resulted in different ferrite grain sizes.

Influence of Austenite Carbon Content

The intent of the current research was to investigate the effect of austenite carbon content on the mechanical performance of TRIP steels. However, attempted processing to develop TRIP steels with different carbon contents in Si-bearing steels resulted in steels with only slightly different carbon contents, 1.06 wt.% C for Low-austenite-LC and 1.0-1.2 wt.% C for Low-austenite-HC steel as mentioned in Table 4.2 in Chapter 4. The quasi-static stress-strain results for Low-austenite-LC and Low-austenite-HC steels are given in Figure 5.36. A comparison of the two steels reveal that Low-austenite-HC steel has a lower degree of strain hardening than the Low-austenite-LC steel and also a higher total elongation indicating that the retained austenite in Low-austenite-HC steel is more stable than the retained austenite in Low-austenite-LC steel. Both steels have similar yield strengths and both exhibit yield point elongation behavior, which is expected since the two materials have similar ferritic grain sizes. Figure 5.37 shows the true stress-strain plots of the Low-austenite-HC TRIP steel as a function of strain rate. The stress-strain plots at each strain rate reveal significant yield drop.

Figure 5.36: Engineering stress-strain curves for the TRIP steels with different retained austenite C-content but with similar austenite volume fraction tested at a strain rate of 0.001 s^{-1} .

Figure 5.37: True stress-true strain plots for the Low-austenite-HC TRIP steel for different strain rates of testing. The curves are offset by a strain of 0.10 for ease of understanding. Duplicate curves are shown superimposed for several of the test conditions.

The yield stress values of the Low-austenite-LC and Low-austenite-HC TRIP steels are compared in Figure 5.38 as a function of strain rate. Both the steels manifest almost similar strain rate sensitivity. The similar strain rate sensitivity of the two steels is probably due to the similar yielding behavior of the two steels. It is also observed that the austenite C-content difference between the two steels may not be large enough to manifest the influence of austenite C-content on the strain rate sensitivity behavior.

Figure 5.38: A comparison of the effects of strain rate on the yield stress of the Low-austenite-LC and Low-austenite-HC steels.

Table 4.2 reveals that the Al-containing TRIP microstructure developed by heat treatment is equivalent to Si-bearing Med-austenite-LC TRIP steel in terms of austenite volume fraction but has different austenite C-contents. Al-bearing TRIP microstructure have significantly higher austenite C-contents than the Si-bearing Med-austenite-LC steel. This presents an opportunity to examine the influence of austenite C-content on the strain rate sensitivity of TRIP steels for two microstructures with similar austenite volume fractions. The Al-bearing TRIP microstructure is designated as Med-austenite-HC TRIP steel in contrast to the Si-bearing Med-austenite-LC steel according to the higher austenite carbon content. Figure 5.39 shows the true stress-strain plots for the Med-austenite-HC Al-TRIP steel as a function of strain rate.

It is interesting to note that the Med-austenite-HC steel shows a similar yielding behavior at all strain rates similar to that of Low-austenite-HC Si-bearing TRIP steel shown in Figure 5.37. At each strain rate of testing, a large yield drop has been observed and the magnitude of yield drop increased with increase in strain rate. The strain hardening behavior remains same at all strain rates.

Figure 5.39: True stress-true strain plots of Med-austenite-HC Al TRIP steel at different strain rates. Each stress-strain plot is offset by an amount of 0.10 strain for visual clarity and several duplicate tests are shown superimposed.

Figure 5.40: Variation in yield stress and ultimate tensile stress of Med-austenite-HC TRIP steel with strain rates.

Figure 5.40 presents the effects of strain rate on the YS and UTS of Med-austenite-HC steel and shows that the effects of strain rate are similar on both parameters. Figure 5.41 compares the yield stress data of the Al-bearing Med-austenite-HC TRIP steel and the Si-bearing Med-austenite-LC steel as a function of strain rate. The results reveal a significant difference in the strain rate sensitivity of the two TRIP steels. The Si-bearing Med-austenite-LC steel shows significantly higher strain rate sensitivity than the Al-bearing Med-austenite-HC TRIP steel. Though the retained austenite fractions are similar in both the steels, the austenite carbon contents are different. The results suggest that a high austenite C-content stabilizes the austenite against deformation-induced transformation and also diminishes the strain rate sensitivity of the multi-phase microstructure.

Figure 5.41: Comparison of yield stress values as a function of strain rates for an Al-bearing Med-austenite-HC TRIP steel and a Si-bearing Med-austenite-LC TRIP steel.

Figure 5.42a shows an additional comparison of the strain rate dependent yield stress of all the TRIP microstructures tested. It is seen that Al-bearing Med-austenite-HC steel shows the lowest strain rate sensitivity of all the TRIP microstructures examined. The strain rate sensitivity value (β) measured up to a strain rate of 10 s^{-1} are summarized in Figure 5.42b and clarify this observation.

Figure 5.42: (a) Comparison of yield stress values as a function of strain rates for the Al-bearing Med-austenite-HC TRIP steel with other Si-bearing TRIP steels and (b) strain rate sensitivity values measured up to a strain rate of 10 s^{-1} for different TRIP microstructures produced in the present test program.

In summary, austenite volume fraction does not seem to influence the strain rate sensitivity of TRIP steels as seen from the results of Figure 5.34. However, an increase in austenite C-content or an increase in austenite stability changes the yielding behavior of the ferrite and significantly reduces the strain rate sensitivity of the TRIP steels. Since an Al addition has a pronounced influence on the increase of austenite stability through increase in austenite carbon content during bainitic transformation, it is expected that an Al addition may reduce the strain rate sensitivity of such steels. A detail research work on the study of the influence of Al and Si on the austenite stability and the austenite transformation behavior during deformation has been described in Appendix I.

6.0 Analysis of Constitutive Equations to Describe the Effects of Strain Rate on the Strength of Sheet Steels.

6.1 Introduction to Application of Constitutive Equations

Analysis of mechanical property data to obtain “universally”-applicable constitutive equations to describe mechanical property data take the following general form:

$$\sigma_F = f(\epsilon, \dot{\epsilon}, T, \text{microstructure}, \dots) \quad [6.1]$$

where σ_F is the flow stress at a specified strain, ϵ is the true strain, $\dot{\epsilon}$ is the true strain rate, T is the absolute temperature, and the microstructure is defined by appropriate parameters including grain size, second phase volume fraction size, distribution, and volume fraction, etc. With deformation, there may also be complex interrelationships between the parameters. For example, plastic strain under adiabatic conditions leads to an increase in test temperature with strain, and in TRIP steels, deformation can lead to martensite formation and a corresponding change in the relative volume fractions of the microstructural constituents.

Development of constitutive equations for advanced multi-phase high strength sheet steels requires three parts: first, constitutive equations to describe the deformation behavior of individual phases; second, an appropriate composite model to describe load distribution between phases and interactions between phases; and finally, for those materials that exhibit microstructural or temperature changes with strain, equations that describe the strain-dependence of these parameters.

Approaches to the description of the deformation behavior of individual phases generally fall into the following categories: (a) fundamental models based on micromechanisms of dislocation motion, (b) phenomenological models designed to develop general equation forms that describe overall stress strain curve shapes, and (c) development of parametric equations based on multi-term polynomial curve-fitting where multiple adjustable parameters are empirically determined to describe experimental data sets. In the development of these models, it is generally assumed

that the primary deformation mechanism responsible for strain hardening remains constant for the range in test variables for which the equations were developed. In the assessment of the deformation behavior of materials as a function of strain rate, it has been shown that deformation primarily by dislocation motion is augmented by strain due to deformation twinning at very high strain rates, e.g. $\geq 10^4 \text{ s}^{-1}$. However, for the range of strain rates of primary interest in this study on sheet steels, i.e. 10^{-3} to $5 \times 10^2 \text{ s}^{-1}$ [28], it is anticipated that contributions due to deformation twinning will be limited, and thus the assessments included here consider primarily models based on single deformation mechanisms, i.e. dislocation motion.

Sheet steels with multi-phase microstructures can be described based on composite models, and the application of rule-of-mixtures approach has been successfully applied to describe the strength of dual-phase steels [29] and microstructural changes associated with strain-dependent transformation of austenite to martensite in steels with metastable austenite [30]. Of particular note is the composite model of Mileiko [31], and applied later by Garmong and Thomson, that has been used to describe the stress strain behavior and instability conditions in a multi-phase, iso-strain composite [32]. In the following sections these concepts are used to describe the deformation behavior of DP and TRIP steels.

6.2 Analysis of the Effects of Strain Rate and Microstructure on the Deformation Behavior of Ferrite

A review of constitutive equation models is presented in Section 3.13 of D. Bruce's thesis [Appendix B]. Based on the results of this review, she developed constitutive equations to describe the effects of cold work, solid solution strengthening and grain refinement in ferrite (i.e. in an IF steel). Her model, based on the Zerilli-Armstrong equation for BCC materials [33], was shown to systematically predict the overall strain hardening behavior as a function of strain rate and microstructural variable and to describe the deformation behavior of HSLA steels. Figures 6.1 and 6.2 illustrate the correlation between model predictions and experimental data. These figures illustrate that the model describes well the overall deformation behavior but does not adequately describe the effects of strain rate on the yield point phenomena. This latter short coming is a characteristic of most all of the models currently being considered in various publications to describe the strain rate dependence of the deformation behavior of sheet steels. If the yield point phenomena are to be incorporated into models, then a two-step model that addresses the difference between discontinuous yielding associated with yield points and continuous yielding associated with uniform dislocation-dislocation interactions during strain hardening must be developed.

Figure 6.1: True stress-strain curves for IF-1 steel at three different strain rates compared with Modified Zerilli-Armstrong model at same strain rates using the constants summarized in Table 6.12 of Appendix B. (a) 10 μm grain size, (b) 25 μm grain size, and (c) 135 μm grain size.

Figure 6.2: True stress-strain curves for (a) HSLA-1 steel and (b) HSLA-2 steel at four different strain rates compared with Modified Zerilli-Armstrong model at same strain rates using the constants shown in Table 6.14 of Appendix B .

6.2 Constitutive Equations Applicable to Dual-Phase Steels

The effects of strain rate on the deformation behavior of multi-phase materials can be modeled by applying an appropriate composite model along with appropriate constitutive equations for the individual phases. Following Mileiko [31] and assuming that a rule of mixtures applies to the deformation behavior of dual phase steels at all strain rates, the effects of strain rate on the martensite volume fraction dependent flow stress and on the effects of martensite on the apparent strain rate sensitivity can be predicted. Neglecting the effects of temperature change with strain rate, the flow stress can be given by the following:

$$\sigma_F(\text{Dual Phase Steel}) = V_M \sigma_M(\epsilon, \dot{\epsilon}) + (1 - V_M) \sigma_a(\epsilon, \dot{\epsilon}) \quad [6.2]$$

where V_M is the volume fraction of martensite, σ_F is the dual phase steel flow stress at a specified strain and strain rate, and σ_M and σ_a represent the strain and strain rate dependent constitutive equations for martensite and ferrite respectively. As shown by Kircher and as illustrated in Figure 5.24 (from Figure 5.7 in Appendix C), for a fully martensitic low carbon steel, over the strain rate range of interest to this study, the deformation behavior is essentially independent of strain rate. With this assumption, the strain rate dependence of the dual phase steel flow stress depends on the strain rate dependence of the ferrite present. The significance of this assumption is illustrated in Figure 6.3 which compares the strain rate dependence of the yield stress in dual phase steels with V_F amounts of 13, 28, 39, and 49 pct. (obtained from Fig. 5.8 a of Appendix C) with predictions based on Eq. 6.2. For these calculations, the effects of strain rate on the yield stress of ferrite were taken from the work of Bruce [Appendix B] and from Fig. 5.24 the yield stress of low carbon martensite was taken as 1100 MPa. The figure shows that the primary effects of strain rate and martensite volume fraction, i.e. that the

effect of strain rate on flow stress decreases with an increase in martensite volume fraction and strength increases with martensite volume fraction, are explained by this simple model. Improvements to the correlations may be obtained with the development of additional data on ferrite and martensite with compositions designed to reflect the specific compositions in the phases of the composite of interest and by incorporating flow curves for each phase in Mileiko's model. However, these results clearly illustrate the degree to which the strain rate sensitivity of the ferrite phase dominates the strain rate behavior of the composite. Thus it is concluded that with the composite model discussed here and the constitutive model for ferrite developed by Bruce, the effects of strain rate on the deformation behavior of DP steels can be described.

Figure 6.3: A comparison of predicted (solid and dashed lines) and measured yield stresses (experimental data points) in a series of dual phase steels with martensite volume fractions between 13 and 49 pct. as a function of strain rate.

6.2 Constitutive Equations Applicable to TRIP Steels

The composite approach of Olson [30], which is actually an extension of the concepts of Mileiko [31] to a material with a metastable components of the microstructure, is applicable as a constitutive model for the description of TRIP steels. These approaches are applicable if constitutive equations for the individual phases are

know, along with a description of microstructural changes with strain. The samples developed for this study to compare and assess systematic variations in retained austenite volume fraction and austenite stability, contained variations in other microstructural variables (e.g. ferrite grain size) that contributed to strength differences in addition to the effects of austenite. An assessment of these variables is contained in Kricher's thesis [Appendix C]. These comparisons showed that all of the TRIP steels exhibited similar sensitivities to strain rate except for alloys A (0.1 C, 1.52 Mn) and D (0.21 C, 1.5 Mn, 1.0 Al) heat treated to have a higher carbon content in the austenite. This material exhibited a slightly higher sensitivity of yield stress to strain rate, but this sample also exhibited the largest ferrite grain size.

7.0 Summary

An experimental test facility to evaluate the effects of strain rate on tensile properties of Advanced High Strength Sheet Steels (AHSS) was established and used to evaluate the effects of strain rate on steels with systematic microstructural variations designed to assess the dominant strengthening mechanisms in these steels. The results show that all materials exhibited an increase in strength with strain rate and the functionality of this increase depended on the effects of strain rate on the properties of the ferritic phase that dominated the microstructures in all of these steels. Constitutive models were considered and developed to describe the effects of microstructural variables on ferrite and on the composite structures of ferrite and either martensite (DP steels) or austenite (TRIP steels). The results of this study are the subject of ongoing further analysis and investigation. As additional information is developed based on the results obtained in this program, these results will be provided to all participants in this study.

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9.0 List of Appendices

Appendix A: George Krauss, Amar K. De, John G. Speer and David K. Matlock, "The Mechanical and Physical Metallurgy of Low-Carbon Sheet Steels Produced by Dual-Phase and TRIP Steel Processing," Special Review Report, MT-SRC-02-007, Colorado School of Mines, Golden, Colorado, April 2002.

Appendix B: Denise M. Bruce, "Dynamic Tensile Testing of Sheet Steels and Influence of Strain Rate on Strengthening Mechanisms in Sheet Steels", MT-SRC-003-018, Ph.D. Thesis, Colorado School of Mines, Golden, Colorado, July 2003.

Appendix C: Ryan S. Kircher, "The Effect of Strain Rate on the Tensile Properties of High Strength Multiphase Sheet Steels," M.S. Thesis, Colorado School of Mines, Golden, Colorado, Draft Version, Final thesis to be published April 2005.

Appendix D: Amar K. De, John G. Speer and David K. Matlock, "Color Tint-Etching for Characterization of Complex Microstructures in Multiphase Steels," Advanced Materials and Processes, vol. 161, no. 2, 2003, pp. 27-30.

Appendix E: D. M. Bruce, D.K. Matlock, J. G. Speer, A.K. De, "Assessment of the Strain-Rate Dependent Tensile Properties of Automotive Sheet Steels", SAE Technical Publication #2004-01-0507, SAE, Warrendale, PA, 2004; Also published in Innovations in Steel Sheet and Bar Products and Processing and Modeling and Testing of Steel Structures, ed. by D.W. Anderson, *et al.*, SAE Publication SP-1837, SAE Warrendale, PA, 2004, pp. 119-128.

Appendix F: D.K. Matlock, D.M. Bruce, and J.G. Speer, "Strengthening Mechanisms and Their Applications in Extremely Low C Steels", International Forum for the Properties and Applications of IF Steels, ed. by H. Takechi, Proceedings of IF Steels 2003, Iron and Steel Institute of Japan, Tokyo, Japan, 2003, pp. 118-127.

Appendix G: I.D. Choi, D.M. Bruce, S.J. Kim, C.G. Lee, S.H. Park, D.K. Matlock, and J.G. Speer, "Deformation Behavior of Low Carbon TRIP Sheet Steels at High Strain Rates", Proc. of 1st ICASS (International Conference for Advanced Structural Steels), Tzukuba, Japan, May 21-24, 2002, pp. 85-89; also published in ISIJ International, vol. 42, No. 12, 2002, pp. 1483-1489.

Appendix H: I.D. Choi, S.J. Kim, D. Bruce, D.K. Matlock, and J.G. Speer, "The Effect of Retained Austenite Stability on High Speed Deformation Behavior of TRIP Steels", Proceedings of ICASS 2004, Chinese Society for Metals, Shanghai, China, April 14-16, 2004, pp. 703-706.

Appendix I: Amar K. De, Ryan S. Kircher, John G. Speer and David K. Matlock, "Transformation Behavior of Retained Austenite in TRIP Steels as Revealed by a Specialized Etching Technique," Proc. Int. Conf. on Advanced High-Strength Sheet

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Appendix J: R.S. Kircher, A.K. De, D.K. Matlock, and J.G. Speer, "The Effects of Strain Rate on the Deformation Behavior of High Strength IF and Bake-Hardenable Sheet Steels for Automotive Applications," Materials Science & Technology 2004 Conference Proceedings, AIST and TMS, Warrendale, PA, 2004, vol. I, pp. 491-498.

Appendix K: Paper Drafts currently under preparation

[1] Amar K. De, John G. Speer and David K. Matlock, "Influence of Austenite Carbon Content on the Deformation-induced Transformation Response in Low Alloy TRIP Steels", paper to be submitted to Acta Materialia.

[2] I.D. Choi, D.M. Kim, S.J. Kim, D. Bruce, D.K. Matlock, and J.G. Speer, "The Effect of Retained Austenite Stability on High Speed Deformation Behavior of TRIP Steels".

[3] I.D. Choi, D.M. Bruce, D.K. Matlock, J.G. Speer, D.M. Kim, and S.H. Park, "The High Speed Deformation Behavior of Automotive Steels".

[4] I.D. Choi, D.M. Kim, D.M. Bruce, D.K. Matlock, J.G. Speer, and S.H. Park, "Effect of Prestrain and Bake Hardening Treatment on the High Speed Deformation Behavior of Automotive Steels", to be submitted to J. of Korea Inst. Met. & Mater. 2005.