

Development of Rapid Heating and Cooling (Flash Processing) Process to Produce Advanced High Strength Steel Microstructures

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Abstract

Flash processing of a AISI8620 steel sheet, which involves rapid heating and cooling with an overall process duration of less than 10 seconds, produced a steel microstructure with high tensile strength and good ductility similar to that of advanced high strength steels. Flash processed steel showed at least 7% higher UTS and 30% greater elongation than published results on martensitic advanced high strength steel. The underlying microstructure was characterized with optical, scanning electron, transmission electron-microscopy, as well as, hardness mapping. A complex distribution of bainitic and martensite microstructures with carbides was observed. A mechanism for the above microstructure evolution is proposed.

Introduction

There is a need for weight reduction of automotive structures in order to achieve improved fuel efficiency, while not compromising the safety of passengers. To meet these two demands, a new class of steels known as advanced high-strength steels (AHSS) was developed by the Ultra-Light Steel Automotive Body-Advanced Vehicle Concept (ULSAB-AVC) Consortium ^[1, 2]. This family of steels shows good formability while maintaining very high strength values [engineering yield strength >300 MPa and ultimate tensile strength >700 MPa] ^[3]. These AHSS grades include dual phase (DP), transformation induce plasticity (TRIP), complex phase (CP) and partially martensitic (PM) steels. All these steels achieve their mechanical properties by engineering the fractions of ferrite, bainite, and martensite microstructure, as well

as, austenite phase. This is achieved either through energy-intensive thermo-mechanical processing steps and/or by expensive alloying additions (e.g. Manganese) [3]

Need for producing low-cost and advanced steel development is imminent. This is demonstrated by research directions of many researchers throughout the world. For example, researchers from Colorado School of Mines have developed the steels that rely on formation of martensite and partitioning of carbon to retained austenite. This heat treatment produces high-strength and -tough microstructures [4]. Similarly, University of Cambridge researchers have developed high-strength high-toughness microstructures by low-temperature austempering process. During this heat treatment, a high-carbon austenite transforms to nano-scale thick bainitic microstructure [5] with no carbides. In addition, researchers from Japan have achieved fine ferrite grain structure by coupled deformation and phase transformation [6] to achieve good combination of strength and ductility. Other researchers have focused on minimizing the packet and sub-unit size in low-carbon martensite to achieve advanced high strength and toughness [7]. In addition to the production of these steels, there exists a critical need to develop reliable and robust processes for forming [8] and welding [9] of these steels too. The above examples are not a comprehensive representation of the published literature. But, it proves that there is an impetus to develop low-cost processing techniques to produce AHSS.

This paper pertains to an innovative heat treatment procedure that shows potential as an alternative route for the current production of AHSS steels [10]. This process has been termed as '*Flash Processing*' for the incredibly short time (< 10 seconds) for heating and cooling of the steel sheets. Throughout this paper this term or an acronym "FP" will be used to denote steels that are processed using this method. Preliminary work has shown that flash-processing of plain carbon-steels (> 0.15C) may lead to high yield strength (>1200 MPa), tensile strength (>1500 MPa) and appreciable ductility (> 7%). As a result, FP processed steels could be classified under the category of AHSS [10].

Cursory evaluation of the above claim leads to skepticism, since it contradicts the well-established theories of phase transformation and microstructure evolution in steels [11]. Heating of steels to the homogenous austenite phase field and rapid quenching will lead to the formation of brittle martensite. During tensile testing of this martensitic microstructure, no appreciable plastic strain is expected. Therefore, the objective of this paper is to relate the observed thermal cycle, microstructure and the observed properties by developing a mechanistic understanding of the microstructure evolution.

Experimental

Steel Composition and Initial Condition

In earlier work, Cola has investigated many steels including AISI8620 steels by flash processing [10]. Results such as 1900MPa at 8-9% elongation have been found with multiple low carbon, lean steels. In the current paper, to limit the effect of the steel compositions on the interpretations of the microstructure and property correlations, only AISI8620-grade steel sheets were investigated. The measured composition of the steel is shown in Table I. The original steel was in the annealed condition and the microstructure constitution was ferrite and uniform dispersion of carbides. The optical microstructure and scanning electron microstructure of the un-processed samples will be presented later. The thickness of the steel sheets was 1.6 mm.

Flash Processing Set-Up and Heat-Treatments

A schematic of the process setup is shown in Fig. 1. The assembly consists of a pair of rollers that transfers the steel sheets through a heating and cooling stage. At a controlled distance from the top pair of rollers, the heating stage is placed. The heating stage is either based on either oxy-propane flame or electrical induction heating. The flame heating consists of 17 flame nozzles spaced evenly to spread the heat over the steel sheets. Spot temperature check using an Ircon® infrared pyrometer showed a small variation of temperature of approximately $\pm 10^{\circ}\text{C}$ from the middle to the edges of the sample. Immediately below the heating port, a cooling trough is placed. This trough can use any cooling medium and in the current

work water is used as a cooling medium. Water in the trough is continuously agitated. The water temperature is maintained by a water chiller and a closed loop circulation systems. To avoid inefficient heating due to the upward flow of steam, a graphite separator film is used to separate the heating and cooling ports. The cooling trough has a slit in the middle of its base to allow the steel to pass through. A rubber lining was used around this slit and it acts as a seal to prevent water from flowing down through the slit along with the processed steel sheets. These seals do not deteriorate during flash processing, since the steel surface reaches the ambient water temperature before the steel exits the water trough.

The above set-up has been equipped with infrared pyrometers, instrumentation to drive the feed rollers, mechanical fixtures for positioning heating and cooling units, and heating intensity controls. During flame heating, the heating intensity is controlled by regulating the oxygen and propane flow rate, as well as, the mixture of the same. In case of induction heating, the power characteristics are controlled to regulate the heating intensity. With the above set up, the steel heating/cooling rate and the peak temperature the steel can be controlled. In addition, this set-up provides flexibility to control the overall heating and cooling cycles. In this experiment, the steel sheets were fed through the heating and cooling system at a rate of 28 cm/min (~11 inch/min) while maintaining the peak temperature at 1100 °C. In order to compare the flash processed samples with a reference sample, another AISI8620 unprocessed steel sheet was subjected to a quench and tempering (QT) treatment. This heat treatment involves austenitizing at 1000°C for 2 minutes, quenching into a water bath and furnace tempering at 220°C for 4 minutes.

Thermal Cycle Measurements and Analyses

Cola ^[10] calculated thermal cycles at different regions using infrared pyrometers and imposed feed rates of steel sheets. However, the detailed temporal variations of temperature were not measured using contact thermocouples. Such measurements are necessary to develop a mechanistic understanding of microstructure evolution during flash processing. However, the oxy-propane flame heating method does not lend itself for the direct measurement of temperature due to possible deterioration

of thermocouples by the high-intensity flame. Therefore, the induction heating method was used during temperature measurement experiments. In order to impose the similar heating conditions as of that of the flame heating, feed rates, temperature gradients and the peak temperature were controlled to be identical. Type-K (Chromel - Alumel) thermocouples were attached to the middle of the steel sheets. The temperature of the steel was measured, as the sheet traverses through the heating and cooling ports. These measurements are made with a high-speed data acquisition system capable of recording temperature at a sampling rate of up to 5 KHz. The measured thermal cycles were analyzed to determine the instantaneous heating and cooling rates throughout the flash processing. In addition, ferrite (α) to austenite (γ) transformation was evaluated by analyzing thermal cycle data using Single Sensor Differential Thermal Analysis (SS DTA) technique [12]. The SS DTA is an innovative technique for in-situ measurement of thermal phenomena taking place during well-described heating/cooling experiments. The sensitivity and accuracy of SS DTA technique, in measuring α to γ transformation during high heating rates ($\sim 500^\circ\text{Cs}^{-1}$), has been documented earlier [13] by direct comparison with dilatometric techniques.

Mechanical Property Measurements

In order to evaluate the spatial variations of steel microstructures, two-dimensional hardness distributions were measured in the cross section of samples. For each sample, more than 2000 indents were made on QT and FP samples using an AMH43 automatic hardness testing system with a load of 300g. The measured hardness distributions were analyzed in a map format and using frequency distribution curves to provide quantitative measure of mechanical heterogeneity of the samples.

Cola [10] measured tensile properties of these steels after flash processing. In this paper, the tensile tests were carried out to reconfirm the Cola's measurements and to extract the strain hardening exponents in the plastic zone. Two samples were extracted from the edge region and one from the middle region. Tensile tests were performed along the longitudinal direction of the coils. The gauge length of the test samples was 25 mm (~ 1 inch). The width of the samples at the gauge length was

6.3 mm. The tests were performed at room temperature with a crosshead displacement rate of 1.27 mm/min. All the tensile testing were performed using testing machines available at Edison Welding Institute. Similar tests were repeated in quenched and tempered (QT) samples also.

Microstructure Characterization

Optical microscopy was used to characterize microstructure from the FP and QT samples. The sectioned samples were mechanically polished down to less than 1 μm finish. The polished samples were then etched in 2% Nital etchant for 5 seconds. In the next step, the samples were characterized using a Quanta200 scanning electron microscope (SEM), equipped with Secondary Electron (SE), Back Scattered (BSE) and X-ray energy dispersive analyses (EDS) detectors. The microscope was operated at an accelerating voltage of 25kV.

Transmission electron microscopy (TEM) samples were prepared by electrolytic thinning of 3mm discs of 90 μm thick samples. A chemical solution of 33% Nitric acid, 67% Methanol was used as the electrolyte. The polishing was carried out at 25V at a current of 10-15 mA and the temperature of the electrolyte was maintained at -25°C . The samples were imaged using two types of transmission electron microscopes. General microstructure and electron diffraction analyses were performed in a Phillips CM12 transmission electron microscope operating at 120kV. In addition, High angle annular dark field (HAADF) and bright field imaging was performed using scanning transmission electron microscopy (STEM) technique in a Tecnai F-20 FEG/TEM operating at 200kV. The STEM HAADF minimizes the contrast due to diffraction effects and facilitates better imaging of the lath structure and the inter-lath boundaries that have very low misorientations. Energy dispersive X-ray analysis (EDS) was used to characterize the chemical composition of the carbides.

Computational Modeling

ThermoCalc[™] [14] software with TCFE5 database was used to estimate the thermodynamic quantities and equilibrium phase transformation temperatures. In

addition, the time-temperature-transformation (TTT) diagrams for the initiation of reconstructive and displacive transformations were predicted using the methodology developed by Bhadeshia [15,16]. The TTT data were then converted into a CCT diagram using the methodology developed by Babu and Bhadeshia [17]. The source code for these calculations can be downloaded from an internet location [18] and example calculations can be made using an on-line calculator [19].

Results and Discussions

Calculated Phase Transformation Characteristics

In order to analyze the results from the flash processing experiments with reference to expected transformation behaviors, the transformation temperatures were calculated using the methodologies described before [14,19] for the 8620 steel composition. The equilibrium A1 and A3 temperatures of this steel were calculated to be 702°C and 806°C, respectively. In addition, the calculated equilibrium volume fractions of cementite (M_3C) and alloy carbides [$M_{23}C_6$ and MC] that can form in these steels as a function of temperature are shown in Fig. 2a. This result shows that as the steel heats above 810°C, it should transform to 100% austenite.

The bainitic and martensitic start temperatures [15, 16] were calculated to be 553°C and 426°C respectively. Calculated TTT and CCT diagrams for the initiation of paraequilibrium[†] reconstructive mode of austenite to allotriomorphic ferrite transformation and displacive mode of austenite to Widmanstätten and bainitic transformations are shown in Fig. 2b. Since the flash processing involves rapid cooling process, the assumption of paraequilibrium is indeed valid. Based on the above results, one can expect complete transformation of austenite to 100% martensite in 8620 steels at cooling rates faster than 70 °Cs⁻¹. These predictive methodologies have been developed using a wide range of steels and their time-temperature-transformation data presented in the literature [16,20]. Therefore, the predictions are expected to be fairly accurate for low-alloy steels including 8620

[†]Transformation controlled by carbon partitioning and the ration of substitutional to iron atoms are configurationally frozen in place on either side of austenite and ferrite transformation interface.

grades. For example, these calculations have accurately predicted the transformation start during continuous cooling of low-alloy steels [21].

Thermal Cycles during Flash Processing

A typical heating and cooling thermal cycle during flash processing is shown in Fig. 3. The plot also shows the heating and cooling rate as a function of time. These thermal profiles can be divided into four regions based on the temperature range. In region I, the steel temperature increases gradually due to heat conduction from the hot stage. In region II, the steel temperature increases rapidly due to its proximity to heat source. In this region, maximum heating rate of $410\text{ }^{\circ}\text{C}\cdot\text{s}^{-1}$ was observed at 780°C . This temperature is in between the calculated A1 and A3 transformation temperatures for 8620 steel. Above $780\text{ }^{\circ}\text{C}$, the heating rate starts reducing indicating that there is a dynamic equilibrium between the heat flux, the heat conduction, and endothermic/exothermic effects due to phase transformations [12]. In region III, the heating rate reaches zero when the sample reaches a peak temperature of 1100°C . As per the equilibrium thermodynamic calculations, at this temperature, the sample should be 100% austenite. After reaching the peak temperature, the sample starts cooling down gradually. The total dwell time for this slow cooling near 1100°C is measured to be 2 seconds. In region IV, within 1.2 seconds of reaching peak temperature, the sample reaches the water bath and starts cooling rapidly. The maximum cooling rate of $3150\text{ }^{\circ}\text{C}\cdot\text{s}^{-1}$ is achieved when the sample reaches a temperature of 393°C . This temperature is much below the calculated B_s (553°C) and M_s (426°C) temperatures [15]. As per the measured thermal cycles and the calculated CCT diagram (See Fig. 2b), the austenite phase from high temperature should transform to 100% martensite during flash processing. Fig. 4 shows the complete thermal profile of two separate FP treatment runs recorded from thermocouples attached at the center of the sheet. This shows the thermal cycle repeatability that can be achieved in flash processing.

Alexandrov and Lippold [12] have developed the SSDTA software to detect transformation temperatures from measured thermal cycles, in a well-described [Newtonian, Gaussian or Rosenthal type] heat-transfer condition. The

transformation temperatures are detected by looking for a departure of measured temperature from a reference heating (or cooling) curve due to endothermic (or exothermic) effects during phase changes. Fig. 5a and 5b show the results of the SS DTA analyses of thermal profiles shown in Fig. 5, for the heating regime only. In these analyses, the reference curves were described by a polynomial function that describes a well-controlled heat transfer conditions.

Both the results show a small endothermic phenomenon around the calculated ^[14] Curie temperature [~ 750 °C] of BCC ferrite and another large endothermic peak around 900°C. These large endothermic peaks are interpreted as A_{c1} (926°C and 930°C) and A_{c3} (1046°C and 1052°C) temperatures by SSDTA software. The reproducibility of these measurements is within the 0.4% accuracy of the type-K thermocouple. The accuracy of these analyses is also affected by the electromagnetic noise generated by the induction heating. It is noteworthy that SSDTA technique estimates a higher and non-equilibrium A_{c1} and A_{c3} temperature, in comparison to equilibrium A_1 (702 °C) and A_3 (806 °C) temperatures. Possible mechanisms for these are discussed later.

Hardness and Tensile Properties

Measured engineering tensile properties are presented in Table II. The table II also shows the tensile properties of the FP samples tested by Cola ^[10] and the current work. The data show that the flash processed 8620 steels do have yield and tensile-strengths greater than 1300 MPa and 1500 MPa, respectively. In addition, the elongation and reduction in area indicate appreciable ductility in these samples. The ductility of 8.8 to 9.9% is indeed greater than the levels reported for martensitic steels ^[22]. The UTS of the FP tensile specimen is comparable to that of the QT sample; however, the FP samples, consistently, show a higher elongation values.

In order to evaluate the formability of FP steels, the ratio of YS and UTS and true stress -true strain characteristics were calculated [see Fig. 6]. The strain hardening exponent (n) and pre-exponent strength coefficient (K) for uniform plastic strain was estimated by fitting the following equation to the plastic regime of the true stress-strain curve.

$$\sigma = K \cdot \epsilon^n \quad (1)$$

The calculated K and n values [see Table III] are similar value to the published grades of AHSS steels for DP and CP steels [22]. The measured UTS and YS from the current work are overlaid on a map of current AHSS steels. This comparison confirms that the flash processed steels (see Fig. 7) are indeed part of AHSS steel family. It is interesting to note that both UTS and the elongation are higher than the reported values for AHSS based on a martensitic microstructure.

Microstructure of Heat-Treated Samples

In this section, the underlying microstructure in original steel sheets before processing and after the FP and QT heat treatments are presented.

Initial Microstructure of the Annealed 8620 steel sheets

Optical microscopy of original steel sheets showed the presence of ferrite grains with carbides [see Fig 8a]. Most of the carbides were smaller than 1 μm and some of them were larger than 2 μm , suggesting a bimodal nature of carbide distribution. These carbides also appear to be aligned in between pan-caked ferrite structure suggesting that these could have precipitated during legacy thermo mechanical processing of these sheets. Scanning electron microscopy (see Fig. 8b) with back-scattered imaging showed that most of these carbides were associated with ferrite grain boundaries. Energy Dispersive X-ray Spectroscopy (EDS) analyses (See Fig. 8c) of these carbides show a higher ratio of Cr/Fe peak intensities compared to that of ferrite matrix. Based on the above analyses and thermodynamic calculations, the carbides are assumed to be Cr rich cementite (M_3C).

Microstructure of the Quench and tempered 8620 steel sheets

Since the QT and FP microstructures showed similar properties, the microstructure of the QT samples was characterized with optical microscopy and scanning electron microscopy. Due to the fine nature of the microstructure, optical microscopy (See Fig. 9a) could not discern the martensitic and bainitic microstructures. The scanning electron microscopy (see Fig. 9b) with back-scattered electron imaging revealed the martensitic lath structure with brightly-imaging carbides. By tracing the lath packet

boundary, typical prior austenite grain size was estimated to be around 20 to 30 μm . Due to the short duration of the austenitizing time (2 min at 1000°C), extensive austenite grain growth was not observed. A quantitative distribution of the austenite grain size in the overall sample was not performed due to the difficulty of identifying all the prior austenite grain boundaries by this technique. A back-scattered electron diffraction (EBSD) technique with crystallographic orientation analyses is required [23] and is the focus of the on-going work.

The volume fraction and size of carbides in the QT samples is smaller than that of the untreated samples (see Fig. 8b). It is possible that the microstructure may contain nanometer sized carbides that formed during the tempering (220°C for 4 minutes) stage. Usually the sizes of these carbides that form during early stages of tempering are below the resolution of scanning electron microscopy [24,25,26,27]. However, it is fair to conclude that initial carbides that are present in the un-treated samples have dissolved during austenitizing at 1000°C for 2 minutes.

Microstructure of the Flash Processed 8620 steel sheets

Since the focus of this paper is related to flash processing, extensive characterization was performed on FP samples. Optical microscopy [see Fig. 10a] of the FP samples again showed features that were difficult to distinguish. However, scanning electron microscopy with back scattered electron imaging revealed interesting features [see Fig. 10b]. Fine lath structures with brightly imaging particles could be observed. By tracing the lath packet boundary [see Fig. 10b], typical prior austenite grain size was estimated. This qualitatively shows that the prior austenite grain size of FP samples ($\leq 10 \mu\text{m}$) is much finer than that of the (20 to 30 μm) QT samples. This could be attributed to the rapid heating and cooling above the Ac3 temperature. In the next step, the composition of carbides was evaluated with EDS analyses. The measured spectrum and the calculated composition of analyzed regions are presented in Fig. 10c. Surprisingly, similar to the original steel samples before processing, these carbides were found to be rich in Chromium. In addition coarser and finer carbides were also observed which is similar to untreated samples. Since the FP process involves rapid heating (>

400°C/s) and cooling (> 3000°C/s) rates, these carbides are interpreted as the undissolved carbides that were present in the original base material.

With cursory observation of optical and scanning electron micrographs, one may conclude that the FP samples are predominantly martensitic. However, to substantiate this conclusion, detailed transmission electron microscopy was performed. A low magnification TEM image [see Fig. 11a] shows a prior austenite-austenite (γ/γ) grain boundary with presence of fine-scale martensitic laths with high dislocation density and coarser bainitic ferrite with slightly reduced dislocation density. The identification of bainitic ferrite was made based on the size of these plates and the presence of sheave like morphology with sub-units as shown by Bhadeshia [26].

In another region of the same TEM sample [see Fig. 11b], three bainitic (marked as 1, 2, 3 in the images) sheaves were observed. It is interesting to note that each and every individual sheave is made up of many ferrite sub-units with similar orientation in space. Electron diffraction analyses failed to identify any retained austenite film in between these sub-units [27]. Crystallographic orientation relationships between these bainitic sheaves were calculated and the summary of the results are presented in inset diffraction patterns in Fig. 11b. The sheaves #1 and #2 are indexed to be in the [111] BCC zone axis with relative misorientation of only 10 degrees. Assuming that these bainitic sheaves have KS/NW orientation relationship [28] with the parent austenite (FCC) phase, one can conclude that the bainitic sheaves #1 and #2 could have formed in a single austenite (γ_A) grain, within the regions that are separated by small angle boundaries. In contrast, the sheave #3 was indexed to be close to the [011] zone axis. Using the axis-angle pair analyses [29], the relative orientation between #3 and #1 or #3 and #2 was found to be of high-angle type. Using the same assumption of KS/NW orientation relationship of bainitic sheaves with austenite and symmetry relations [28], one can also conclude that the #3 sheave formed in adjacent austenite grain (γ_B) and the orientation relationship between γ_A and γ_B should be of high-angle type. Copious presence of such austenite grains with large angle boundaries will promote bainitic and/or

martensitic plates with wide varying orientations and may improve the properties [30]. It is stressed that the above results are typical and one cannot conclude similar crystallographic conditions throughout the FP sample. To evaluate this throughout the sample, we need to do extensive EBSD analyses and correlate the orientation distribution functions to the properties, which is the focus of the ongoing work.

In another FP sample, HAADF STEM image showed extensive carbide distributions. Some of them were associated with the ferrite sub-unit grain boundaries (See Fig. 12a). Size of these carbides was about 100nm. Elongated carbides between the lath boundaries are probably cementite. The precipitates are predominantly darker than the matrix suggesting that their average atomic number is lower than the iron matrix. However, the precipitate thickness effect is convoluted with the atomic number contrast and further work is required to establish the chemical composition of these precipitates. Some of the large carbides (marked by arrow) show no correlation to the ferrite boundaries (Fig. 12b). The EDS analyses show that these carbides were rich in silicon and chromium (Fig 12c). The detailed analyses of elemental distribution within these carbides can only be attained by atom probe field ion microscopy [24, 25]. Due the rapid rate of heating and cooling, the diffusion controlled dissolution and/or growth is expected to be sluggish. Therefore, these large carbides are interpreted as the un-dissolved carbides from the original steel sheets that existed before processing. From the above TEM analyses one can conclude that the FP microstructures contain bainitic ferrite, martensitic laths, as well as, uniform distribution of large and small carbides.

Evaluation of Mechanical Heterogeneity in Heat Treated Samples

Although, the FP samples showed this unique mixture of bainite, martensite and carbide microstructure, it is necessary to evaluate the heterogeneity of these microstructures within the overall sample. Since it is practically impossible to evaluate the microstructural heterogeneity through the series of optical, SEM and TEM techniques throughout the sample, hardness mapping was adopted as a measure of the heterogeneity [17]. In this method, series of hardness indents were made on the samples and the mechanical heterogeneity [31] is estimated by

analyzing the hardness distribution. The average size of the indents in the FP sample was $42\mu\text{m}$ which is bigger than the prior austenite grain size and also the bainitic and martensitic packet sizes. It is highly possible that the different microstructure distribution may provide similar hardness values. Therefore, the above measure has to be used only as a way to evaluate the possible scatter in final mechanical properties [32]. The mechanical heterogeneity results from QT and FP samples are discussed below [see Fig. 13].

The hardness maps from the cross-section of QT [Fig. 13a] sample showed a large soft region (blue color) in the edge1 region. In addition, the hardness variations are larger in the QT sample. This is attributed to the in-efficient quenching during manual insertion of the austenitized sample into the water. This is also reflected in the results of the tensile tests, where QT samples showed low yield strength [see Fig. 4a and Table II] and high strain hardening exponent. The hardness distribution curve from the QT sample is shown in Fig. 13b. The above data was interpreted by fitting multiple peaks to the distribution. The fitted peaks show that the QT sample can be divided into two regions, i.e., large ($v_{QT1} = 88\%$) region 1 with an average hardness of 475 ± 47 HV with large scatter and a small ($v_{QT1} = 12\%$) region 2 with an average hardness of 483 ± 5 with small scatter.

In contrast, the mechanical heterogeneity of the flash processed samples was less than QT samples [see Fig. 13b and Fig. 13c]. The image also shows that a slightly harder region (yellow color) is in the middle of the FP steel sheet compared to the edge of the steel sheet. The reader may recall the description of the FP process setup [see Fig. 1] in which the middle of the steel sheet gets exposed to the highest heat density, as a result highest heating rate and high peak temperature. Similar to QT sample analyses, the calculated hardness distribution was analyzed by fitting multiple peaks to the distribution. The fits show that the QT sample can be divided into three distinct regions, i.e., region 1 ($v_{FP1} = 14\%$) with an average hardness of 498 ± 7 , region 2 ($v_{FP1} = 74\%$) with an average hardness of 512 ± 11 and region 3 ($v_{FP1} = 12\%$) with an average hardness of 534 ± 9 HV. This presence of distinct and

homogeneous mixing of soft and hard regions in the FP sample is tentatively attributed to the improved mechanical properties of the flash processed steels.

Mechanism of Microstructure Evolution during Flash Processing

Based on the above results, one can conclude that the flash processing leads to a complex microstructure that contains martensite, bainite, and carbides. In addition the above complex microstructure also leads to a preferred mechanical heterogeneity that is better than the reference QT sample. In addition, the FP steels can be classified under the category of AHSS based on the strength observed. Furthermore, the flash processed steels show at least 7% higher UTS and 30% higher elongation than published martensitic advanced high strength steels [22] (see Fig. 7). However, these results have to be discussed in context of the mechanism for the microstructure evolution during flash processing.

On Heating– Carbide Dissolution

As per equilibrium thermodynamic calculations, the 8620 steels should be 100% austenite above 800°C. From the measured thermal cycle during the flash process, the peak temperature was 1100° C. Therefore, the sample should have completely transformed into austenite at this temperature and all the carbides should have dissolved. However electron microscopy showed that the carbides have not dissolved completely. This discrepancy is attributed to a very short dwell time in the austenitizing region which limits the dissolution of carbides. This hypothesis is supported by the observation of less number of carbides in the QT samples, which were austenitized for 2 minutes at 1000°C before quenching. It is well known that, extended time in the austenite phase field will promote the complete dissolution of carbides [33] by enhancing the substitutional element diffusion.

On Heating– Ferrite to Austenite Transformation

Another possibility of such unique microstructure evolution can be attributed to an increase in A_{c1} and A_{c3} due to rapid heating rates ($> 400 \text{ }^\circ\text{C s}^{-1}$) during flash processing. This hypothesis is in agreement with neural network analysis [34] and Gaussian process modeling [35]. The level of superheating required to nucleate and grow the austenite phase depends on the alloy composition, initial microstructure

and the heating rate. This increase in Ac1 temperature also been rationalized with the need to nucleate austenite phase from a ferrite + carbide microstructure [36]. On the other hand, if the initial microstructure contains retained austenite, there is no need for nucleation barrier [37]. Since the initial microstructure of 8620 steel sheets contained only ferrite and carbide, there is a need to nucleate austenite. Therefore, the rapid heating rate should lead to an increase in Ac1 above equilibrium value. Similar arguments can be made for an increase in Ac3 temperature due to reduced dwell time for the diffusion-controlled growth of austenite into the existing ferrite matrix. It noteworthy that the published research [34, 35] did not consider very high ($> 100 \text{ }^\circ\text{C s}^{-1}$) heating rates. Therefore, we need to consider the validity of the continued increase in Ac₁ and Ac₃ far above the equilibrium Ac₃ temperature. Recently, Elmer et al [38] have developed overall transformation kinetic model for ferrite to austenite transformation formation in 1005 steel based on synchrotron diffraction measurements during a heating cycle of a weld. The model was calibrated using the data from low heating rates below $100 \text{ }^\circ\text{C s}^{-1}$. However, the model calculations were extrapolated to higher heating rates. The calculations suggest a superheating of 100°C temperature or more above the equilibrium Ac₃ for the completion of austenite formation at heating rates above 300°C s^{-1} . This result supports the notion of nonequilibrium Ac₁ and Ac₃ temperature during flash processing (heating rate $> 410^\circ\text{C s}^{-1}$). Moreover, the SSDTA analyses of the heating curve (Figs. 5a and 5b) indicate that the Ac₁ temperature is definitely above 900°C ; 925.7°C (Fig. 5a) and 929.6°C (Fig. 5b). Similarly the Ac₃ temperature is calculated to be above $1000 \text{ }^\circ\text{C}$; 1045°C (Fig. 5a) and 1052°C (Fig. 5b), which is $200 \text{ }^\circ\text{C}$ higher than the equilibrium Ac₃ (806°C). Such a large increase in Ac₁ and Ac₃ temperature will also affect the carbide dissolution, since the presence of austenite is necessary for the initiation of dissolution. In addition, the time taken for austenite grain growth will also be reduced. This is supported by the small prior austenite grain size ($<20 \text{ }\mu\text{m}$) measured in the flash processed samples.

As a result of small dwell time above Ac₃ temperature, the carbon diffusion and redistribution within the austenite grain is expected to be incomplete [39]. Hence the

steel that that is being cooled from peak temperature will have inhomogeneous carbon distribution within the austenite matrix, which could trigger complex decomposition of austenite into different ferrite morphologies and martensite.

On Cooling – Austenite Decomposition

Based on the calculated CCT diagram and cooling rate ($> 3000 \text{ }^\circ\text{C}\cdot\text{s}^{-1}$) during flash processing, the steel should have transformed to 100% martensite. To evaluate the effect of carbon concentration gradients within the austenite on the final microstructure, the CCT diagrams for 8620 steels with different carbon concentrations ranging from 0.03 to 0.21 wt.%, with the same substitutional alloying element concentrations, were calculated using the methodology described earlier [16, 17]. These series of CCT diagrams are also overlaid with the measured cooling curve [see Fig. 14]. The comparison shows that even with very low carbon content in the austenite the microstructure should be 100% martensite. However, the microstructure contains bainitic ferrite, martensite and carbides.

This discrepancy is attributed to the inadequacies of transformation kinetic modeling methodologies. These models were developed using the TTT data generated by traditional heat-treatment conditions [16, 20] involving slow heating and isothermal hold at austenitizing temperature exceeding 5 minutes, and homogenous austenite. The extension of these models to austenite with inhomogeneous carbon concentration may not be reliable. These situations are similar to the rapid heating and cooling conditions experienced by heat-affected-zone of welds.

Based on the above results and discussions, a hypothetical microstructure evolution during flash processing is proposed [see Fig. 15]. At room temperature, the initial microstructure contains ferrite and carbides [Fig. 15a]. On rapid heating, the austenite nucleation is expected to occur at the carbide-ferrite boundaries [36] and the austenite is expected to grow at the expense of ferrite and carbide dissolution. Due to rapid heating rates, the temperature at which the austenite nucleation and growth occurs may be elevated to higher temperatures ($> 900^\circ\text{C}$) [see Fig. 15b]. But before the complete dissolution of cementite, the cooling cycle starts. At this juncture, the austenite is expected to have inhomogeneous distribution of carbon

content [Fig. 15c]. With the on-set of rapid cooling, the carbon enriched regions may transform to martensite and carbon-depleted regions may transform to bainite or low-carbon martensite. This leads to a final microstructure with complex distribution of bainite, martensite and carbides [Fig.15d]. This hypothetical mechanism suggests that the initial microstructure will have strong effect on the final microstructure and properties achieved by flash processing. Ongoing experiments are focusing on evaluating these effects and methods to track the phase transformations during heating and cooling for wide range of steels to prove or disprove this hypothesis.

Summary and Conclusions

A rapid heating ($>400^{\circ}\text{C}\text{s}^{-1}$) and cooling ($>3000^{\circ}\text{C}\text{s}^{-1}$) flash process has been developed that produces a microstructure with good combination of yield (1280MPa), ultimate tensile strength (1600MPa) and appreciable ductility of up to 9.9%. . The above strength and ductility levels are significantly better than martensitic-based AHSS (1400-1500 MPa UTS). Based on the above data, the flash processed steels can be classified as part of the advanced high strength steel family. As the overall processing time (< 10 seconds) is very short, this process could be an alternative route for producing AHSS sheets.

The underlying microstructures in flash processed samples were characterized with optical and analytical electron microscopy. The characterization results showed that the steel contains bainite, martensite and carbides. The hardness map analyses showed that the flash processed samples contain a minor fraction of soft microstructure interspersed within a harder microstructure. This unique distribution of soft and hard microstructure is correlated with the improved YS, UTS and ductility.

Traditional phase transformation models indicated that, for the cooling rates measured in the flash process, the microstructure should be 100% martensite after processing. This discrepancy is addressed with a theoretical mechanism that involves (a) an increase in A_{c1} and A_{c3} due rapid heating rate, (b) incomplete

dissolution of carbides, and (c) decomposition of austenite with non-uniform carbon concentration.

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Tables

Table I: Composition of AISI8620 steel.

Element	Fe	C	Si	S	P	Mn	Cr	Ni	Co	Mo	Cu
Wt.%	Balance	0.21	0.27	0.002	0.009	0.73	0.48	0.48	0.007	0.156	0.178

Table II: Measured engineering tensile properties of FP (from previous and current work) and QT samples (current work). The tests were performed according to ASTM E8 specifications

Sample ID	Ultimate Tensile Strength (MPa)	0.2% Yield Strength (YS)(MPa)	Reduction of Area (%)	Elongation (%)	Ratio of YS to UTS	Source
Ref-FP1	1685.77	1241.75	-	6.3	0.737	Ref. 10
Ref-FP2	1694.04	1314.83	-	7.3	0.776	Ref. 10
Ref-FP3	1676.12	1275.53	-	7.1	0.761	Ref. 10
Ref-FP4	1669.91	1292.77	-	6.9	0.774	Ref. 10
FP#1	1664.1	1442.8	39.4	8.8	0.867	Current
FP#2	1619.3	1386.9	38.0	9.9	0.856	Current
FP#3	1520.7	1300.0	38.3	9.9	0.855	Current
QT#1	1607.6	1333.8	37.4	6.8	0.829	Current
QT#2	1657.9	1464.1	49.8	10.0	0.883	Current
QT#3	1642.8	1487.6	14.8	4.3	0.905	Current

Sample dimensions used in Ref 10: width 13 mm; thickness 1.5mm

Sample dimensions in the current work: width 6.4 mm; thickness 1.23mm

Table III: Calculated pre-exponent (K) and strain hardening exponent (n) derived by fitting the equation (1) to the plastic regime of the true stress-strain curve.

	FP#1	FP#2	FP#3	QT#1	QT#2	QT#3
K (MPa)	569.6	622.1	613.7	1305.6	536.3	543.9
n	0.1542	0.1568	0.1348	0.1950	0.1342	0.1471

Figure Captions

Fig. 1 Schematic illustration of the experimental set-up of the flash processing technique.

Fig. 2a Calculated variation of carbide volume fraction as a function temperature obtained using ThermoCalc ^[14] and TCFE5 database.

Fig. 2b TTT and CCT diagrams showing reconstructive transformation and displacive transformation Transformation regions.

Fig. 3 Heating/Cooling rate (a) & Temperature variation (b) Vs. Time in a typical FP.

Fig. 4 Thermal profile of two samples showing repeatable thermal cycles.

Fig. 5a SSDTA analysis of the heating region of sample 1 showing two endothermic thermal events.

Fig. 5b SSDTA analysis of the heating region of sample 2 showing two endothermic thermal events.

Fig. 6 Comparison of true stress Vs. true strain of Q&T and Flash processed specimen.

Fig. 7 Overview of published tensile strength and ductility that can be achieved for wide range of advanced high strength steels [22] and measured tensile strength and ductility of flash processed steels.

Fig. 8(a) Optical image-Microstructure of 8620 un-processed sample.

Fig. 8(b) SEM -Microstructure of 8620 un-processed sample.

Fig. 8(c) EDS- Comparison of matrix and particle EDS signals of the 8620 un-processed.

Fig. 9(a) Optical Image- Q&T 8620 sample.

Fig. 9(b) SEM- Microstructure of 8620 Q&T sample with a prior austenite grain boundary.

Fig. 10(a) Optical Image- Flash processed 8620 sample.

Fig. 10(b) SEM- Microstructure of 8620 Flash processed sample.

Fig. 10(c) EDS- Comparison of matrix and particle EDS signals of the Flash

Fig. 11(a) TEM image showing a g/g grain boundary with parallel laths growing out of it.

Fig. 11(b) TEM Image showing three bainitic sheaves (1, 2, 3) in FP sample.

Fig. 12(a) Carbides in between ferrite sub-units and (b) high magnification of the same.

Fig. 12(c) EDS showing chromium enrichment.

Fig. 13(a) Hardness contour map of a cross section of Q&T sample.

Fig. 13(b) Hardness histogram of Q&T sample.

Fig. 13(c) Hardness contour map of a cross section of FP sample.

Fig. 13(d) Hardness histogram of FP sample.

Fig. 14 CCT curve of the AISI 8620 showing the formation of martensite for the flash cooling rate.

Fig. 15 Schematic Illustration of microstructure evolution: a) Un-processed ferrite with carbides, b) Growth of austenite and start of carbide dissolution c) Inhomogeneous austenite with reduction in carbides d) Martensite with bainitic plates formed from inhomogeneous austenite.