Quenching and Partitioning Response of a Si-Added TRIP Sheet Steel

A.M. Streicher,[1] J.G. Speer,[1] D.K. Matlock,[1] and B.C. De Cooman[2]

[1] Advanced Steel Processing and Products Research Center;
Colorado School of Mines, Golden, CO 80401, USA
Tel.: 303-273-3025
Fax: 303-272-3016
E-mail: astreich@mines.edu, jspeer@mines.edu, dmatlock@mines.edu

[2]Laboratory for Iron and Steelmaking, Department of Metallurgy and Materials Science;
Gent University, Technologiepark 903, B-9052, Gent, Belgium
Tel.: 32-9-264-5761
Fax: 32-9-264-5833
E-mail: bruno.decooman@ugent.be

ABSTRACT

Mechanical property results are presented for sub-sized sheet tensile samples of a 0.19C-1.59Mn-1.63Si TRIP steel composition, heat-treated by a recently developed process, quenching and partitioning (Q&P). The Q&P process was proposed to produce steels with controlled volume fractions of carbon-enriched retained austenite at room temperature. For Q&P processing, sheet steel samples were heated to form austenite (either completely or partially), followed by quenching to a temperature between M_s and M_f to create controlled amounts of martensite, followed by a thermal treatment to partition carbon into the remaining austenite. All processing occurred isothermally in molten salt or in low-melting point metallic baths. The ability to create new sheet microstructures using Q&P is investigated, and the mechanical property results are compared to data for other high strength sheet products. The results suggest that Q&P is able to extend the strength of TRIP steels to higher levels than obtained by conventional processing.

INTRODUCTION

A novel process, quenching and partitioning (Q&P), has been developed recently based upon a new understanding of carbon partitioning from martensite into austenite after quenching[1]. The goal of Q&P is to produce specific volume fractions of carbon-enriched retained austenite[1-3]. While the thermodynamic calculation details are not presented here, in the absence of carbide formation it is found that the austenite composition can be closely approximated by assuming that virtually all of the carbon in the martensite partitions to the austenite, and applying the appropriate carbon mass balance based on the amount of retained austenite present after quenching[1]. An implication of these carbon balance considerations is that there is a tradeoff between the amount of austenite present and the degree of carbon enrichment that is possible in a given steel, although higher overall carbon levels in the steel provide the potential for greater amounts of higher carbon austenite.

The Q&P processing schematic is shown in Figure 1. Corresponding schematic microstructures are also shown for each processing step in Figure 1. The example shown indicates an initial full austenitization step, although both fully austenitic and intercritical initial microstructures, prior to "quenching" and "partitioning" treatments, may be employed using appropriate processing parameters. Carbon enrichment of austenite occurs by partitioning from carbon-supersaturated martensite. Martensite volume fraction is controlled by cooling to a specific quench temperature between M_s and M_f. The quench temperature, designated QT in Figure 1, controls the martensite fraction according to, for example, the Koistinen-Marburger[4] relationship. The M_s temperature can be estimated from available equations such as the Andrews equations, which incorporates the effects of alloy additions such as C, Mn, Cr, Ni, and Mo[5]. A similar equation[6] that includes the effects of Si and Al is also available, and was used in this work to determine M_s temperatures. Subsequent holding of the austenite/martensite mixtures at appropriate partitioning temperatures (PT), which may or may not differ from the initial martensite quench temperature, should result in austenite with increased stability (through increased austenite carbon content) and carbon-depleted martensite. Since complete partitioning assumes no loss of solute carbon to competing reaction products, such as transition carbides or cementite, microconstituents that effectively consume carbon should be avoided. If carbide precipitation is avoided, the level of carbon

enrichment of the austenite when partitioning is completed can be approximated by assuming that virtually all of the carbon partitions to the austenite[1,3].
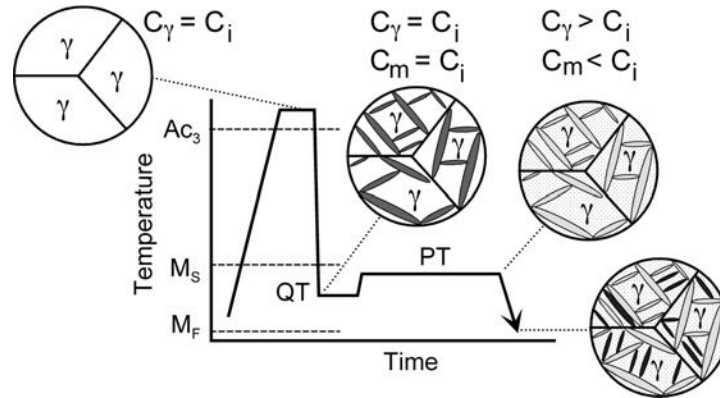

Figure 1.   Schematic illustration of the Q&P process for producing of austenite-containing microstructures. C_i, C_γ, C_m represent the carbon contents of the initial alloy, austenite, and martensite, respectively. QT and PT are the quenching and partitioning temperatures.

The absence of carbide formation is a fundamental element of the CPE model, since the existence of metastable equilibrium between ferrite and austenite is precluded if the more stable ferrite plus iron carbide equilibrium is achieved[1,3]. Any carbide formation effectively "consumes" carbon, since it is no longer available to enrich the austenite. Thus, it is necessary to understand and control carbide precipitation processes that may occur during any partitioning treatments associated with the Q&P process. It is well known in the bainite transformation literature that cementite formation can be eliminated or suppressed through additions of silicon[7,8], and also that aluminum and even phosphorus can have this effect[9]. TRIP sheet steels are potential candidates for Q&P processing, since alloying elements such as Si and Al are already added to prevent carbide formation. Q&P has therefore been proposed as an alternative processing approach to form high strength TRIP sheet steels, with the added benefit of enhanced microstructure control including the final stability (through carbon content) of the retained austenite. It is anticipated that Q&P processing may yield higher strength TRIP sheet steels with martensite/austenite mixtures, and reduced equiaxed ferrite fractions. In this paper, mechanical property results are presented for Q&P heat-treated sub-sized tensile samples with varying amounts of intercritical ferrite to assess the ability to create a new family of steels extending the strength levels of existing TRIP products.

EXPERIMENTAL

Sub-sized tensile samples of the TRIP steel composition listed in Table I were Q&P heat-treated and tested. Intercritical and austenitizing treatments were initially performed on heat-treat coupons for 180s in molten salt to determine appropriate temperatures for obtaining specific fractions of intercritical ferrite and austenite. From metallography, the Si-containing TRIP steel was determined to be fully austenitized above 900°C, while an intercritical treatment of 820°C produced approximately 75% austenite with 25% ferrite. Consequently, austenitizing or intercritical temperatures of 950°C or 820°C, respectively, were used to produce material containing either 0% or 25% equiaxed ferrite.

TABLE I - Chemical Composition of Experimental High-Si TRIP Sheet Steel (wt. pct.)

C	Mn	Si	Al	P	N	Cr	S
0.19	1.59	1.63	0.036	0.013	0.0109	0.03	0.002

Austenite volume fractions were determined using x-ray diffraction (XRD) to assess retained austenite amounts in the Q&P heat treated condition. Samples for XRD were obtained from heat-treated sub-sized tensile sample grip ends, located just before the fillet and reduced section. It was assumed that sample plastic deformation did not occur in the grip section of the specimen during mechanical testing. Specimens were ground to mid-thickness to remove near-surface microstructures and polished for XRD. Molybdenum radiation was used and eight diffraction peaks, including four austenite and four ferrite peaks, were monitored. Samples were scanned continuously over a 2θ range from 18° to 45° using a step size of 0.02° and a dwell time of 1.2s. A Siemens D5000 Kristalloflex diffractometer operating at 45kV and 50mA was used. The intensity and position of each peak was determined using peak fitting software. A relationship incorporating the integrated intensities of the {111}, {200}, {220}, and {311} austenite peaks and the {110}, {200}, {211}, and {220} ferrite peaks, in addition to R-values for each peak, was used to quantify austenite volume fraction for the steel used in this study[10]. This approach is reported to account well for possible texture effects, since numerous austenite and ferrite peaks are included in the calculation[10]. R-values are a function of θ, hkl, and the crystal structure and composition of each phase, and are representative of calculated theoretical intensity values[10,11]. 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[11]. Ferrite and martensite were both treated as body centered cubic for these calculations. The $K_{\alpha 2}$ contribution to intensity was stripped for retained austenite determination,

and integrated intensity incorporated only the contribution of $K_{\alpha 1}$. Austenite carbon content was estimated from the austenite {220} peak position (determined from peak fitting software) using the expression[11]:

$$a_0 = 3.555 + 0.044x \tag{1}$$

where a_0 is austenite lattice parameter in Angstroms and x is carbon content in weight percent. Carbon content was estimated from the {220} lattice parameter, since peak position could be most accurately determined for this high angle peak. The $K_{\alpha\ wtd.}$ wavelength (a weighted average) was used to calculate {220} austenite peak position; the $K_{\alpha 2}$ contribution was not stripped for austenite carbon content determination so that more accurate peak positions, especially of lower intensity peaks, could be obtained.

RESULTS AND DISCUSSION

Fully austenitized microstructure (2-step Q&P processing)
Figure 2 shows the *final predicted* austenite fraction after partitioning and cooling to room temperature (solid line) vs. the quenching temperature prior to partitioning (after full austenitization). The calculation essentially applies the Koistinen-Marburger[4] relationship to both the initial quench, and then the final quenching step to room temperature (~25°C) after full partitioning. The results indicate an "optimum" quenching temperature that yields a maximum amount of retained austenite.[3] The microstructural components include the austenite and martensite present at the quench temperature (after the initial quench), and the additional martensite that forms during the final quench to room temperature. Above the peak temperature, substantial austenite fractions remain after the initial quenching step, but the austenite stability is too low during final quenching, and increasing amounts of fresh martensite (M_{FRESH}) are found at higher quench temperatures, reducing the final austenite fraction at room temperature. Below the peak temperature too much austenite is consumed during the initial quench prior to carbon partitioning, and the carbon content of the retained austenite is greater than needed for stabilization at room temperature. The peak is found at the particular quench temperature where martensite formation is just precluded during the final quench (i.e. where M_{FRESH} vanishes). This temperature corresponds to austenite having an M_s temperature of room temperature after full partitioning. The calculations made using this methodology were used to guide the selection of quench temperatures to achieve substantial austenite fractions. Two-step Q&P heat treatments, with partitioning at temperatures above the quench temperature[2], were envisioned for study. The heat-treatments on sub-sized tension specimens were to consist of austenitizing for 180s at 950°C, followed by quenching to 220, 240, and 260°C and holding for 3s, followed by partitioning at 350, 400, or 450°C for 10, 30, or 100s. Molten salt was used for heat-treatment stages above 350°C, and a tin-bismuth bath was used to achieve lower processing temperatures. All samples were water-quenched to room temperature after partitioning. Due to salt pot failures at high temperature, only samples processed by austenitizing for 180s at 950°C, followed by quenching to 220°C for 3s, and partitioning at 350, 400, or 450°C for 10, 30, or 100s are reported here. Duplicate sub-sized tensile specimens heat-treated in this sequence were then tensile tested at a displacement rate of 0.1 in/min.

Table II summarizes average mechanical property results from duplicate samples and austenite results obtained for the different partitioning temperatures and times examined for the fully austenitized condition. Uniform elongation, independent of gauge length, is also reported in Table II in addition to total elongation because sub-sized tensile samples with 25.4mm gauge lengths were tested. Uniform elongations were determined from peak loads obtained during the tensile test; elastic contributions were subtracted from the strain at the ultimate tensile strength, calculated from peak load. Figure 3 shows engineering stress vs. engineering strain curves for single samples austenitized at 950°C for 180s, quenched to 220°C and held for 3s, then partitioned at 350°C for 10, 30, or 100s. The stress-strain behavior shown in Figure 3 generally depicts the behavior of these materials, and indicates very high strength levels.
As shown in the results of Figure 3 and Table II, very high yield and ultimate tensile strengths were achieved with limited ductility, which was anticipated for these microstructures containing significant amounts of martensite. The retained austenite results suggest that the quench temperature (220°C) in this instance may still not be optimized. Austenite carbon content is not reported in Table II for these samples, due to limitations in accurately determining peak position for small intensities, although carbon contents over 1% are estimated. Q&P processing involving quenching to 240°C, expected to produce the maximum amount of final austenite after partitioning and cooling to room temperature, was not performed due to salt pot limitations, as previously mentioned. Nonetheless, the 220°C/350°C,10s Q&P specimen appears simultaneously to exhibit high strength and elongation relative to the other specimens, indicating the likely occurrence of a favorable microstructure.
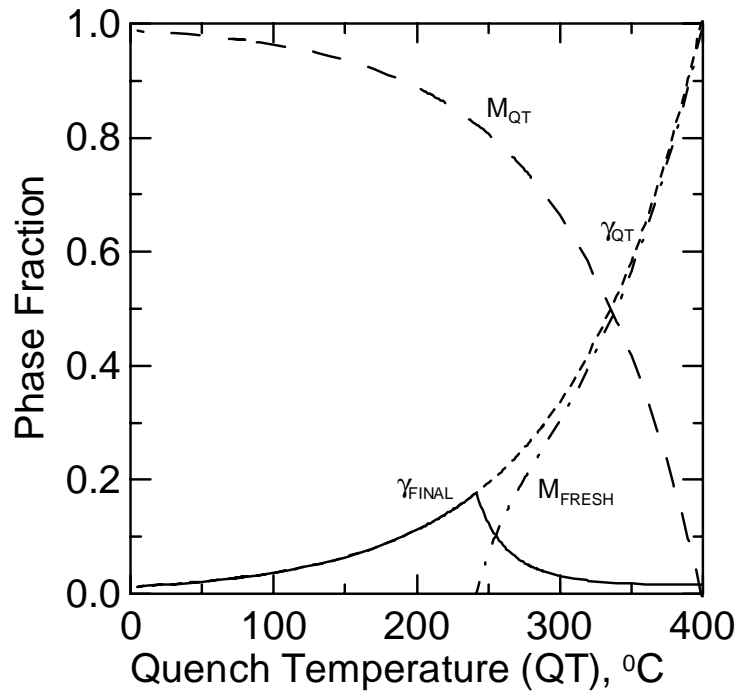
Figure 2.   Final austenite fraction vs. quench temperature for a Si-containing TRIP steel composition after full austenitization.
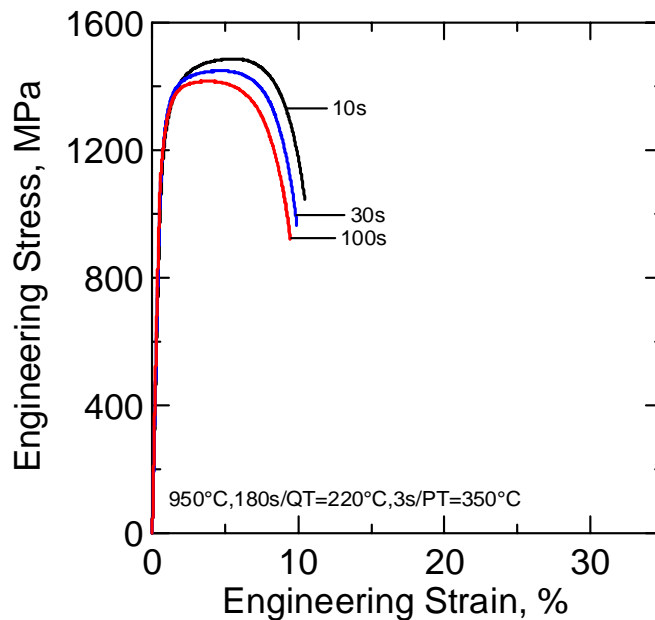

Figure 3.   Engineering stress vs. engineering strain for sub-sized tensile samples austenitized at 950°C for 180s, quenched to 220°C for 3s, and partitioned at 350°C for 10, 30, or 100s (2-step).

Intercritical microstructure (1-step and 2-step Q&P processing)

Due to the high strength levels achieved after austenitization at 950°C followed by Q&P processing, samples were heat-treated to produce lower strength microstructures containing intercritical ferrite (in addition to martensite and retained austenite) for improved ductility. Figure 4 shows predicted final austenite fraction after intercritical annealing to produce approximately 75% intercritical austenite and 25% intercritical ferrite. Q&P heat-treatments of sub-sized tensile samples were performed based upon the guidance provided in Figure 4. As shown, a quench temperature of approximately 240°C is expected to produce the maximum amount of final austenite. Samples were intercritically annealed at 820°C for 180s, quenched to 200, 220, 240, or 260°C, and partitioned for 10, 30, 100, or 1000s. Several quench temperatures were investigated to "bracket" the estimated optimum quench temperature and to assess sensitivity to quench temperature.

TABLE II – Average Tensile Test Results of Sub-sized Samples after 2-step Q&P Heat Treatment Following Full Austenitization at 950°C (0.19%C-1.59%Mn -1.63%Si TRIP Sheet Steel).

Heat Treatment	0.2% Yield (MPa)	UTS (MPa)	Uniform Elongation (%)	Total Elongation (%)	Retained Austenite (%)
950°C,180s/220°C,3s/350°C,10s	1201	1483	4.7	9.0	2.6
950°C,180s/220°C,3s/350°C,30s	1258	1446	3.7	7.9	3.1
950°C,180s/220°C,3s/350°C,100s	1232	1424	3.3	6.9	2.7
950°C,180s/220°C,3s/400°C,10s	1182	1388	4.5	7.9	3.1
950°C,180s/220°C,3s/400°C,30s	1224	1380	2.5	6.0	3.0
950°C,180s/220°C,3s/400°C,100s	1223	1368	2.7	5.9	2.2
950°C,180s/220°C,3s/450°C,10s	1247	1366	2.6	7.9	2.9
950°C,180s/220°C,3s/450°C,30s	1219	1309	2.9	7.0	2.6
950°C,180s/220°C,3s/450°C,100s	1165	1248	5.2	7.4	1.0
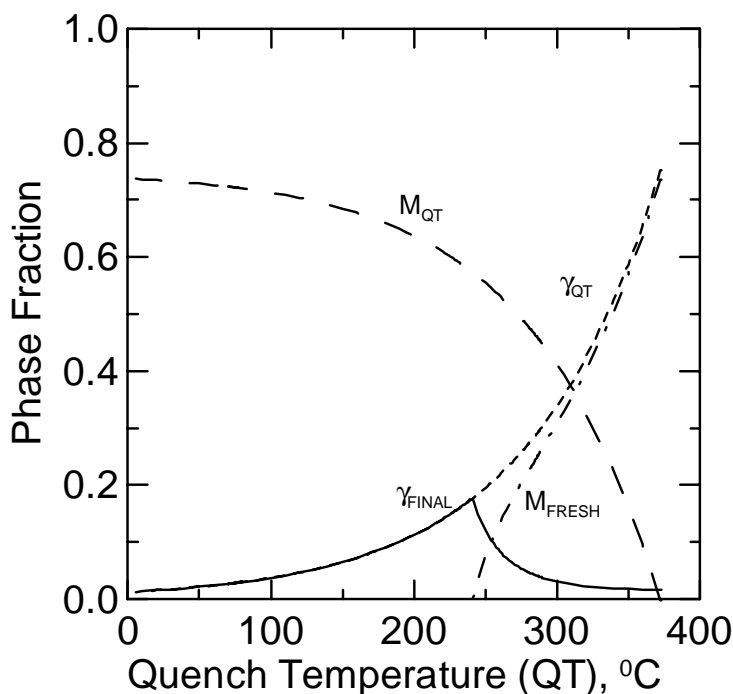
Figure 4.   Final austenite fraction vs. quench temperature for a Si-containing TRIP steel composition after intercritical annealing to produce approximately 75% intercritical austenite and 25% intercritical ferrite.

Table III summarizes tensile test results after "1-step" Q&P processing (where the quenching and partitioning temperatures are identical[2]). One-step Q&P processing consisted of intercritical annealing at 820°C for 180s, quenching to 200, 220, 240, or 260°C, followed by partitioning at the quench temperature for 10, 30, 100, 100, or 1000s. Austenite peak intensities were limited for 1-step heat-treated samples, and austenite amounts were determined to be less than 3%. Heat-treatments of 820°C/240°C,1000s and 820°C/260°C,1000s qualitatively seemed to produce slightly more residual austenite (~3%), with carbon content estimated to be 0.9%. Longer partitioning times at these higher quench temperatures could be of interest with respect to future 1-step Q&P processing studies.

Mechanical properties and austenite volume fraction and carbon content results are shown in Table IV for 2-step processing, with partitioning at temperatures above the quench temperature. The 2-step heat-treatment consisted of intercritical annealing at 820°C for 180s, quenching to 200, 220, 240, or 260°C for 10s, followed by partitioning at 400°C for 10, 30, 100, or 1000s. A holding time of 10s in the quenchant was used to ensure thermal equilibration at the quench temperature. Average mechanical property results for duplicate samples are provided.

TABLE III – Average Tensile Test Results of Sub-sized Samples after Intercritical Annealing at 820°C and 1-step Q&P Heat-Treatment (0.19%C-1.59%Mn -1.63%Si TRIP Sheet Steel).

Heat Treatment	0.2% Yield (MPa)	UTS (MPa)	Uniform Elongation (%)	Total Elongation (%)
820°C,180s/200°C,10s	740	1424	8.5	10.9
820°C,180s/200°C,30s	687	1408	8.7	13.1
820°C,180s/200°C,100s	724	1401	7.9	10.8
820°C,180s/200°C,1000s	636	1335	9.5	11.9
820°C,180s/220°C,10s	733	1432	8.4	10.8
820°C,180s/220°C,30s	719	1424	7.4	11.6
820°C,180s/220°C,100s	711	1390	8.5	11.4
820°C,180s/220°C,1000s	632	1331	9.0	12.9
820°C,180s/240°C,10s	715	1385	8.4	12.7
820°C,180s/240°C,30s	750	1384	8.6	10.9
820°C,180s/240°C,100s	706	1371	8.6	13.3
820°C,180s/240°C,1000s	599	1304	10.0	12.1*
820°C,180s/260°C,10s	723	1393	7.9	12.6*
820°C,180s/260°C,30s	679	1369	8.9	11.3
820°C,180s/260°C,100s	691	1338	9.1	11.4
820°C,180s/260°C,1000s	584	1241	10.7	14.4

* total elongation reported for one sample

The goal of the Q&P processing after intercritical treatment was to achieve high strengths and improved ductilities in comparison to the results of Figure 3, through additions of equiaxed ferrite and additional retained austenite to the microstructure. Figure 5 shows representative tensile test results after 1-step processing (intercritical annealing at 820°C for 180s, quenching to 260°C, followed by partitioning for 10, 30, 100, or 1000s). Continuous-yielding stress-strain curves were obtained from 1-step processing, consistent with stress-strain behavior typically exhibited by martensitic microstructures. Additionally, Figure 5 shows that partitioning times of 1000s generally resulted in lower ultimate tensile strengths and greater ductility compared to those shown for samples partitioned at the same quench temperature for shorter times.

Representative tensile test results are shown in Figure 6 for sub-sized samples 2-step Q&P heat-treating with higher partitioning temperatures. Samples were intercritically annealed for 180s at 820°C, quenched to 240°C and held for 10s, then partitioned at 400°C for 10, 30, 100, or 1000s. Lower tensile strengths were obtained for the 2-step processed samples than the 1-step processed samples. Longer partition times further decrease tensile strength for a given series, and generally increase ductility. A more pronounced transition from elastic to plastic deformation is exhibited in comparison to 1-step processed samples. Higher quench temperatures resulted in slightly greater elongations. Samples quenched to 240°C and 260°C show particularly promising results, consistent with the predicted maximum amount of retained austenite being at a quench temperature of about 240°C. The elongations from the 2-step results (Table IV) suggest that austenite might indeed be present at higher levels in samples quenched to temperatures near 240°C and partitioned at 400°C, and the x-ray measurements of austenite fraction confirm this behavior.

Figure 7 shows retained austenite fraction vs. partitioning time after intercritical annealing at 820°C, quenching to 200, 220, 240, or 260°C, followed by up-quenching and partitioning at 400°C for 10, 30, 100, or 1000s, and water-quenching to room temperature. The associated carbon concentrations in the retained austenite are plotted in Figure 8 vs. partition time. It should be noted that water quenching from 820°C (i.e. a condition that corresponds to a partitioning time of zero) resulted in no measurable amount of austenite. At long partitioning times, austenite volume fraction decreases, which might be attributed to increasing amounts of carbide formation, or complications associated with competition from concurrent bainite transformation (especially at increased quench temperatures, where greater amounts of untransformed austenite are present before partitioning). The carbon concentration of the austenite is shown in Figure 8, and generally increases with partitioning time for a given quench temperature, except after 1000s at a quench temperature of 200°C. The overall trend is to reach higher carbon levels with increasing partitioning time, consistent with carbon diffusion controlled processes. Carbide precipitation (possibly in combination with some bainite formation) may also occur during the partitioning treatment. Carbide formation would result in reduced enrichment of the austenite, while formation of significant amounts of carbide-free bainite might further enrich the remaining austenite at long times, which could stabilize austenite, but also result in reduced austenite fractions. Bainite formation could be avoided in higher

alloy steel compositions, and further work is currently underway to examine such steels. Small XRD peaks near 21° and between 31.3-31.4° 2θ (Mo-radiation), possibly corresponding to {112} and {123} cementite, exist.

TABLE IV – Average Tensile Test Results of Sub-sized Samples after Intercritical Annealing at 820°C and 2-step Q&P Heat-Treatment (0.19%C-1.59%Mn -1.63%Si TRIP Sheet Steel).

Heat Treatment	0.2% Yield (MPa)	UTS (MPa)	Uniform Elongation (%)	Total Elongation (%)	Retained Austenite (%)	Austenite Carbon Content (%)
820°C,180s/200°C,10s/400°C,10s	781	1179	9.9	12.8	8.4	1.2
820°C,180s/200°C,10s/400°C,30s	843	1157	9.3	14.3	7.6	1.3
820°C,180s/200°C,10s/400°C,100s**	848	1135	8.5	14.3	7.4	1.5
820°C,180s/200°C,10s/400°C,1000s	840	1097	9.9	15.1	5.8	1.3
820°C,180s/220°C,10s/400°C,10s	702	1186	12.5	17.0	8.5	1.2
820°C,180s/220°C,10s/400°C,30s	837	1138	9.9	16.3	8.6	1.3
820°C,180s/220°C,10s/400°C,100s	840	1118	10.1	14.9	5.0	1.4
820°C,180s/220°C,10s/400°C,1000s	817	1071	10.3	15.3	3.1	1.4
820°C,180s/240°C,10s/400°C,10s	664	1186	11.7	15.2	7.6	1.2
820°C,180s/240°C,10s/400°C,30s	736	1089	11.8	14.9	8.7	1.4
820°C,180s/240°C,10s/400°C,100s	801	1087	11.3	18.5	5.8	1.5
820°C,180s/240°C,10s/400°C,1000s	792	1056	11.6	17.6	5.6	1.5
820°C,180s/260°C,10s/400°C,10s	642	1183	12.7	16.0	7.9	1.3
820°C,180s/260°C,10s/400°C,30s	687	1088	13.9	18.8*	9.6	1.4
820°C,180s/260°C,10s/400°C,100s	750	1045	10.6	14.4*	9.0	1.5
820°C,180s/260°C,10s/400°C,1000s	734	1009	14.5	17.5	6.1	1.5

*total elongation reported for one sample,

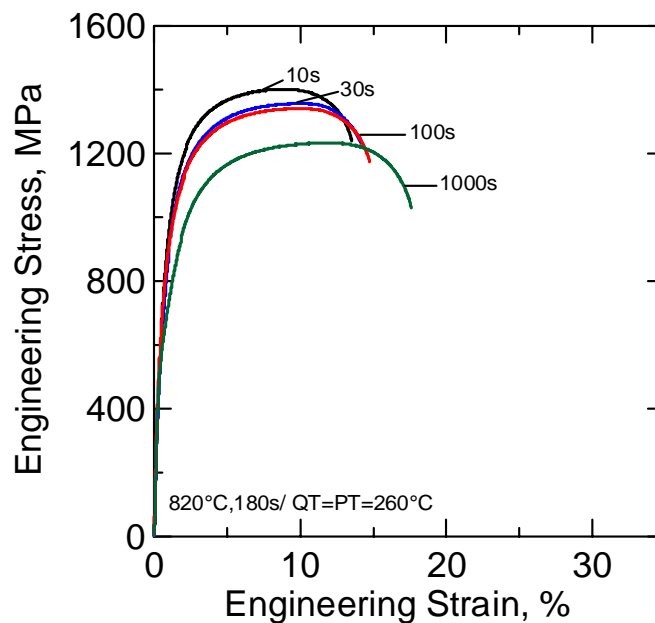**all properties reported for 1 sample

Figure 5.   Engineering stress vs. engineering strain for sub-sized tensile samples intercritically annealed at 820°C for 180s, and quenched to 260°C and partitioned for 10, 30, 100, or 1000s (1-step).
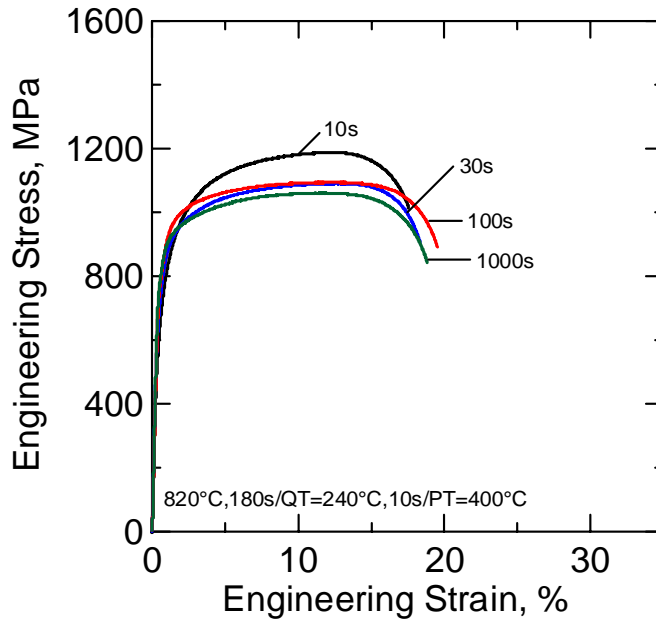
Figure 6.   Engineering stress vs. engineering strain for sub-sized tensile samples intercritically annealed at 820°C for 180s, quenched to 240°C for 10s, and partitioned at 400°C for 10, 30, 100, or 1000s (2-step).
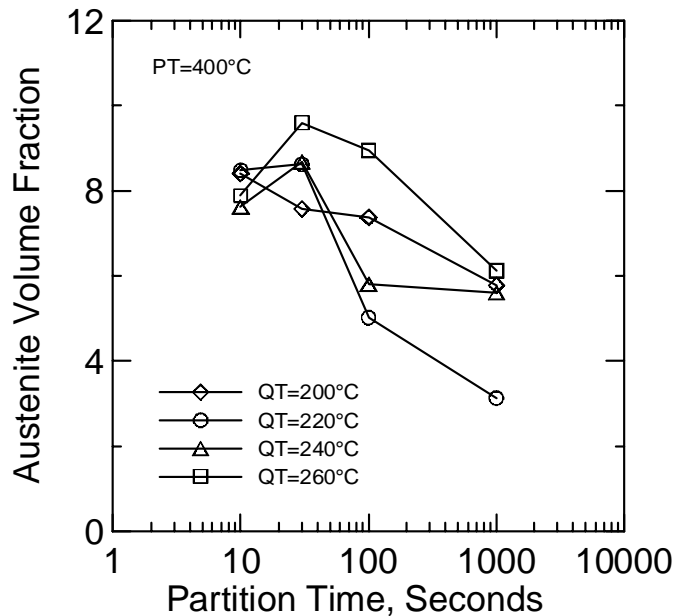

Figure 7.   Final austenite fraction at room temperature in a Si-TRIP sheet steel after 2-step Q&P processing (PT=400°C).

Conventional TRIP processing

For comparison with Q&P, samples of the steel in Table I were also processed by conventional TRIP heat-treating, consisting of intercritical annealing at 820°C for 180s and austempering at 400°C for times of 10, 30, 100, or 1000s without an intermediate quenching step, followed by final water-quenching to room temperature. Mechanical property results are summarized in Table V, and Figure 9 shows the stress-strain curves obtained for these TRIP heat-treated samples. The ultimate tensile strength and elongation results after austempering for 10s at 400°C, were similar to some of the Q&P results. Transformation of low stability austenite to martensite upon final quenching might explain the higher ultimate tensile strengths achieved by the short austemper time, compared to the other TRIP heat-treratments. At longer holding times, the conventional TRIP steel shows significantly lower tensile strengths, and increased elongations.

The 2-step Q&P heat-treating produced microstructures with high tensile strengths and good elongation properties. At short treatment times, austempered TRIP steels have properties similar to Q&P heat-treated steels, although yielding behavior is slightly different, while long austempering times resulted in higher elongations and lower strengths than Q&P heat-treated steels. Figure 10 shows uniform elongation vs. ultimate tensile strength for both 1-step and 2-step Q&P heat-treated samples, along with results for the same steel conventionally TRIP processed (austempered). Envelopes have been added to distinguish the type of processing, and highlight the overlap and "continuity" in mechanical properties to higher strength levels than are typical for TRIP

sheet steels. Thus, the results in Figure 10 suggest that Q&P might provide a means to increase the strength levels of this family of products.
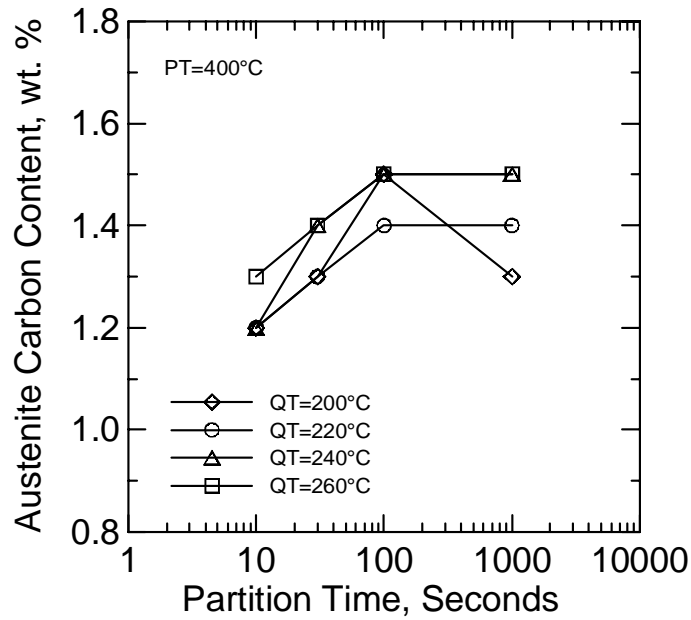

Figure 8.   Measured carbon concentration of retained austenite in a Si-TRIP sheet steel after 2-step Q&P processing (PT=400°C).

TABLE V – Average Tensile Test Results of Sub-sized Samples after Conventional TRIP Processing (0.19%C-1.59%Mn - 1.63%Si TRIP Sheet Steel).

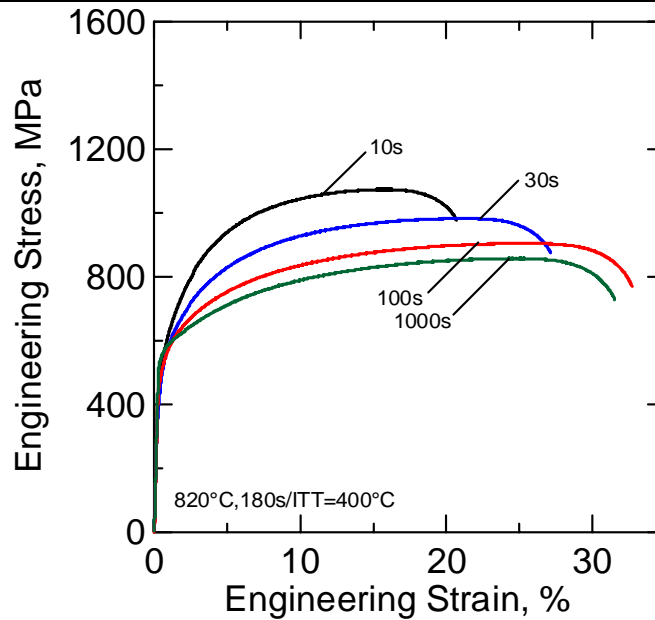| Heat Treatment	0.2% Yield (MPa)	UTS (MPa)	Uniform Elongation (%)	Total Elongation (%)	Retained Austenite (%)	Austenite Carbon Content (%)
820°C,180s/400°C,10s	503	1072	15.8	19.6	7.1	1.1
820°C,180s/400°C,30s	476	981	20.5	25.2	8.9	1.3
820°C,180s/400°C,100s	493	905	24.8	29.4	11.4	1.5
820°C,180s/400°C,1000s	551	859	23.6	26.6	12.3	1.4

Figure 9.   Engineering stress vs. engineering strain for samples intercritically annealed at 820°C for 180s, and austempered at 400°C for 10, 30, 100, or 1000s. ITT indicates the isothermal transformation (austempering) temperature.
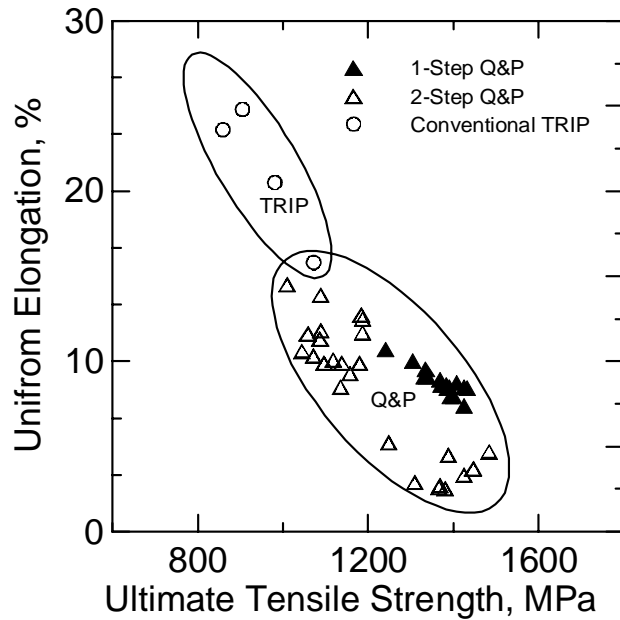
Figure 10.   Average uniform elongation vs. average ultimate tensile strength of duplicate tensile samples 1-step and 2-step Q&P heat-treated, compared to results after conventional TRIP processing (austempering).

Microstructure

Microstructures of 2-step Q&P heat-treated samples were investigated using scanning electron microscopy (SEM), normal to the plane of the sheet. Figure 11 shows the microstructure of an 820°C/200°C/400°C,10s heat-treated sample, etched in 2% nital. Dark feature-less intercritical ferrite is observed, in addition to smaller regions of martensite-austenite (M-A) constituent. The lath martensite structure is clearly indicated, although the substructure is much more pronounced than is typical for M-A constituent in dual-phase or conventional TRIP steels. This etching response is presumably associated with the partitioned martensite-austenite mixture. Coarser lath-like features (highlighted by the arrow) that appear similar to carbide-free bainite plus retained austenite are also present in the microstructure of Figure 12 after the 820°C/260°C/400°C,30s heat-treat sequence. For comparison, Figure 13 shows the microstructure produced from conventional 820°C/400°C,10s TRIP heat-treatment, clearly showing carbide-free bainite plus retained austenite. The lath-like features in the conventional austempered condition appear slightly finer than those shown in Figure 12 for the 2-step Q&P heat-treated steel. The substructure and morphology differences apparent in the Q&P heat-treated condition are believed to provide a clear distinction between partitioned martensite, and bainite. The amount of bainite is diminished at low quench temperatures (illustrated here by comparison of Figure 11 with Figure 12), consistent with this interpretation. Higher resolution studies will better characterize the microstructural features.
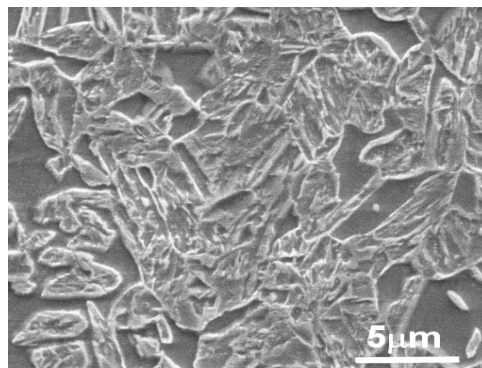

Figure 11.   Microstructure of Si-TRIP composition after intercritical annealing (75%γ) followed by quenching to 200°C and partitioning at 400°C for 10s, etched in 2% nital. This microstructure contains 8.4% austenite.
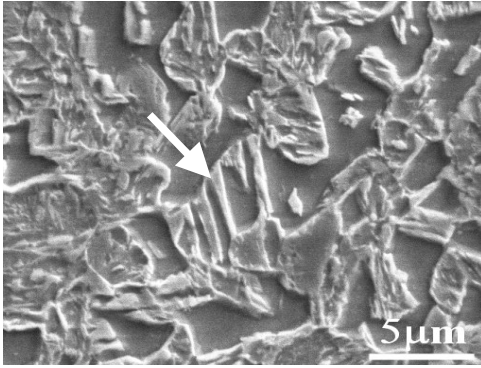
Figure 12. Microstructure of Si-TRIP composition after intercritical annealing (75%γ) followed by quenching to 260°C and partitioning at 400°C for 30s, etched in 2% nital.
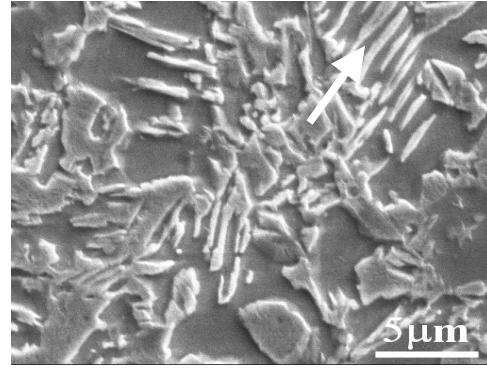

Figure 13. Microstructure of Si-TRIP composition after intercritical annealing (75%γ) followed austempering at 400°C for 10s, etched in 2% nital. (Arrow indicates carbide-free bainite.)

Comparison with other high strength sheet steels

The goal of this work was to produce a new family of steels that extended the strength levels of existing TRIP products using Q&P processing. Figure 14 shows total elongation vs. ultimate tensile strength for 1-step and 2-step Q&P heat-treated material, in addition to property results obtained from conventional austempering of the Si-TRIP composition used in this work. ULSAB (Ultra-Light Steel Auto-Body program)[12] and Ispat-Inland[13] reported mechanical properties of TRIP, dual-phase (DP), and martensitic (M) product are also shown for comparison. General Motors Worldwide Engineering Standard minimum requirements[14] are also indicated in Figure 14. Minimum elongation and ultimate tensile strength values and requirements are plotted, in addition to actual values obtained from this work. (Total, rather than uniform elongations are plotted here, due to the nature of the data available for comparison. Differences in specimen dimensions for the various ductility results should be recognized.) From the results of Figure 14, it is clearly evident that Q&P material does in fact extend the strength range of TRIP steels. High strengths and good ductilities were obtained, suggesting that various fractions of intercritical ferrite in conjunction with carbon-depleted martensite and retained austenite, produced by Q&P, offer unique and attractive property combinations. It appears that the Q&P process is an attractive method of producing new high strength steels, although it is recognized there are relatively few production facilities capable of applying the requisite thermal profiles at present.
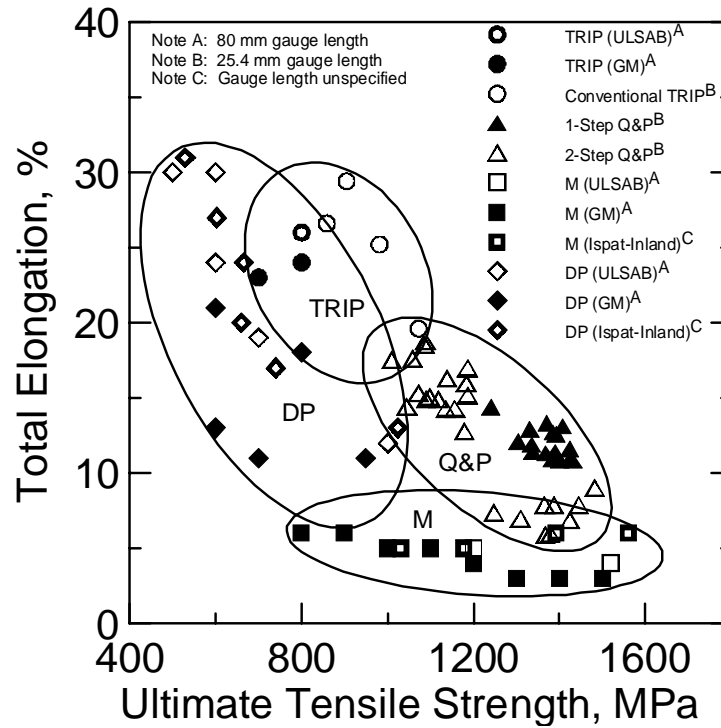

Figure 14. Total elongation vs. ultimate tensile strength for TRIP, DP, M, and Q&P materials.

CONCLUSIONS

1. New microstructures that extend the strength levels of current TRIP steels have been produced by Q&P processing. Further opportunity remains to explore available property combinations, however. Increased intercritical ferrite and optimized austenite fractions, along with 1-step Q&P processing at high quench temperatures with long partitioning times, may be worthy of additional investigation.

2. SEM microstructure evaluations show intercritical ferrite, in addition to martensite and retained austenite. Significant substructure is apparent in the Q&P heat-treated condition, making partitioned martensite generally distinguishable from bainite. Various amounts of carbide-free bainite plus austenite (typical of conventional TRIP steels), appear to be present for the highest quench temperatures examined (after partitioning at 400°C).

3. Carbon enrichment of austenite generally occurred as partitioning times increased to 1000 s for samples that were 2-step Q&P heat-treated. The retained austenite fraction decreased after long partitioning times at 400°C, however, and the presence of carbides is suggested in the x-ray diffraction results.

ACKNOWLEDGEMENTS

REFERENCES

1. J.G. Speer, D.K. Matlock, B.C. De Cooman, and J.G. Schroth, "Carbon Partitioning Into Austenite After Martensite Transformation", *Acta Materialia*, Vol. 51, 2003, pp. 2611-2622.

2. D.K. Matlock, V.E. Bräutigam, and J.G. Speer, "Application of the Quenching and Partitioning (Q&P) Process to a Medium-Carbon, High-Si Microalloyed Bar Steel", Proceedings of THERMEC' 2003, (Uetikon-Zurich, Switzerland: Trans. Tech. Publications, Inc., 2003) pp. 1089-1094.

3. J.G. Speer, A.M. Streicher, D.K. Matlock, F. Rizzo, and G. Krauss, "Quenching and partitioning: A Fundamentally New Process to Create High Strength TRIP Sheet Microstructures", Austenite Formation and Decomposition, Ed. E.B. Damm and M. Merwin, ISS/TMS, Warrendale, PA, USA, 2003, pp. 505-522.

4. G. Krauss, STEELS: Heat Treatment and Processing Principles, ASM International, Metals Park, OH, 1990.

5. K.W. Andrews, "Empirical Formulae for the Calculation of Some Transformation Temperatures", *Journal of Iron and Steel Institute*, July 1965, pp. 721-727.

6. M. De Meyer, J. Mahieu, and B.C. De Cooman, "Empirical Microstructure Prediction Method for Combined Intercritical Annealing and Bainitic Transformation of TRIP Steel," *Materials Science and Technology*, 18, 2002, pp. 1121-1132.

7. H.K.D.H. Bhadeshia, Bainite in Steels, The Institute of Materials, 2001, London.

8. H.K.D.H. Bhadeshia and D.V. Edmonds, "The Bainite Transformation in a Silicon Steel", *Metallurgical Transactions A*, Vol. 10A, 1979, pp. 895-907.

9. M.F. Gallagher, J.G. Speer, D.K. Matlock, and N.M. Fonstein, "Microstructure Development in TRIP-Sheet Steels Containing Si, Al, and P", 44th MWSP Conf. Proc., Iron and Steel Society, 2002, Warrendale, PA, pp. 153-172.

10. Retained Austenite and Its Measurement by X-Ray Diffraction, SP-453, Society of Automotive Engineers, Inc., January 1980, Warrendale, PA, p.12.

11. B.D. Cullity, Elements of X-Ray Diffraction, 2nd Ed., Addison-Wesley Publishing Co., Inc., 1978.

12. American Iron and Steel Institute, http://www.steel.org/autosteel/ulsab_avc/ttd6_appx2_main.htm, February 24, 2004.

13. Ispat-Inland Steel Company, Automotive Applications, http://www.ispat.com, March 8, 2004.

14. General Motors Worldwide Engineering Standard, December 2000.