Processing Opportunities for New Advanced High Strength Sheet Steels

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Abstract

Currently there is considerable interest in developments leading to new advanced high strength sheet steels (AHSS) for automotive and other transportation applications that demand high strength, light weight materials. Design requirements will involve material properties with strengths greater than currently available dual phase and TRIP steels (steels in this group are referred to as the “First Generation” of AHSS), with good ductility and formability, but produced at a cost less than the high ductility stainless steels or high manganese TWIP steels (materials referred to as the “Second Generation” of AHSS). Recent studies have shown that materials that satisfy the required property/cost combinations will include complex microstructures containing high amounts of retained austenite in combination with a high strength constituent that may be ultrafine grained ferrite, martensite, bainite, and or combinations of ferrite-based constituents. In this paper, selected methodologies leading to the production of new AHSS materials will be reviewed and assessed to provide a framework for consideration of new concepts and processing routes that will be required in production operations.

Keywords: *AHSS, TRIP, DP, TWIP, Steel.*

INTRODUCTION

Recently, significant interest has been expressed by steel producing and using industries for development of the next generation of advanced high strength sheet steels (AHSS) to meet the demanding requirements for excellent combinations of strength, formability, toughness, cost, etc [1–3]. To produce the new AHSS sheet steels, unique processing schemes will be required. It is anticipated that these will be based initially on modifications to routes currently employed to produce dual phase (DP) and transformation-induced plasticity (TRIP) steels. In this paper, selected approaches to produce new AHSS products, as well as to enhance existing AHSS steels, are presented. Results presented here are taken primarily from the research work recently published in the proceedings of the International Conference on New Developments in Advanced High Strength Sheet Steels, held in Orlando, Florida on June 15–18, 2008 [4], and the discussion concentrates principally on ferritic-based materials.

AHSS FAMILIES

AHSS steels in use today have been classified as “First Generation,” i.e. primarily ferrite-based steels which include e.g. DP, TRIP, complex-phase (CP), and martensitic (MART) steels or “Second Generation,” i.e. austenitic steels with high manganese and sometimes aluminum contents. The latter steels are closely related to some austenitic stainless steels. Of significant growing interest [1] is the potential to produce new steels, referred to as the “Third Generation” of AHSS with properties superior to the First Generation of AHSS. It is anticipated [2, 3] that development of steels with strength/ductility combinations greater than available in the First Generation of AHSS, but without the full cost of the stabilizing elements and processing required for the austenitic Second Generation AHSS steels, will require unique alloy/ microstructure combinations. To produce materials with the desired final properties, several microstructural features must be simultaneously and independently controlled. These include: number of constituents (e.g. phases) and constituent volume fractions, sizes, distributions, and mechanical properties (e.g. strength and strain hardening behavior). To illustrate the potential for developing materials with the desired microstructure/ property combinations, Figure 1 compares predicted effects of systematic microstructural variations, obtained by increasing the martensite volume fraction (up to 70 pct), in two hypothetical two-component composite materials of either ferrite plus martensite or stable austenite plus martensite [2, 3]. These predictions are superimposed on a strength-ductility map often used to compare sheet steel properties [1] and which identifies property bands associated with First and Second Generation AHSS. For these calculations, specific properties for each phase were assumed based on literature data and predictions utilized a composite model for two ductile constituents [2]. Figure 1 shows that the properties of a wide variety of sheet steel families, including the First Generation AHSS steels, are reasonably predicted with a two-component system based on ferritic steels and that steels in the Third Generation band may be produced by various combinations of martensite with stable austenite. Similar calculations based on metastable austenite have shown that the resulting composite properties depend sensitively on austenite stability against deformation induced transformation to martensite [3]. Composites with austenite that transforms to martensite at low strains exhibit properties similar to those shown in Figure 1 for the ferrite/martensite combination, while composites with austenite having enhanced stability against deformation exhibit properties within the Third Generation of AHSS band [3].

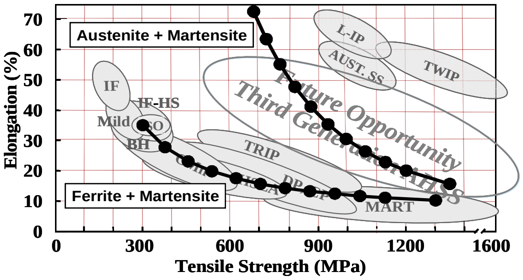

Figure 1: Predicted [2, 3] strength/ductility combinations for two hypothetical steel families with systematically varied martensite volume fractions with property bands for various classes of conventional and AHSS steels [1]. Predictions shown are for composites consisting of ferrite+martensite or stable austenite + martensite

While the predictions shown in Figure 1 provide insight into microstructural combinations required for the new AHSS, simplified assumptions were embedded in the composite model and consideration of actual deformation behavior of and interactions between constituents will be required to predict optimal microstructures. Nonetheless, it appears clear that the Third Generation of AHSS will consist of complex microstructural combinations, with significant use of both a) high strength constituents (e.g. martensite, bainite, ultrafine grained ferrite, etc) and b) high ductility constituents with significant strain hardening capacity (e.g. austenite with controlled stability against deformation induced transformation to martensite) [2, 3]. While development of modified TWIP steels and Second Generation AHSS are important and are receiving considerable attention, particularly in Europe and Asia, this paper concentrates on ferritic based materials. A discussion of developing Third Generation AHSS by starting with Second Generation AHSS products is left to future publications.

CONVENTIONAL COLD ROLLED AHSS PROCESSING

DP and TRIP steels can be produced as hot or cold rolled products, but current interest is focused more on cold rolled low carbon steels subjected to thermal cycles such as illustrated schematically in Figure 2 [5]. Low carbon steels rapidly cooled from an intercritical temperature result in DP steel microstructures consisting primarily of ferrite and martensite and which may contain small amounts of retained austenite. For appropriately designed steel compositions, if cooling is arrested and transformation is accomplished at an intermediate temperature, typically above the martensite start temperature, then TRIP steels consisting of microstructures with bainitic ferrite and larger quantities of austenite result after final cooling to room temperature. In either processing route, the final microstructure depends on control of a large number of variables including alloy content, starting microstructure, annealing temperature prior to cooling, time at temperature, heating and cooling rates, and incorporation of additional thermal cycles associated with hot dip galvanizing, etc. It should be noted that alloy designs have been based primarily on low carbon steels due to welding considerations and modifications to weld designs or joining processes may allow the use of higher carbon grades in the future, enabling new process/product concepts that are considered infeasible at present.

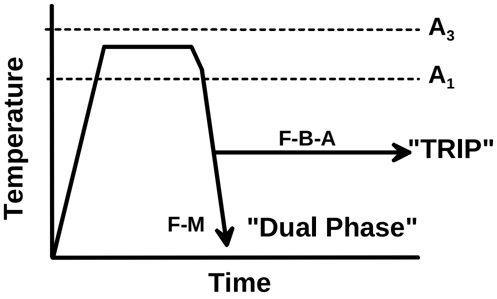

Figure 2: Schematic time-temperature diagram for TRIP and DP steel processing [5]

DP STEEL PROCESSING OPPORTUNITIES

Dual phase steels are the primary AHSS products currently being incorporated into future vehicles, and thus there remains significant economic incentive to produce DP steels with enhanced properties, i.e. higher strength with improved formability. The strength of cold rolled and annealed DP steels is primarily controlled by the martensite volume fraction, while for steels with a constant martensite volume fraction, the ductility and formability, including resistance to localized fracture, are controlled by the relative strength of the martensite and ferrite, and the martensite particle size and distribution [5, 6]. To achieve the necessary hardenability to produce martensite on cooling, DP steels often contain high Mn contents which may be greater than 2 wt pct. The presence of the high

Mn content may lead to banding and result in local inhomogeneous deformation and less than optimal formability [7].

In processing DP steels the importance of the hot rolled microstructure achieved on coiling has received limited attention in the literature. Melo *et al.* [8] compared the effect of coiling temperature (450 v. 650^{o}C) after hot rolling on the DP microstructures developed after intercritical annealing of a cold rolled 0.15C-2.14Mn-0.4Si (in wt pct) steel, and resulting light optical micrographs illustrating the importance of coiling temperature are shown in Figure 3. A banded pearlitic microstructure existed in the sample coiled at 650^{o}C (Figure 3(c)) while the material coiled at 450^{o}C was uniformly bainitic (Figure 3(a)). The fine structure associated with bainite was maintained in the cold rolled samples. On subsequent DP processing from 760^{o}C, the bainitic material produced a more uniform martensite distribution (Figure 3(b)) than was evident in the DP microstructure (Figure 3(d)) produced from the material coiled at 650^{o}C.

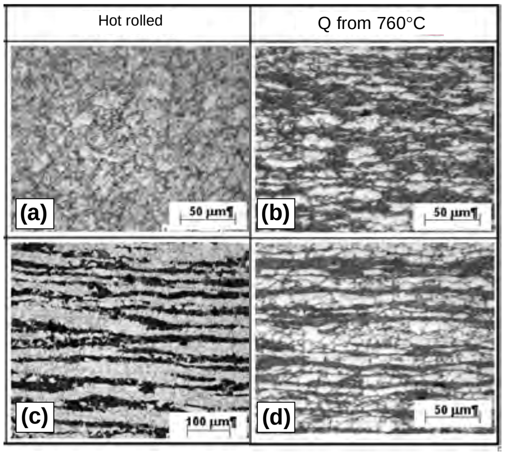

Figure 3: Light optical micrographs showing the effect of microstructure after coiling at 450^{o}C (a) or 650^{o}C (c) on DP microstructures produced after quenching from 760^{o}C, (b) and (d) respectively. Adapted from [8]

A novel path to produce DP steels with enhanced properties was illustrated by Azizi-Alziamini *et al.* [9] using a 0.17C-0.74Mn steel in which the five-stage processing history shown in Figure 4 was employed to produce a ultrafine grained DP steel. After an ice-brine quench (Step I), the steel possessed a fully martensitic microstructure that was tempered in Step II prior to cold rolling (Step III). A recrystallization pre-anneal at 550^{o}C (Step IV) produced an ultrafine aggregate of ferrite and carbide (cementite). Stage V incorporated a rapid heating rate and short times at the intercricical annealing temperature, both chosen to minimize ferrite grain growth and promote carbide dissolution and austenite formation. On quenching, an ultrafine structure developed with a martensite volume fraction of approximately 0.42 having a uniform distribution of 2 μm martensite islands. While not shown, in this material the strength difference between the ferrite and martensite should be less than in other DP steels owing to the grain size refinement of the ferrite, and consistent with the modeling predictions of Krempaszky *et al.* [7], should lead to improved overall material performance in tests limited by decohesion fracture and non-uniform second phase distribution. In a related study [10], incorporation of an additional reduction step in the interval normally associated with the runout table, can be used to enhance deformation induced ferrite formation resulting in ultrafine grained DP products.

Another path to precondition sheet steels prior to DP processing, particularly for steels of interest for hot dip galvanizing operations, has also been considered recently. The response of materials to DP processing depends on the distribution of alloying elements in addition to carbon, and renewed interest has been expressed [11] to implement unique processing histories to modify Mn distributions to enhance the Mn content in the austenite present at the intercritical annealing temperature. This approach was evaluated recently by comparing the response to DP processing of a 0.2C-2.0Mn steel coiled at 550^{o}C with the same steel after an additional hot band annealing treatment (0.5 hr at 750^{o}C). Addition of the extra step produced material with a better combination of strength and ductility with reduced sensitivity to intercritical annealing conditions. This example, along with others highlighted above and in later sections, illustrates that flexibility in processing is needed to facilitate implementation of new microstructure concepts.

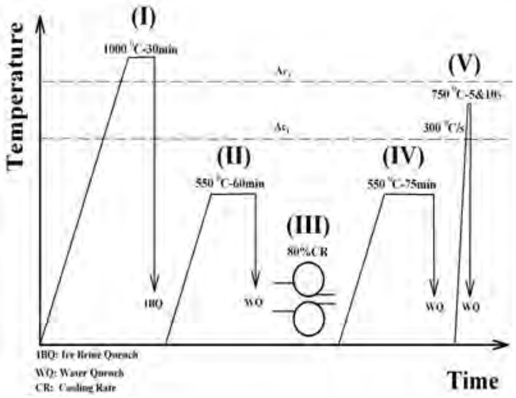

Figure 4: Schematic time-temperature history used to produce ultrafine grained DP steels [9]

TRIP STEEL PROCESSING OPPORTUNITIES

Modifications to conventional TRIP steel alloys and processing routes potentially offer the most direct paths to extending material properties into the third generation AHSS strength-ductility band shown in Figure 1. In comparison to currently available TRIP steels, to achieve third generation AHSS properties, two primary modifications to TRIP steel microstructures are required [3]: a) increase the volume fraction of austenite, and b) increase the stability of austenite against transformation to martensite with strain. Austenite stability increases with an increase in carbon and alloy content, a reduction in austenite particle size [12], and segregation of other austenite stabilizing elements to austenite during intercritical annealing.

The ability to modify austenite stability during the isothermal bainitic heat treatment in Figure 2 depends sensitively on the alloy/time/temperature combinations employed. This is illustrated a recent comparison of TRIP steels containing aluminum. Spenger *et al.* [13] showed for an Al-stabilized TRIP700 that an increase in time at the bainitic isothermal hold temperature of 400^{o}C increased the initial austenite volume fraction from approximately 0.1 to 0.12 and the stability against transformation with strain. In contrast, with a 0.18C-1.58Mn-1.3Al alloy for use in hot dip galvanized TRIP600 products, Li *et al.* [14] showed that for comparable annealing times the amount of retained austenite decreased with an increase in isothermal transformation temperature above 460^{o}C, an effect that was enhanced for longer transformation times.

The potential to use nitrogen to stabilize austenite in TRIP steels with commercially interesting compositions was considered in the study by Lee *et al.* [14], on nominally 0.08C-1.5Mn-1.5Si steels with and without 0.03 Al. The Al-addition served to getter nitrogen, producing steel with limited free nitrogen. Figure 5 shows the effects of hot dip galvanizing temperature for a constant processing time on the initial retained austenite content of the two alloys. The alloy without the aluminum addition exhibited a higher amount of retained austenite at all temperatures, and the increased stability of the austenite was attributed to rejection of nitrogen from cementite during bainite formation and corresponding stabilization of austenite due to an increase in the local nitrogen content in the austenite adjacent to cementite-austenite interfaces. Use of nitrogen as an austenite stabilizer was suggested to provide improved mechanical properties, although the extent of the effect is limited by nitrogen solubility, and 200 ppm N was indicated to be the likely limit using conventional steelmaking practice.

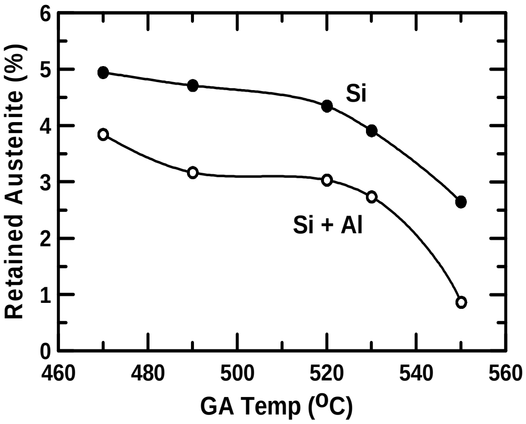

Figure 5: The effects of galvanizing temperature on the retained austenite content in a steel modified with Al to getter nitrogen in comparison to the same steel without Al. Adapted from [14]

MARTENSITIC STEEL PROCESSING OPPORTUNITIES

Fully martensitic sheet steels are receiving increased attention as they possess high strength and are currently being incorporated as formable automotive steels as a result of developments in hot stamping where the high strength martensite develops as a consequence of die quenching after stamping. As most martensitic steels are employed after tempering, Säglitz *et al.* [15] evaluated the tempering response of a typical B-modified 0.22C press-hardening steel after water quenching. Figure 6 shows a summary of their results plotted as ultimate tensile strength versus total tensile ductility. As-quenched martensite has high strength with ductility that is essentially equivalent to tempered structures produced by low temperature tempering. The observation of equivalent ductilities indicates that the as-quenched material should perform at least as well as low-temperature-tempered microstructures. The data show that increased tempering in the range of 150 to 350°C decreases total elongation somewhat, a consequence of a form of tempered martensite embrittlement due to enhanced void nucleation at carbide particles formed during tempering. This observation further suggests that the use of materials in the as-quenched or auto-tempered conditions are appropriate as the degree of tempered martensite embrittlement does not appear severe.

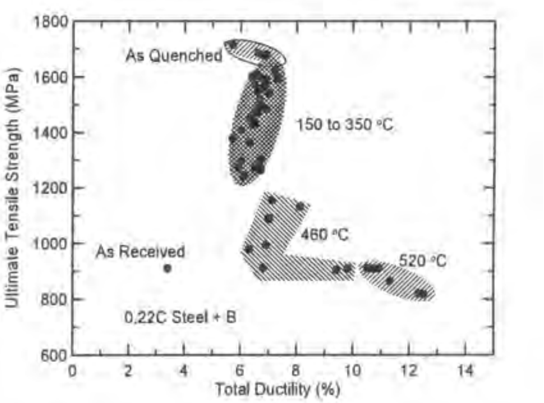

Figure 6: Combinations of ultimate tensile strength and total elongation for martensitic 0.22 pct C boron steel tempered at various times and temperatures. Adapted from [15]

NOVEL PROCESSING OF AHSS CONTAINING MARTENSITE

Recently there has been considerable research interest in cold rolled and annealed materials rapidly cooled to a temperature below the M_s but above room temperature after heating to $T > A_3$ or to an intercritical annealing temperature. This sequence is akin to a modified martempering process as employed in quenched and tempered steels, but is designed to modify the microstructure rather than the thermal stresses. After cooling, the material is held at the quench temperature (QT) or is reheated slightly and held for a specified time prior to cooling to room temperature. Materials produced by this interrupted quenching process have been referred to by a variety of terms including "quenching and partitioning (Q&P)" [16, 17], "TRIP-dual" [18], "TRIP aided bainitic ferrite (TBF)" [19], "isothermal quench and tempering (IQ + T)" [20], and "quenching, partitioning, and tempering (Q-P-T)" [21]. For the discussion that follows, steels from these studies are referred to as "Q&P" steels, although Q&P processing usually implies quenching temperatures meaningfully below the martensite start temperature (M_s). On quenching, the amount of initial martensite that forms depends on the M_s for the austenite present during annealing, and the difference between the M_s and QT. Subsequent changes in the microstructure on isothermal holding prior to cooling to room temperature depend on alloy content, temperature, and holding time. Figure 7 summarizes results of Jun and Fonstein [18] on a nominally 0.2C-1.7Mn-1.5Si Al-killed steel processed with various thermal histories including "TRIP-dual" isothermally transformed at 400°C (essentially a TRIP austempering process as shown in Fig. 2) or transformed below the M_s (a process similar to 1-step Q&P [16, 17]), a conventional Q&T process for sheet quenched to room temperature prior to tempering, and two different Q&P processing histories processed by quenching to either 240 or 260°C followed by an isothermal hold at 400°C (this process is equivalent to those referred to previously as 2-step Q&P [16, 17]). In a format analogous to Figure 1, Figure 7 shows that for similar strength levels, Q&P steels exhibit higher ductilities than the martensitic Q&T grades and thus offer a potential processing path to produce new third generation AHSS steels.

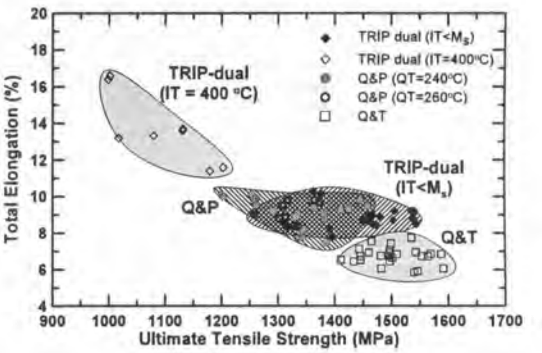

Figure 7: The effects of processing histories on the strength-ductility combinations obtained using a 0.2C-1.7Mn-1.5Si Al-killed steel. Adapted from [18]

During the intermediate isothermal hold, several potential microstructural modifications can occur including: carbon partitioning from the martensite to the austenite resulting in increased volume fractions of austenite at room temperature [16], isothermal martensite formation [22], transition carbide formation [23], migration of the martensite/austenite interface, carbon trapping at dislocations, bainite formation, and cementite formation [24]. If carbon partitioning can be promoted in lieu of the other possibilities, then the potential to produce new high strength steels with controlled amounts of ferrite (controlled by the annealing temperature prior to quenching), martensite, and retained austenite can be realized. One way to promote austenite stabilization is to add Si to suppress cementite formation and Figure 8 illustrates this effect in the recent work of Santofimia *et al.* [24] on a nominally 0.2C-3.5Mn steel with either 0.45 or 1.54 Si. For all partitioning times, the high Si steel had higher retained austenite contents due to the suppression of cementite formation and for times greater than 100 s, there was little change in the final austenite content, indicating the presence of a wide processing window. Other studies on lower Mn steels with Si or Si + Al have shown that Mo additions also promote austenite stability by possibly retarding bainite formation through a solute drag like effect [25].

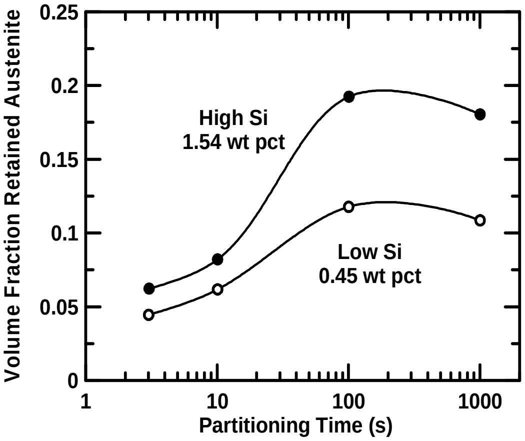

Figure 8: The effects of Si content (0.45 or 1.54 wt pct) and processing time on the volume fraction of retained austenite in a 0.2C-3.5Mn steel Q&P processed with QT = 240°C and PT = 350°C. Adapted from [24]; error bars deleted for clarity

While not covered extensively here, it is worth noting that new processing approaches are being considered to develop Q&P steels for processing on a runout table after hot rolling [26] and AHSS sheet steels with controlled microstructural gradients or bimodial grain size distributions [7]. These alternate approaches to develop new steel microstructures may prove interesting in the future.

IMPLICATIONS FOR PROCESSING AND EQUIPMENT REQUIREMENTS

Many of the new processing routes that are currently being evaluated, or have been proposed, require careful control of heating and cooling rates and deformation histories that may exceed capacities of current hot and cold rolling, annealing, and galvanizing lines. As a result, there is significant ongoing development and implementation of new processing steps to facilitate production of new materials. As illustrated in Figure 4, both high heating rates and cooling rates enhance production of ultrafine grained DP steels. Rapid heating rates can be obtained through installation of inline induction heating systems at critical stages in the process stream. For example, installation of an induction heater between the annealing furnace and zinc pot, in conjunction with a rapid cooling system on exit of a continuous annealing furnace, would allow the possibility of producing Q&P products on a hot dip galvanizing line. Installation of an induction heating coil and/or a controlled cooling system on a runout table would provide the opportunity to control coiling temperature for the purpose of more completely using coiling as one step in the overall heat treating process. Of particular note is the compact cooling system discussed by Sprock *et al.* [27] where cooling, using high water flow rates and critical control of water nozzle geometries, equivalent to that achieved in a conventional laminar flow system was obtainable in a space of 7 m in contrast to the 57 m required for conventional processing. The compact size of this system was shown to offer a novel system design that possibly could be adapted to both new and existing lines. While not exhaustive, these examples clearly illustrate that equipment manufactures are prepared to respond to the potentially more severe processing conditions (i.e. high heating and cooling rates) that will be required to produce several of the new Third Generation of AHSS steels.

As most automotive sheet steels are coated for corrosion control, another important process parameter that must be considered is the surface chemistry of the as-heat-treated sheet prior to entering a coating operation. Many of the alloy systems discussed above, particularly the TRIP and Q&P steels, involve Si additions to control transformation response on cooling. However, high Si steels are known to exhibit significant difficulties in conventional heat treating and galvanizing operations owing to silicon-containing oxides on sheet surfaces. The presence of surface oxides necessitates modifications in processing to remove or modify the oxide to ensure complete Zn wetting of the steel surface during hot dip galvanizing. The formation of surface oxides is reportedly minimized by utilizing a protective atmosphere furnace with hydrogen in place of conventional inline annealing furnaces [28], although availability of these types of systems in current mills is very limited. Van De Putte *et al.* [29] recently illustrated the degree that surface oxides can be modified in their study of austenitic annealing of a 0.23C-1.45Si-1.61Mn-0.023Al steel in a N_2-5%H_2 atmosphere. Figure 9 shows the surface Si and Fe concentrations in samples rapidly heated (open symbols) or soaked for 70 s (solid symbols) at the indicated temperatures. There appears to be a critical temperature where there is a distinct transition in surface Si concentration. The disappearance of the oxide at the surface was interpreted to occur by the reduction of the SiO_2 film by dissolved carbon in the steel, and zinc was shown to wet samples annealed above the transition temperature.

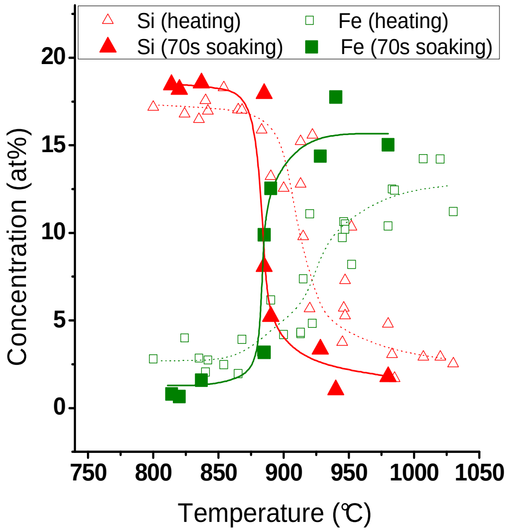

Figure 9: Effects of annealing temperature on the surface Si and Fe concentration (heating experiments—dotted lines and isothermal annealing—solid lines). Adapted from [29]

SUMMARY

Recent results from a 2008 international conference on AHSS were used to highlight opportunities for modifications to existing and development of novel paths for the production of new high strength sheet steels with strength-ductility combinations superior to those available today. The steels of interest will contain significant amounts of a highly ductile constituent (e.g. austenite) and high strength constituents, including martensite, bainite, ultrafine grained ferrite, etc. The selected examples illustrate that significant opportunities for the production of new AHSS products have been identified, although most require higher heating and/or cooling rates, along with multiple thermal or deformation processing cycles not currently available on most production lines. The challenges for the future will be to design methodologies to economically implement these new processing routes.

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