Weldability Investigations of Advanced High Strength Steels Produced by Flash Processing

THESIS

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By

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Abstract

Recently, a new heat treatment process (flash processing) has been shown to take low-alloy steel and create steels with advanced high strength steel (AHSS) level properties (ultimate tensile strength> 1600MPa and ductility>8%) that are better than most available martensitic AHSS. This steel has also shown better ballistic protection capability than currently available high hard, rolled homogenous, and titanium armors. The unique process works by rapid induction heating into the austenite phase field and subsequent quenching in less than ten seconds. The resultant mixture of carbides, bainite, and martensite allows for the mechanical properties it achieves. However, the steel has not been evaluated for its weldability, which could limit both the ballistic and mechanical property advantages that it has over currently used materials. A well documented decrease in strength, ductility, and toughness in the welds occurs when the heat-affected zone (HAZ) becomes softer compared to its initial state. This has been well documented for AHSS, armor materials, and other thermo-mechanically processed steels.

The goal of this research is to examine the effect of different welding conditions on flash processed steel microstructure and resultant mechanical properties. Low heat input gas metal arc welding (GMAW) was performed as an initial process that is typically done for joining armor steels. A comparison to a currently used high hard armor steel showed lower hardness and a larger softened region softening in the flash process HAZ.

To further understand the microstructure evolution for both steels, HAZ physical
Simulations were conducted to examine the effect specific peak temperatures for a given heating and cooling rate. It was found that flash processed steel softened to 170HVN (from base metal hardness of 540HVN) when intercritically heated close to the $A_{c3}$ temperature. High hard steel softened to 290HVN (from base metal hardness of 530HVN) when heated below the $A_{c1}$. The reasoning was found that flash process steel transforms to the soft ferrite microstructure when intercritically heated with more ferrite forming as the more of the initial material is transformed to austenite. High hard transforms only to martensite when heated above the $A_{c1}$, therefore softening only when the original martensitic microstructure is tempered. The main difference between these effects is the initial microstructure and initial alloying additions in the composition (3.5wt% for high hard compared to 1.8wt% for flash process) that stabilized austenite for the high hard material.

To reduce the softening that is deleterious to ballistic properties, high power laser welding was used to reduce the heat input from 19kJ/in with GMAW to 4kJ/in. This resulted in the increase of the softest hardness in flash processed steel to 300HVN and 340HVN for high hard steel. The width of the softened region was also reduced from over 25mm using GMAW to 6mm. The softened microstructure was that of tempered initial microstructure for both steels.

To examine the effect of these welding conditions, tensile testing was performed. It was found that the tensile strength for flash process reduced from 260ksi in the as-received material to 120ksi when using GMAW and 215ksi when laser welding. The ductility also showed a decrease from 14.8% in the base metal to 9.8% when using
GMAW and 6.9% when laser welded. Similar reductions in strength and ductility were seen in high hard armor. The failure of both welding conditions occurred in the softened region and fractography showed the presence of quasi cleave and small ductile dimples in the fracture surface with deeper ductile dimples present in the GMAW.

Preliminary ballistics testing showed a successful protection from a NIJ level III threat 10mm from either side of the laser weld centerline, while this distance was 22mm from either side of the gas metal arc weld centerline (i.e. failure of the test occurred within a 20mm window of the laser weld centerline and within a 44mm window of the GMAW centerline).
Dedication

This document is dedicated to my parents who have always supported and encouraged me through all endeavors in life.
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Chapter 1: Introduction and Motivation

With the increasing concern over energy efficiency, the transportation industry is trying to find ways to reduce energy consumption to make transportation more economical and environmentally friendly. One solution involves the lightweighting of materials. The automotive industry has been active in this regard for over the last decade [1, 2]. A consortium entitled Ultra-Light Steel Automotive Body-Advanced Vehicle Concept (ULSAB-AVC) Consortium has set their main task to lightweighting through the development of Advanced High Strength Steels (AHSS).

Along similar circumstances, the military has also been trying to reduce weight to improve survivability, expense, and maneuverability. Recent reports published by the National Research Council have shown developments in this regard for land, air, and sea based vehicles [3, 4]. While lower density materials such as titanium and aluminum are often used in aircraft, steel is still the predominant usage material in both ships and land-based vehicles such as tanks, trucks, and infantry fighting vehicles. The extensive knowledge of steel fabrication, availability of material, and material cost are among the main reasons for using steel in armor applications [5]. In addition, the ballistic worthiness extends to a wide range of threat levels that alternative materials have not matched [6].
Recently, a new heat treatment process has been developed to create AHSS level properties from low-carbon and low-alloy steel [7-9]. The process uses a rapid heat treatment by continuously feeding steel sheet through an induction or flame heating unit to heat the steel above the A₃ temperature and rapidly quenching with a quench bath. The result is an unconventional microstructure of carbides, martensite and bainite. The properties achieved in flash process are displayed on a plot developed by Zrnik and modified by Lolla et al. [8, 10] in Figure 1.1. This process has also shown ballistic properties that exceed those of currently available armor materials.
The development of better armor steels in recent history has shown the increasing difficulty to form and weld [11]. This results in the armor materials being unusable for structural applications and being added on for protection, adding to the weight of the overall vehicle. An effort to incorporate the weldability of these steels is of importance to be able to reduce the weight and cost of the overall vehicle. The ability to retain strength, ballistic worthiness, and toughness after welding will allow considerable advances in armor design [3].

Since the deployment of future AHSS, armor steel, and in particular, flash processed steels is highly dependent upon the effect of welding, it is crucial that these
steels be weldable and all regions affected by welding meet the properties necessary for their applications. One particular deficiency occurring during welding is the effect of softening in the heat-affected zone (HAZ), as seen in pipeline, armor, automotive, and other thermo-mechanically controlled processed (TMCP) steels [12-20]. The softening usually corresponds with decreased strength, toughness, and ductility. In addition, softening results in a decrease in ballistic properties [11, 21, 22]. Published research has not been performed on welding’s effect on flash processed steels. Thus, this document will present the various welding processes and simulations used to determine the effect of welding on this material. This includes baseline fusion welding processes that are currently used in addition to physical simulations concurrent with the conditions seen in these typical welding conditions. In addition, other welding processes will be explored for their potential to try and maintain base metal properties during welding.

The layout of this thesis will include a background literature review that discusses important topics that pertain to the research performed. The objectives for the research are discussed in chapter three while the experimental procedures are emphasized in chapter four. Chapter five contains all of the results and discussions of the research. First, the base metal properties for the two steels are discussed in terms of hardness, microstructure characterization, and mechanical properties measured by tensile testing. Second, the baseline gas metal arc welding conditions are discussed in similar terms as the base material properties along with fractography of the tensile tested samples. The third part of this section goes into depth on the phase transformation analysis of single-pass heat-affected zone physical simulations. Dilatometry, hardness, and characterization
efforts are the primary focus of this section to describe the effect of different HAZ thermal cycles on both materials examined. The fourth section of the fifth chapter discusses the use of high energy density (HED) laser welding and the subsequent effect on the hardness and microstructure of the base materials. The same analysis techniques used in the GMAW are used with the HED testing conditions. Finally, a summary of the results and primary conclusions are featured in chapter six. The seventh chapter discusses the potential for future work. This includes areas that have some preliminary results including multi-pass HAZ physical simulations and ballistic testing. Further testing to verify HED welding effect on HAZ properties is proposed using similar techniques to the single and multi-pass HAZ physical simulations. In addition, toughness evaluation has been proposed to understand its role in these steels’ applications.
Chapter 2: Background

2.1. Flash Processed Steel

Initial development of flash process steels looked at a wide range of low carbon and low alloys steels such as 1020, 1040, 8620, and 4130. Initial data showed improvement in strength and ductility compared to as-quenched and quenched and tempered conditions with similar composition.

2.1.1. Process Set-Up

A schematic setup of the process is seen in Figure 2.1 [8]. The steel sheet is fed at a constant rate through either vertically or horizontally through guide rollers. The rollers direct the steel into the heating element that is either made of several oxy-propane flames or an induction heating element where the steel is heated above the A₃ temperature. A few millimeters below the heating element, an agitated and chilled water quench bath cools the sample rapidly. To avoid steam from rising in the vertical position and creating heating discrepancies, a graphite separator film is used between the heat source and quench bath. Infrared pyrometers are used within the assembly both for temperature monitoring and control. The distance between the heating element and quench bath can be changed in addition to the feed rate to better control heating and cooling rates, peak temperature, and time at peak temperature [9].
2.1.2. Thermal Cycle Analysis

Cola first used pyrometers to measure the temperature at fixed points for the primary reason of process control [7]. In order to monitor the thermal cycles better, Lolla et al. used type K thermocouples attached to the inner diameter a pipe to monitor the thermal cycle during processing [8]. The heating and cooling rates are seen in Figure 2.2a and the recorded thermal cycle in Figure 2.2b. The heating rate maximum of 410°C/s is reached near 780°C between the A₁ and A₃ temperatures. The peak temperature is near 1100°C, well into the austenite phase field. The close proximity of the quench bath allows the cooling rate to be over 3000°C/s. The thermocouple data agrees with normal steel phase transformation analysis that the microstructure fully
austenitized and was quenched fast to avoid any reconstructive transformation and create a fully martensitic microstructure.
Figure 2.2: Thermal Profiles from 8620 flash processed steel. (a) heating and cooling rates achieved (b) Temperature vs. time plot [8]

In order to determine if the sample was completely austenitized, the thermocouple data was processed using a technique called Single-Sensor Differential Thermal Analysis
(SSDTA) [23, 24]. In this technique, a calculated reference curve is used to analyze against the measured thermal profile of the sample in question. The process uses a reference sample that is made by analytical heat flow conditions similar to that of the sample and contains no enthalpy changes. Deviations from this curve will mark phase transformation phenomena on-heating and cooling based on the enthalpy change. In the work by Lolla [9], they investigated AISI 4130 pipe with these conditions. The result on-cooling showed two separate deviations from the reference curve, marking two phase transformations as seen in Figure 2.3. The temperatures and resultant microstructure evaluation agree that martensite and bainite are both present. On heating data presented for 8620 steel showed an increase of both $A_{c1}$ and $A_{c3}$ temperatures. The increase was over 200°C compared to equilibrium calculations. This increase in critical temperatures has shown to be in agreement with Gaussian process modeling and neural network analysis [25, 26].
2.1.3. Initial Composition and Microstructure Before and After Flash Processing

Lolla et al. report using both AISI 8620 and AISI 4130 as the initial microstructure in their in-depth research of the process [8, 9]. The initial 8620 microstructure obtained had significant amount of spheroidized carbides as seen in Figure 2.4a-b. The composition of the 8620 material is seen in Table 2.1 with low alloying additions of Cr, Ni, Mn, Si, and Mo. After flash processing, the presence of the carbides remained from the initial microstructure in addition to the creation of martensite and bainite when rapidly cooled as seen in Figure 2.4c-d. The mechanism will be explained
shortly, but this goes against common steel metallurgy knowledge of phase
transformations [27]. Transmission electron microscopy (TEM) was used to identify that
both bainite and martensite were present after processing as seen in both Figure 2.5 and
Figure 2.6. Sheaves of ferrite subunits with cementite between these subunits were
observed that are characteristic of upper bainite [28]. Lower bainite, characteristic with
carbides present within the ferrite where also observed during TEM observation. When
energy dispersive x-ray spectroscopy was performed on the matrix and carbides, it was
found that the carbides in both conditions were enriched with chromium.
Figure 2.4: (a) Optical microscopy picture of 8620 before flash processed (b) SEM image of 8620 before flash processed (c) Optical microscopy of 8620 after flash processed (d) SEM image of 8620 after flash processing (adapted from [1, 2])

Table 2.1: Composition of 8620 [9]

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>Ni</th>
<th>Co</th>
<th>Mo</th>
<th>S</th>
<th>P</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Composition</strong></td>
<td>0.21</td>
<td>0.27</td>
<td>0.73</td>
<td>0.48</td>
<td>0.48</td>
<td>0.007</td>
<td>0.156</td>
<td>0.002</td>
<td>0.009</td>
</tr>
</tbody>
</table>
Figure 2.5: Austenite-Austenite grain boundary with martensite formed in one grain, and parallel sheaves of bainite growing from the other [8]
Figure 2.6: TEM micrograph with three bainitic sheaves growing from an austenite grain boundary and corresponding diffraction pattern for ferrite [8]

2.1.4. Mechanism

The explanation for the observed bainite when well-established principles dictate there should only be martensite is explained by a couple reasons. First, the initial Cr-enriched carbides are not fully dissolved when heated to the peak temperature in the austenite phase field as the equilibrium thermodynamics dictates. In addition, the time spent in the austenite phase field is not long enough for full partitioning of the carbon homogenously throughout the material. The sluggish dissolution of the Cr-enriched M_3C carbide limits the austenite Cr and C content. Recent simulation work [29] confirms the
sluggish dissolution and predicts the ability for bainite to form from regions of austenite near these carbides. However, these results have not been experimentally proven besides the end result seen after flash processing. A schematic of the microstructure evolution developed by Lolla et al. is seen in Figure 2.7 and described here. Initially in region 1 (a), there is the initial microstructure of ferrite and carbides that is starting to be heated by conduction from the steel closer to the heating source. In region two (b), the formation of austenite begins at carbide-ferrite interfaces and rapidly grows toward the ferrite as opposed to the cementite that is seen in pearlitic steels [25, 30]. In region three (c), full austenization is realized for a short duration with carbides not decomposing fully as would be expected. Finally after the quenching in region 4 (d), the microstructures of bainite and martensite are seen along with the carbides that were initially present.
2.1.5. Mechanical Properties

2.1.5.1. Tensile Strength and Ductility

The strength and ductility of flash processed 8620 from Cola and Lolla et al. is listed in Table 2.2 and the 8620 flash processed samples are plotted on Figure 1.1. The strength and ductility are consistently higher than those presented by Zrnik. However, there has been considerable debate between testing methods and accuracy of reporting of this data that uses ASTM standards and those used in other countries [31]. However, the experiment shows results that have FP material with better strength and ductility compared to the quenched and tempered (QT) samples.
Table 2.2: Tensile Test Results of FP 8620 and quenched and tempered 8620[8] Ref-FP are from Cola [7]

<table>
<thead>
<tr>
<th>Sample ID</th>
<th>Ultimate tensile strength, MPa</th>
<th>0.2% yield strength, MPa</th>
<th>Reduction of area, %</th>
<th>Elongation, %</th>
<th>Ratio of YS to UTS</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ref.FP1</td>
<td>1685.77</td>
<td>1241.75</td>
<td>...</td>
<td>6.3</td>
<td>0.737</td>
</tr>
<tr>
<td>Ref.FP2</td>
<td>1694.04</td>
<td>1314.83</td>
<td>...</td>
<td>7.3</td>
<td>0.776</td>
</tr>
<tr>
<td>Ref.FP3</td>
<td>1676.12</td>
<td>1275.53</td>
<td>...</td>
<td>7.1</td>
<td>0.761</td>
</tr>
<tr>
<td>Ref.FP4</td>
<td>1669.91</td>
<td>1292.77</td>
<td>...</td>
<td>6.9</td>
<td>0.774</td>
</tr>
<tr>
<td>FP no. 1</td>
<td>1664.1</td>
<td>1442.8</td>
<td>39.4</td>
<td>8.6</td>
<td>0.867</td>
</tr>
<tr>
<td>FP no. 2</td>
<td>1619.3</td>
<td>1386.9</td>
<td>38.0</td>
<td>9.9</td>
<td>0.856</td>
</tr>
<tr>
<td>FP no. 3</td>
<td>1520.7</td>
<td>1300.0</td>
<td>38.3</td>
<td>9.9</td>
<td>0.855</td>
</tr>
<tr>
<td>QT no. 1</td>
<td>1607.6</td>
<td>1333.8</td>
<td>37.4</td>
<td>6.8</td>
<td>0.829</td>
</tr>
<tr>
<td>QT no. 2</td>
<td>1657.9</td>
<td>1464.1</td>
<td>49.8</td>
<td>10.0</td>
<td>0.883</td>
</tr>
<tr>
<td>QT no. 3</td>
<td>1642.8</td>
<td>1487.6</td>
<td>14.8</td>
<td>4.3</td>
<td>0.905</td>
</tr>
</tbody>
</table>

From Lolla’s Thesis [9], they justify this strength using strengthening models developed by Tomita and Okabayshi [32] and Young and Bhadeshia [33]. The trend Tomita and Okabayshi found showed the increase in strength of steel as the volume of bainite increases against the simple rule of mixtures in Figure 2.8. The main reasons for this increase in strength differ between Tomita and Young slightly, but they agree that there is a plastic restraining effect of the martensite on the bainite. The difference between the two models is that Young and Bhadeshia attributed much of the strengthening to carbon portioning between the bainite and martensite with the bainite rejecting carbon making the martensite stronger with increased carbon content. Tomita and Okabayashi stated that there was refinement of the austenite grain size due to the growth of the bainite making the martensite stronger due to a Hall-Petch type relationship. These considerations allow the ability for flash process to obtain the
strength and ductility that is seen from characterization, thermal analysis and tensile testing.

Figure 2.8: Lolla compares Okabayashi and Tomita data to the rule of mixtures method showing increased strength from increasing bainite fraction in martensitic steel [9]

2.1.5.2. Hardness

The hardness of flash processed steels shows a bimodal and trimodal distribution in 4130 and 8620 steels, respectively (Figure 2.9) [9]. The distribution of the hardness is attributed to the microstructure of softer bainite and harder martensite that correlate with equations based on composition seen in the thesis by Lolla [9]. Back-calculating the carbon content and assuming uniform concentration showed good correlation with
expected bulk concentration giving another confirmation of the mixed bainite and martensite microstructure.

Figure 2.9: Hardness Histograms for FP 8620 (top) showing trimodal distribution in hardness and 4130 showing bimodal distribution in hardness (bottom)

2.2. Armor Steel

As mentioned in the introduction, armor steel has been the traditional material for land and sea based vehicles because of its well known protection against multiple threats,
established fabrication methods, lower material costs, relative ease of repair in the field, and large commercial availability [3, 5, 6]. In recent years, lightweighting has become a primary concern due to the increased savings and desire to create better materials than those currently available [3, 4].

2.2.1. Processing and microstructure

Prifti et al. give a good history of the development of modern day armor steel [22]. The general trend was to maintain fabrication ability while increasing the protection level. The method for doing so was increasing the purity of the cast steels [34-36] and increasing the hardness from 30 HRc to 45HRc by increasing alloying content, specifically nickel, chromium, molybdenum, and boron [22]. The reason for increasing hardness is the reporting of multiple sources showing a correlation between higher hardness (up to 52 HRc) and better ballistic properties [11, 21, 22]. Recent development has seen further increase in hardness of high hard armor (HHA) and very hard armor with hardness above 500 and 600 Brinell Hardness, respectively [5, 11]. These steels are completely martensitic and tempered after processing, but their high carbon contents leads to difficulty in welding and forming, as discussed later in this chapter. The main code that dictates the compositional limits and hardness range on armor steels is MIL-DTL-12560J for wrought homogenous armor steels such as RHA. MIL-DTL-46100E is another code for quenched and tempered, higher hardness, materials and used for thinner thickness armors such as HHA. The creation of newer, ultra high hard steels that have a hardness in the range of 515-650BHN are outside the scope for both these standards;
however, a current standard is being developed for them because of the improvement against ballistic threats [37].

The two main military codes, MIL-DTL-12560J and MIL-DTL-46100E, both have separate processing conditions for specific properties. In MIL-DTL-12560J, two different heat treatments, high and low temperature, are used. First, a low tempering temperature (177°C) that provides high strength and hardness for the ballistic performance at the sacrifice of ductility and toughness is implemented. The second tempering temperature, at higher temperatures (427°C), provides higher ductility and toughness for shock resistance at the expense of lower strength and hardness. Jena et al. did experimentation on the effect of tempering temperature which showed for a constant tempering time of 2 hours, lower temperatures (100°C-300°C) had much better resistance to bullet penetration through a given 25mm thickness as seen in Figure 2.10 [38]. The main reasoning for this increase in properties is attributed to allowing sufficient time to relieve quenching stresses and allow recovery of dislocation structures within the material. However, with increased temperatures, coarsening of the martensite laths by precipitation of cementite and potential tempered martensite embrittlement occurs with the increasing temperatures [38, 39].
A debate in the armor steel literature revolves around the effect of retained austenite. It is reported by Jena et al. [40] that retained austenite leads to a decrease in armor resistance as evidenced in increased retained austenite from 2.46% to 7.43%. Lower hardness regions attributed to retained austenite were found near those regions that failed in the higher retained austenite samples. However, Maweja et al. [41-43] reports that retained austenite benefitted ballistic properties via nature of deformation transformation to twinned martensite when impact occurred during ballistic testing. X-ray diffraction (XRD) showed that there was systemic decrease in the before and after ballistic test, respectively [41]. The difference in these views could be on the location of the retained austenite in relation to the ballistic event. If the stress is not high enough far away from the impact, or too close to the surface and experiences high temperature strain event from the ballistic shot, the retained austenite could not change to martensite and
still be a potential location for crack initiation or propagation. However, if the retained austenite is located in the material in such a way that it can twin to form martensite as observed by Maweja [42], the retained austenite could be beneficial.

2.2.2. Mechanical Properties

This section is meant to discuss the wide reporting in literature on the usefulness of armor steel for survivability (e.g. ballistic resistance and shock resistance) in relation to standardized structural integrity properties (e.g. strength and hardness).

As mentioned previously, hardness is one of the most reported properties for the correlation of increased hardness with ballistic performance [5, 37, 44-47]. Gooch and Showalter [5, 37, 44] investigated the new Ultra High Hard armors (UHH) that showed the response of increased hardness from MIL-DTL-46100E (477-534 BHN) to 570-640HVN showed an increase of upward 20% based on $V_{50}$ velocity measurements (Figure 2.11) [37]. The $V_{50}$ velocity is defined as the projectile speed that results in complete perforation of the armor steel 50% of the time. Dikshit showed the increased benefit of armor hardness to the velocity at which plates were perforated (Figure 2.12). However, there is no discussion on the actual microstructures or condition of the material besides hardness. However, more recent reports have shown that hardness may just be a correlation and nothing more [21, 41-43]. Demir studied the effect of AISI 4340 and DIN 100Cr6 with varying hardness for both materials. He found that increased hardness to about 50HRc resulted in better ballistic protection, but above this threshold the protection decreased. The rationalization was that although the increasing
hardness level reduced the projectile from progressing through the plate (decreasing the perforation ability) the toughness was not high enough to handle the impact load and catastrophic failure occurred through the plate by gross crack formation [21]. Maweja et al. found similar findings and also proposed that the low ratio of yield strength to ultimate tensile strength has a positive effect on ballistic properties in their experimentation [42, 43].

![Graph showing ballistic performance comparison]

Figure 2.11: Increase in ballistic performance when comparing UHH steel to RHA and HHA counterparts of MIL-DTL-46100E [37]
Figure 2.12: Dikshit's findings of increased hardness benefitting ballistic resistance to plate perforation velocity (bullet speed after it goes through the steel) [46]

Another aspect of hardness to consider is that of the projectile compared to the base metal. A study by Anderson showed that if the projectile was harder than that of the base metal, reduction in the $V_{50}$ (Anderson uses $V_{BL}$) decreases by more than 50 m/s (Figure 2.13) [48].
From most of the findings, normal strain rate conditions from tensile testing and Charpy toughness tests have led to inconclusive results. A test developed by Hopkinson aptly named the split-Hopkinson bar test (Figure 2.14) analyzes materials at high strain rates higher than $10^2 \text{s}^{-1}$ [49]. Various researchers have used this method to help understand the dynamic fracture toughness of armor materials when ballistic arrangements cannot be easily completed [50-53].

*Figure 2.13: Ballistic limit velocity ($V_{BL}$) compared to projectile hardness [48]*
Most of the background discussed this far involves the material as it is originally made, without discussing the effect of welding. However, it is important to know the complex interactions that occur during the process and final state of the material in order to effectively use the material in a manufactured product. This will be discussed in the next section.

2.3. Welding of Steels

2.3.1. Steel Welding Metallurgy

Welding metallurgy is normally associated with both the fabrication and service life characteristics. During fabrication, the chemical composition, thermal cycle interactions, and inherent process (e.g. arc-welding, high energy density, solid-state) are
major factors [54]. The complex thermal gradients, solidification and segregation, and heat flow into the surrounding heat affected zone (HAZ) causing austenite grain growth are of great importance to the initial part and service life of the end product [55, 56].

The main regions of a weld consist of the fusion zone, HAZ, and base metal (Figure 2.15). The fusion zone can either be autogenous or have filler metal added. The HAZ contains four main regions as follows. (1) The coarse grain HAZ (CGHAZ), where peak temperature is well above the Ac3 temperature (T_p >> Ac3). Under these conditions, full transformation of ferrite to austenite occurs with extensive austenite grain growth. (2) The fine grain HAZ (FGHAZ), where the peak temperature is just above the Ac3 (T_p > Ac3). Under these conditions, full transformation of ferrite to austenite occurs with minimal austenite grain growth. (3) The intercritical HAZ (ICHAZ) region for which, the peak temperature is between the Ac1 and Ac3 (Ac1 < T_p < Ac3). Under this condition, partial transformation of ferrite to austenite occurs. (4) The subcritical HAZ (SCHAZ), where the peak temperature is below the Ac1 (T_p < Ac1) and no measurable transformation to austenite is observed. As stated in the background to flash processed steel and evidenced above, the condition in which the austenite is in before cooling has a fundamental determination of the final microstructure.
2.3.2. **Welding Armor Steels**

Since armor materials generally have an initial high hardness, they are prone to softening in the heat-affected zone that results in reduction of ballistic and mechanical properties. There have been several studies on the HAZ softening of armor steels that show the reduced ballistic limit in the softened region [17, 18, 20, 58-60]. Mohandas et al. researched several armor steels with varying alloying concentrations and correlated higher carbon equivalency with increased softening potential [17]. This is primarily due to the tempering of the martensite present that creates enriched carbon cementite, while the martensite transitions to ferrite [61]. Detailed discussion of the softening mechanism will be discussed in the next section. Reddy et al. conducted several experiments where they investigated the effects of welding processes and heat input on softening and ballistic properties. Figure 2.16 shows the relation of shielded metal arc welding (SMAW) heat input on the hardness across the weld area [18]. As anticipated, the

---

**Figure 2.15**: Regions of steel weld from Caron et al. [57]
increased heat input results in a larger softened region due to increased heating of the metal outside of the fusion zone. In the same paper, they correlated the ballistic effectiveness with these different regions (Figure 2.17). It is important to note the increased ballistic performance of this material with the lower heat input. Unfried et al. modeled the softening effects and compared them experimentally by analyzing the microstructure of SMAW conditions. Using ThermoCalc and DICTRA, and heat flow software AC₃, he showed good agreement between the model and experiment within 5HVN in most regions (Figure 2.18) [20].

Figure 2.16: Effect of heat input on HAZ softening in shielded metal arc welds adapted from Reddy et al. [18]
Figure 2.17: Ballistic behavior in different regions of a welded structure [18]
Several researchers have also examined the role of welding processes and filler metals on mechanical properties (strength and toughness [62-64]), ballistic properties[60, 65-67], and hydrogen cracking potential [68-70].

Ade has shown that pulse spray is the best transfer method for GMAW due to increased sidewall penetration that increases the effective angle and toeline extension (defined in Figure 2.19) that correlate with reduced cracking susceptibility (Figure 2.20) [65]. He also reports that armor steel welds are usually designed in such a way to hide them from ballistic attack. However, they must be able to survive shock testing. Those welds that are deemed “ballistic” must pass a test where they can crack no more than 15” from a 75mm projectile being fired in an h-plate configuration (Figure 2.21). Various tests are made to validate the welding before it is subjected to a shock test in an H-plate configuration. These tests include the y-groove restraint and simulated H-plate that both

Figure 2.18: Hardness comparison of experimentally made SMAW and model using commercial software AC3 [20]
use similar residual stress levels when welding to the H-plate to determine if welding process, parameters, or base material will have issues before testing [58, 65].

Figure 2.19: Definition of toeline extension and effective angle. Ade has shown increasing both dimensions reduces ballistic failure [65]

Figure 2.20 Different lengths of cracking experienced for GMAW transfer modes when H-plate tested [65]
Hydrogen induced cracking (HIC) is of concern in the weld and HAZ of armor materials due to the high susceptibility of the martensitic microstructure, higher levels of residual stress from thick armor plating, and the presence of hydrogen from either the welding process (i.e. electrodes, base metal) or manufacturing conditions present (i.e. ...
humidity, oils and greases) [55]. Magudeeswaran et al. did several experiments involving investigating the susceptibility of HIC of the welding process and filler metal. The use of the mercury method to measure diffusible hydrogen and the implant test (Figure 2.23) were used to investigate the role of low hydrogen ferritic, austenitic, and nickel filler metals for both SMAW and flux core arc welding (FCAW) processes. The result was of minor differences in the diffusible hydrogen found, but the lower critical stress (LCS) during implant testing was found to best in the high nickel filler metals due to the completely austenitic microstructure that have high solubility and lower diffusivity of hydrogen compared to the austenitic filler metal (contains some ferrite content) and the ferritic filler metal. The FCAW process was shown to have a higher LCS, mostly due to the heat input being lower. Conversely, for the tensile and toughness measurements, the low hydrogen ferritic filler metal was found to be best due to the acicular ferrite present in the microstructure that has been shown to have higher toughness and reduced softening of the microstructure. Preheat has also been examined as a method to maintain the HIC susceptibility with positive improvements in cracking in the y-groove test [58]. However, the softening is also increased as a result of preheat, leading to ballistic protection deterioration.
A potential improvement in ballistic properties could lie in the increase in hardness of the fusion zone by the addition of hardfacing filler metals. Reddy investigated the use of hardfacing material as the weld metal, but it resulted in brittle failure of the weld when tested with ballistics [67]. Investigations varying the amount of hardfacing and austenitic filler metal where investigated that showed improvement with the hardfacing metal between layers of austenitic filler or as the top layer (Figure 2.24). A potential issue found with this welding arrangement was cracking by creation of brittle precipitates in the dissimilar mixed zones of the base metal and hardfacing material [66]. The addition of a buttering layer of austenitic filler before the subsequent filling of the joint responded in no cracking and better ballistic properties.
Limited research is available, but the evaluation of laser welding could pose increased properties in armor applications due to the low heat input and ability to make full penetration welds autogenously. Bassett studied the comparison of hybrid GMA/laser, laser welding, and GMAW processes on ballistic protection and softening extent. The width of the HAZ decreased from 60mm in GMAW to 20mm in hybrid to 10mm in laser welding [59]. The resulting preliminary ballistics tests did not create a “detectable failure window” for either laser or hybrid welding, while failure did occur in the HAZ of the GMAW.

Figure 2.24: Hardfacing and austenitic weld metal combinations effect on ballistic worthiness [67]
2.3.3. Softening in Heat Affected Zone (HAZ)

Softening in the HAZ occurs when the base metal hardness is harder than that of the HAZ. This is prevalent in circumstances where the initial microstructure of the base metal has harder microstructures such as martensite and bainite. Depending on the chemical composition of the original material and thermal cycle imposed in the HAZ, softening can develop. As described above, armor steels readily see this softening issue due to most of these steels initial microstructure being mostly martensitic. When tempered, martensite has numerous locations where cementite can form, such as the packet boundary, prior austenite grain, and dislocations (Figure 2.25) [71].

Figure 2.25: Schematic of lath martensite and locations denoted by arrows for precipitation of cementite [71]
A primary driving force for softening is the prior austenite grain size and cooling rate for many steels, including low-carbon bainitic steel (Figure 2.26) [72]. In this case, assuming the steel is fully austenitized, the faster cooling rate will favor martensitic transformation formation and a harder microstructure. When cooling slower, enough time is allowed for the nucleation and growth of softer ferrite and bainite. The prior austenite grain size will grow with increasing cooling rate as the time in the austenite phase field and temperature for grain growth will be favorable.

![Figure 2.26: Experimental correlation of prior γ grain size and hardness as a function of cooling rate in a low carbon bainitic steel [73]](image)

Similar to armor steels, AHSS, such as dual phase steel, commonly used in the automotive industry also soften due to martensite content [15]. The dual phase (DP) microstructure of martensite and ferrite can be adjusted to reach various strength and ductility levels depending on the amount present for each constituent. Softening in DP
steel is a strong function of the carbon content in the martensite [13]. Carbide forming elements such as Cr and Mo actually help resist the softening as they tend to keep the carbon content lower in the martensite originally. This results in decreased softening when the martensite tempers due to the lack of carbon and slow kinetics. Similarly, when the amount of martensite is reduced, the softening decreases because there is less volume fraction of martensite being tempered [15]. The extent of softening is often reported going to the $A_{c1}$ temperature, so reducing the time at temperature by moving to low heat input processes like laser welding has alleviated this problem [74]. However, a recent study using nano-indentation shows that the extent of softening extends out farther than the $A_{c1}$ temperature most researchers report as the limit of softening [75]. The explanation is the resolution of microhardness capability to examine the phase boundaries of martensite tempering. Experiments with laser welding have shown that there will still be a soft zone present (Figure 2.27), but properties are increased with decreased heat input [76].
Pipeline steels, such as x100 and HSLA 80 and 100, have also shown softening [16, 19, 77]. For increased alloying content from HSLA80 to 100, the softening decreases because the majority of the microstructure will transform fully to martensite on-cooling [19, 78]. The development of allotriomorphic ferrite in the fine grain HAZ allows this steel to soften [19]. Moving to lower heat input processes, Miranda showed that there is still softening in laser welded X100 steel. However, compared to the conventional gas tungsten arc welding, the softening was reduced and toughness properties were shown to improve [77]. Arc welding techniques such as temper bead welding have shown some success, but still impart softening on the HAZ at some location that minimizes the benefit on toughness [79].

The development of ultra fine grain ferritic steels by thermo mechanically controlled processing (TMCP) also experience softening due to the chemical composition
and prior processing being affected by the HAZ thermal cycle [12, 80, 81]. The initial microstructure is refined by recrystallization in rolling passes in the austenite regime and the addition of grain refining elements such as vanadium and niobium [12]. When welding these steels, rapid softening occurs with increasing heat input up to an extent and levels off (Figure 2.28). While the hardness level may level off, the width of this region will keep increasing, making this area more susceptible to reductions in strength and toughness (Figure 2.29). These fine grains change the nucleation and transformation compared to work originally modeled by Ion et al. that is commonly used as a starting point for modeling these behaviors [12, 82]. Therefore, further evaluation and development of these models are needed. The nucleation of ferrite on the grain refining carbides also allows softening to occur and failure in the HAZ of the welded part (Figure 2.30). The use of laser welding has been used to limit the nucleation and growth of these fine grain steels [80, 81]. Wang investigated the effect on toughness of simulated laser HAZ and a HAZ with a post weld heat treatment. The result showed that the toughness was higher when laser welded than the base metal or the tempered region (Figure 2.31) [81]. TEM observation showed increase in packet size of the martensite after tempering, decreasing the toughness. The use of liquid nitrogen was also explored to provide faster cooling rates to prevent the softening [80]. The results surprisingly showed that the liquid nitrogen did not help with the strength or level of HAZ softening until the temperature was below -100°C when strength was increased and failure would occur in the weld metal as opposed to the softened HAZ.
Figure 2.28: Minimum Hardness as function of heat input for a TMCP steel [12]

Figure 2.29: Effect of Heat Input on width of softened zone in a TMCP steel [12]
Figure 2.30: Failure during tensile testing due to HAZ softening in a TMCP steel [12]

Figure 2.31: Effect of tempering on laser welding ultra-fine grain steel toughness [81]

2.3.4. Multiple Welding Pass Effects

There is a large amount of literature, especially for pipeline steels, that discuss the effects of multi-pass welding [16, 83-88]. As seen in Figure 2.32, the subsequent
reheating of a prior CGHAZ gives rise to the naming of the most recent pass and coarse grain heat-affected zone. For instance, in area C, the 2nd pass heats the first zone’s CGHAZ into the intercritical region, so it is aptly named the intercritically reheated coarse grain heat-affected zone (IC CGHAZ). This particular region also develops the lowest toughness of any region in a multi-pass structure (Figure 2.33). The reason for this drop in toughness is the creation of a martensite-austenite (M-A) constituent. This occurs most unfavorably in the intercritical region because only partial transformation of austenite occurs, allowing enrichment of the austenite to occur. It has been reported that a bulk carbon concentration of 0.06 weight percent can have M-A constituents with 1.32 weight percent in these regions [85]. The insufficient time for substitution element diffusion during welding is one of the reasons why these can occur. The transformation behavior of the martensite is has been identified as twinned consistent with the high carbon concentration [85]. The formation of the majority of M-A constituents is on the prior austenite grain boundaries [86], but sometimes the austenite nucleates on the bainite sheath and to form stringy M-A constituents seen as arrow A in Figure 2.34 [88]. The mechanistic failure occurs due to high stress at the M-A constituent/ferrite interface [87, 89, 90]. Besides slower cooling rates or changing composition, one solution is to perform a post-weld heat treatment (PWHT) with temperature above 350°C to decompose the M-A constituent (Figure 2.35) [88]. The reduction of stress at the M-A constituent ferrite interface is one of the reasons PWHT helps increase the toughness. Studies by Fairchild and Bonnevie have shown that Si and C content both delay the tempering behavior of M-A constituents, but they will decompose into carbides.
Figure 2.32: Schematic of weld locations during reheating of CGHAZ in a multi-pass weld [85]

Figure 2.33: Reduction in Charpy toughness due to M-A constituent [85]
Figure 2.34: M-A constituents seen in a micrograph of simulated IC CG HAZ (A) string M-A constituent (B) blocky M-A constituent [88]

Figure 2.35: Effect of peak temperature temper on M-A constituent density in a Si-added low carbon steel [84]
The prediction of these constituents, and in general all other microstructures in welding, help set process windows and understanding of what can occur when given an initial composition and microstructure. However, there are inherent issues within the initial material that make these predictions difficult, especially when using bulk compositions. This next section will discuss the role of heterogeneity in steel making and its effect on the initial material and their weldability.

2.4. Heterogeneity in Steels

2.4.1. Macrosegregation

Macrosegregation occurs during the initial solidification of an ingot or continuously cast part [91-93]. In the ingot (Figure 2.36a) the first dendritic solidification from the bottom creates negative segregation from rejection of the solute and equiaxed pure solid. V-segregation in the centerline results from enriched solute formed at the end of solidification with large equiaxed grains. The A-segregation occurs from the columnar grains reaching the equiaxed region of the ingot where solute from these two regions meet. The final stage of solidification and highest solute content is in the hot top that forms due to buoyancy driven convection and shrinkage fluid flow.

During continuous casting (Figure 2.36b), the mushy zone should contain the liquid, but macrosegregation can occur when insufficient roller constraint allows the liquid to become enriched and concentrated at the centerline. The thermal contraction during cooling also aids in the development of the macrosegregation [91]
2.4.2. Microsegregation

Microsegregation in the literature is often times referred to as banding. This occurs due to micro-scale segregation from the initial casting. Banding would not be expected if steel was just the Fe-C system because of the high diffusion rate of carbon. The problem arises when another element that affects the activity of C (i.e. Mn) causes both to segregate and on-cooling creates banding. There are two types of microsegregation as defined by Kirkaldy, pre- and trans-segregation [94-96]. Pre-segregation is defined as the segregation created during solidification and rejection of solute from solid to liquid (Figure 2.37a). Trans-segregation refers to the solid-solid segregation such as that in ferrite pearlite steels (Figure 2.37b). One big determinate of banding in steel is the prior austenite grain size. As shown in Figure 2.38, if prior grain austenite grain size is less than the segregation (Case I) the banding will be more dominate because ferrite will nucleate and grow first in Mn poor regions, rejecting carbon.
and Mn solute into the higher Mn content regions. This will continue throughout these Mn poor regions only allowing pearlite to form in the higher C and Mn regions. However, if the prior austenite grain size is larger, ferrite will have a more difficult time nucleating and growing only in the Mn deprived regions, but banding could still occur. Similar circumstances can occur when MnS inclusions are in the steel, as the region around them will be diluted of Mn, allowing ferrite bands to form around the MnS (Figure 2.39).

Figure 2.37: (a) Pre-segregation from ingot cast of steel  (b) resultant trans-segregation of ferrite and pearlite after rolling [97]
Mechanical properties can also be affected by banding, especially ductility and toughness while tensile properties are usually the same. A major aspect banding effects is applications where thermal operations (e.g., carburizing gears) depend on a nominal composition for quench and temper process to create the right hardness and strength. D’Errico saw that banding in a gear lead to quench failure because of banded microstructure that was assumed to be homogenous [98]. Krauss and Majka evaluated
the band spacing by producing engineered alternating layers of Mn rich and poor 5140 steel and performing tensile tests at various cooling rates from austenitized state [94, 99]. When cooled quickly, the steels had a wide scatter in property because there was still insufficient time for diffusion of the banded microstructure. However, as the cooling rate was slowed to 1°C/min, the properties leveled off and the banding was no longer present due to time at temperature allowed for similar decomposition into ferrite and pearlite in both the high Mn and low Mn regions. Grange studied the effect of clean and “dirty” (sulfide inclusion containing) steel with roughly the same composition [97]. He found that roughly the same properties tensile and yield strength occur between banded, non-banded, and inclusion containing steel. There was anisotropy of the inclusions and banded steel that also affected the toughness deleteriously.

Krebs studied the effect of banding as an effect on dual phase steel. The initial F/P microstructure lead to a similar ferrite/martensite banding due to the intercritical processing of DP steel to create austenite in the pearlitic regions and when quenched, form martensite there. The results found micro-chemical segregation was different compared to the morphological nature of the DP ferrite and martensite (Figure 2.40) [100].
2.4.3. Heterogeneity effect on Welding

Similar to the effect D’Errico experienced, welding will be affected by the levels of banding within the material. Most welding procedures and PWHTs are designed for a nominal composition and microstructure. The main detraction concerns the prediction and modeling of these microstructures and properties during welding. The variation in composition throughout the initial microstructure due to banding nullifies the efforts to predict the microstructure in the HAZ regions. Aidun and Savage have shown that this is indeed the case, where a common low carbon steel that had banding resulted in an unexpected hardness of over 60HRc in the HAZ due to strong F/P banding. The only solution that could be made for repairing this structure was to completely homogenize the steel before welding [101]. Another study using the temper bead technique [102] showed...
that toughness was decreased greatly due to banding in the microstructure. The result of the temper bead had limited success, but subsequent PWHT did increase properties of the steel. Whether this was due to banding, is not apparent. What is apparent, however, is the difference in fracture surface comparison of a PWHT sample and as-welded condition as seen in Figure 2.41 where ridges from the banding are seen in the aw-welded sample (b).

![Figure 2.41: Charpy Toughness result of banded4140 steel (a) with PWHT (b) as-welded][102]

Little information is published on the effect of segregation banding on armor steels, but a report released by the Australian Department of Defense also reports the reduction of fracture toughness, fatigue, and stress corrosion cracking in armor steel BISALLOY 500 [103]. The banding lead to cracking or secondary cracking from the initial propagation path within the bands during Charpy impact testing. While all cracks
were classified as detrimental, there is potential benefit from secondary cracking to fatigue crack propagation, but is hypothesized as being harmful to ballistic properties. Inclusions have been shown to have deleterious effect on adiabatic shear bands (ASB) in armor materials [53]. ASBs are formed under very high strain rates and characteristic in explosions and ballistic impact. The formation begins when a rapid increase in temperature below $A_{c1}$ from adiabatic heating occurs because of the inability to conduct heat away from the high strain creating many dislocations, very fine carbides and martensitic structures [104]. The bands etch white and have similar hardness to martensite. The load carrying capacity of this structure is low and can lead to failure as seen in Figure 2.42 [50]. If the inclusions are near the area prone to ASB formation, cracking can initiate even in ductile materials [105]. ASBs can be heat treated to eliminate the cracking potential, but it is often difficult to do in the field.

![Figure 2.42: Formation of adiabatic shear band (ASB) from impact loading on both sides of an armor steel (left) Cracking initiating within an ASB (right) [50]](image-url)
Chapter 3: Objectives

The main objective of this research was to determine the effects of welding on flash processed steel microstructure and mechanical properties. To accomplish this goal, flash processed 4130 and comparative, currently deployed, high hard armor steel were analyzed to see the effects of welding. Preliminary work and literature showed potential areas to examine specifically. In particular, the specific objectives of this work include:

1. Show why and how flash process steel softens in the HAZ and compare these results to high hard material.
2. Evaluate currently used welding processes and mitigation techniques for HAZ softening and evaluate the effectiveness for flash processed steel.
3. Evaluate different welding processes form normally used that could improve properties affected by HAZ softening.
Chapter 4: Materials and Procedure

4.1. Materials

The two materials investigated in this work were flash processed 4130 and high hard armor supplied from SFP Works LLC. The high hard was from another unknown producer, but met material specifications for HHA. The composition for each material is seen in Table 4.1. The table also summarizes the AWS D1.1 (IIW) carbon equivalent [106] values as denoted in Equation 4.1 along with calculated A₁ and A₃ temperatures using ThermoCalc TCFE5 database [107] and calculated Mₛ and Bₛ temperatures [108, 109].

Table 4.1: Material Composition and Properties

<table>
<thead>
<tr>
<th>Material</th>
<th>Fe</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
<th>Cu</th>
<th>W</th>
<th>Ti</th>
<th>Co</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>High Hard (approx wt%)</td>
<td>Bal.</td>
<td>.3</td>
<td>.43</td>
<td>.9</td>
<td>.9</td>
<td>.55</td>
<td>.55</td>
<td>.005</td>
<td>.1</td>
<td>.01</td>
<td>.03</td>
<td>.005</td>
<td>.05</td>
</tr>
<tr>
<td>FP 4130 (wt%)</td>
<td>Bal.</td>
<td>.28</td>
<td>.2</td>
<td>.5</td>
<td>.015</td>
<td>.88</td>
<td>.17</td>
<td>.005</td>
<td>.03</td>
<td>.003</td>
<td>.008</td>
<td>.002</td>
<td>.001</td>
</tr>
</tbody>
</table>

\[
CE = C + \frac{(Mn + Si)}{6} + \frac{(Cr + Mo + V)}{5} + \frac{(Ni + Cu)}{15}
\]

Equation 4.1: Carbon Equivalent as defined by AWS D1.1[106]
The representative microstructure of each material can be seen in Figure 4.1. Both FP and HH microstructures have martensitic lath structure as evidenced more clearly in the SEM backscattered electron (BSE) imaging. The presence of carbides, as discussed in the background, appear in the FP sample and image brightly in the BSE image below. It is uncertain from the images if bainite is present in the FP sample, but high hard does appear to be fully martensitic.

Figure 4.1: Base Metal Microstructure of FP 4130 (a) optical and (c) back scattered electron (BSE) imaging and HH (b) optical (d) BSE imaging
4.2. **Welding Processes**

4.3. **Gas Metal Arc Welding**

From the review of literature on welding of steels and the effect of HAZ softening, low-heat input direct current, electrode positive (DCEP) GMAW was performed on both ¼” thick FP and HH steel with identical parameters with 0.045” E110C-K4 filler wire (composition in Table 4.2). Shielding gas mixture of 75%Ar, 25%CO\textsubscript{2} was used in the 6-pass 22.5° v-groove joint. As usually done with fusion welding processes on armor materials [110], a matching material backing bar of equal thickness was placed behind the v-groove. A schematic using EWI’s WeldPredictor\textsuperscript{TM} [111] is shown in Figure 4.2. The average heat input of 19.4kJ/in is based off the average current of 250 amps, voltage of 31 volts, and 24 inches per minute (ipm) travel speed. Arc efficiency is ignored in this heat input in order to compare nominally with laser welding that does not have a well defined process efficiency, especially between various types of laser equipment [112-114].

\textit{Table 4.2: E110C-K4 filler wire compositor}

<table>
<thead>
<tr>
<th>Material</th>
<th>Fe</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
</tr>
</thead>
<tbody>
<tr>
<td>E110C-K4</td>
<td>Bal.</td>
<td>.003-.006</td>
<td>.66</td>
<td>1.5-1.8</td>
<td>.6</td>
<td>2.5</td>
<td>.6</td>
</tr>
</tbody>
</table>
4.4. Autogenous Laser Welding

Laser welding was accomplished with assistance from EWI. A 15kW IPG Ytterbium 200μm Fiber Laser operating at 6kW with a 333μm spot size focused at the top of the weld surface was used for all laser welding. The work was split in two phases consisting first of bead-on-plate (BOP) welds to find correct parameters, and the second phase used the best parameters to do zero-gap butt welds. 6kW was chosen as a ‘rule of thumb’ from experience at EWI to use 1kW of power for each millimeter of thickness. So, the 6.35mm (1/4”) of thickness for both materials equates to using the aforementioned 6kW of power. Both sides of the weld were shielded with pure Argon gas in addition to a trail shield to prevent oxidation. Two air knifes were also employed to prevent contamination of the optics from plasma and fine particulates created during the welding process.

BOP welds were made on both materials to find the proper travel speed from 60-120ipm for the given laser power of 6kW. This was done to limit HAZ softening with a high travel rate, but slow enough to avoid root humping and spiking that is common in laser welding due to capillary instability and fluid flow of the weld pool at too high of travel speeds [115, 116].
As will be discussed in the next chapter, a travel speed of 90ipm was used for the zero-gap butt welds giving a heat input of 4kJ/in ignoring efficiency factors. The edges of the plate in contact for the butt weld were initially wire-EDM cut for the best edge retention. However, since the process was done in water, minor corrosion occurred that was deemed unsuitable for laser welding. Therefore, a milled edge was used for the butt welds. Radiography was used in order to make sure full penetration and no weld defects occurred during the laser welding. The parameters for radiography were 320kV, 5 milliamps and exposure for twelve seconds at a distance of 36” from source to film. A standard ASTM wire set A image quality indicator (also known as penetrameter) with 0.008” wire to ensure the correct sensitivity was used with these parameters.

4.5. Phase Transformation Analysis

In order to evaluate the HAZ properties more precisely, a Gleeble 3800 was used to control the peak temperature, heating and cooling rates for specific HAZ regions. The system utilized low force jaws to allow proper dilation of the samples when heating and cooling. The thermal cycles were measured via a percussion welded Type K thermocouple to the middle of the sample. A quartz dilatometer is also used to track phase transformations via dilation on heating and cooling. This dilatometer uses a quartz push rod with LVDT and built-in air-cooling for minimal distortion of quartz rods. The dilatometer comes into contact at two points along the sample’s diameter. The dilatometer, similar to the thermocouple, is placed on the sample in the middle of the 1” free span between the copper jaws. The heating of the sample is achieved by passing high current through the sample and associated resistance heating. A sealed argon
environment is used to prevent oxidation and thermocouple detachment during the thermal simulation. The material was prepared from ¼” thick plate by wire-EDM. The sample dimensions were cylindrical rods 4” long and 5mm in diameter. A similar setup used by Kullman was used in the Gleeble 3800 as seen in Figure 4.3 for the experimentation [117].

Using Rosenthal thin plate solution considerations and heat input conditions from the GMAW of approximately 19 kJ/in (7.5 kJ/cm), the physical simulation parameters were selected (i.e., heating rate of 100°C/s and a cooling rate of 10°C/s) [118].

Table 4.3 shows the various peak temperatures used in this study and their choice will be described further in later chapters. Figure 4.4 shows the resultant graph of temperature versus time illustrating select peak temperatures and the constant heating and
cooling rate in FP samples. Departures from linear cooling rate were observed due to the release of latent heat during decomposition of austenite. At temperatures less than 200°C, the inability of Gleeble to maintain the 10°C/s cooling rate is due to lack of forced cooling in our experiments.

Table 4.3: Peak Temperatures used in the Phase Transformation Analysis

<table>
<thead>
<tr>
<th>HAZ Type</th>
<th>Coarse Grain HAZ</th>
<th>Fine Grain HAZ</th>
<th>Intercritical HAZ</th>
<th>Subcritical HAZ</th>
</tr>
</thead>
<tbody>
<tr>
<td>Peak Temp [°C]</td>
<td>1300</td>
<td>1100</td>
<td>1000</td>
<td>900</td>
</tr>
<tr>
<td></td>
<td>800</td>
<td>785</td>
<td>770</td>
<td>755</td>
</tr>
<tr>
<td></td>
<td>740</td>
<td>720</td>
<td>700</td>
<td></td>
</tr>
</tbody>
</table>

Figure 4.4: Measured Thermal Cycle profiles of phase transformation analysis for selected peak temperatures

4.6. Mechanical Testing

Tensile testing was performed on both FP and HH steels in accordance with ASTM E8 subsize. The base metal, GMAW, and laser welds were all tested with the same 100kN hydraulic load frame with hydraulic wedge grips. The GMAW welds were
ground flat to a uniform thickness before final machining. All samples (laser, GMAW, and base material) were machined to ASTM sub-size samples via wire-EDM. Testing was done in triplicate for improved accuracy. An ASTM E83 Class A 1” extensometer was used to measure elongation data for all testing.

A laser extensometer was also used to verify the elongation data for the FP GMAW samples. The laser employed a moving window average of 512 scans with a 100Hz refresh rate. The target distance was 305mm and the device has a resolution of 0.001”. It is important to note that the laser extensometer is not actually calibrated to an ASTM standard, and is done on an as-is basis. The primary use of this device was to make sure the mechanical extensometer was fully reading the elongation in the FP GMAW samples. A picture of the testing setup can be seen in Figure 4.5.

Figure 4.5: Tensile testing setup with both mechanical and laser extensometers
A representative fracture surface was analyzed in a Philips ESEM FEG-30 scanning electron microscope at 15kV for each condition tested. The samples were cut, cleaned with acetone, and mounted onto stubs with conductive carbon paint before analysis in the SEM.

4.7. Metallographic Preparation

Samples for metallographic and hardness mapping were prepared by mounting in an automated mounting press using conductive mounting powder. Grinding was performed sequentially with 320, 400, 600, and 800grit SiC paper. In between steps, the samples were rinsed and dried with cool air. Subsequently, polishing was performed sequentially with 6μm and 3μm diamond paste with diamond extending oil as lubricant washing with soap and water and drying in cool air between steps. Final polishing was achieved with 0.05μm colloidal silica with rinse in hot water to ensure colloidal silica did not solidify on the sample.

For EDS analysis, the samples were cleaned in an ultrasonic bath of ethyl alcohol before further polishing on a vibration polisher with non-crystallizing 0.05μm colloidal silica. This was done to ensure the sample was as flat and polish-defect free as possible.

For imaging purposes, the sample must be etched to differentiate the different phases and microstructures of the materials. Several etchants are available for use on steels [119], but 2% and 5% Nital were used for the optimal image quality. An important note is that etching was usually done after hardness mapping to ensure quality indents were made on the flat specimen surface.
4.8. **Hardness Mapping**

The use of an automatic microhardness indenter became essential in this work for analysis of phase transformation analysis and both welding processes. The mapping allowed to see the variation in hardness spatially in 2-D as opposed to 1-D line traverse. As will be seen in subsequent chapters, this allows for better analysis of heterogeneity within the initial microstructure that affects the results for all of the experiments. A Leco AMH43 testing system is used to make and measure the indents automatically after setting up the area and load to be used. For the majority of all experiments, a 300g load was used.

The use of an IGOR Pro macro designed by the author was used to map the hardness distribution with x and y coordinates and regular 16 bin rainbow Vickers hardness scale. Hardness Histograms were also made with this program with frequency distribution curves and binning methods proposed by Scott [120]. The tools provided quantitative measure of the reactions in the weld and HAZ areas in addition to determining the peak temperatures most susceptible to HAZ softening.

4.9. **Optical Microscopy**

An Olympus GX-51 inverted microscope with an Olympus DP71 digital camera were used for optical pictures that included software for white and color balancing for optimal picture quality.

4.10. **Scanning Electron Microscopy**

The use of a Philips ESEM FEG-30 was used for the majority of SEM work that included imaging and EDS analysis. The microscope uses a field-emission gun source
that gives higher resolution compared to tungsten source SEMs. It was essential that the samples were cleaned of contamination by ultrasonic cleaning and a 100°C oven bakeout of one hour before analysis. The main use of the SEM was the greatly increased maximum magnification compared to optical microscopes available. The evaluation of pearlite, M-A constituents, laths in martensite, and carbides were much easier with this magnification level especially with the fine grain size and carbides present using the secondary electron (SE) detector.

The use of backscattered electron (BSE) imaging in the a+b configuration allowed the ability to see z-contrast of heavier element. This aided in analysis of carbides and other constituents within the material. For fractography, the a-b configuration of BSE imaging allowed topographical evaluation in some cases better than that of secondary electron imaging that often has shadowing effects.
Chapter 5: Experimental Results and Discussion

This chapter is broken down into four main sections: base metal analysis, GMAW results and discussion, single-pass HAZ physical simulations results and discussion, and laser welding results and discussion. The base metal properties for the two steels are discussed in terms of hardness, microstructure characterization, and mechanical properties measured by tensile testing. The baseline gas metal arc welding conditions are discussed in similar terms as the base material properties along with fractography of the tensile tested samples. Phase transformation analysis of single-pass heat-affected zone physical simulations are examined with dilatometry, hardness, and characterization efforts are the primary focus of this section to describe the effect of different HAZ thermal cycles on both materials examined. The final section analyzes the laser welding and the subsequent effect on the hardness and microstructure of the base materials. The same analysis techniques used in the GMAW are used with these testing conditions.

5.1. Base Metal Characterization

5.1.1. Flash Process 4130

In order to understand the underlying microstructure for the flash processed 4130, as-received steel before flash processing was looked at in addition to the post-flash processed material.
5.1.1.1. Prior to Flash Process

Initially, the exact processing conditions are not exactly known for the AISI 4130 material, except that it has been spheroidized. Unpublished work by Cola has shown that spheroidized Cr-enriched carbides give the best properties in terms of ballistic resistance. The hardness map (Figure 5.1) and histogram (Figure 5.2) show a spread in hardness from 147HVN to 182HVN with the mean being 162HVN. Figure 5.3 shows the microstructure of the prior to flash processing 4130. Figure 5.3a shows a hard region from the corresponding hardness map in Figure 5.1 that contains ferrite and spheroidized cementite from the potential original pearlite. Liu investigated a similar occurrence where he found local precipitation of $M_7C_3$ carbides in addition to widmanstatten ferrite from the former cementite particles and martensite [121]. There could be similar conditions occurring here, giving rise to the variation in hardness, even though the microhardness indents will average out the hardness due to their size being on the order of 40μm while these microstructures are on the order of sub-microns. Lower hardness regions (Figure 5.3b) contain more polygonal ferrite and less of the carbide containing regions. Figure 5.3c shows another area of interest in the original sample, where an orange constituent is seen in the carbide region. When analyzed with BSE imaging, the constituent appears very dark, signifying light elements due to $z$-contrast. In addition, to EDS spot scan analysis (Figure 5.4) showing increased titanium, carbon, and nitrogen leads to the conclusion this is a titanium carbonitride. Similar analysis of an armor steel investigated by Unfried showed similar TiCN within armor steels that function primarily as prior austenite grain size refiners, by impinging austenite grain growth as the form
along grain boundaries initially when cast and have stability well into the austenite regime [20].

![AISI 4130 hardness map before flash processing](image1)

**Figure 5.1** AISI 4130 hardness map before flash processing

![Histogram of AISI 4130 before flash processing](image2)

**Figure 5.2** Histogram of AISI 4130 before flash processing
Figure 5.3: Microstructures seen in AISI 4130 before flash Processing. (a-b) optical on left, SEM SE image on right (c) optical on left, SEM BSE image on right
(a) Hard region from hardness map in Figure 5.1.
(b) Softer region from hardness map in Figure 5.1
(c) Evidence of TiCN from BSE imaging and EDS data from Figure 5.4.
Figure 5.4: EDS spot scan analysis of dark imaging square constituent seen in Figure 5.3c. The high levels of Ti, C, and N lead to the postulation this is a titanium-carbonitride

5.1.1.2. After Flash Process

The same heat of material analyzed in the previous section was then flash processed using proprietary thermal cycling similar to that discussed in the background of flash processed steel. The hardness map (Figure 5.5) and histogram (Figure 5.6) ranged from 501HVN to 607HVN with average of 536HVN. It is noticeable in the microstructure pictures in Figure 5.7, that the carbides are still present after flash processing. The difference between the hardness levels in Figure 5.5 appears to coincide with the banding phenomenon seen in Figure 5.7b. The hardness increases by over
100HV. The main reason for this hardness could be attributed to the higher precipitate content initially within this area. It was noticed that the hardness of the non-flash processed region within these bands had increased hardness as well. Even with scanning electron microscopy, it is difficult to evaluate the microstructure exactly in these banded regions. It is noticeable that there are varying degrees of prior austenite grain size within these areas as seen in Figure 5.7b.

![Flash processed 4130 hardness map](image)

*Figure 5.5: Flash processed 4130 hardness map*
Figure 5.6: Hardness histogram of flash processed 4130 material
In the hardness map, there is a slight trend of harder microstructure on the upper right and lower left and softer microstructure in the middle. This is more evident in the high hard material that will be discussed shortly. This change in structure may come from the initial quenching process giving rise to harder microstructure and smaller grain size being closer to the surface of the plate. As a result, a hardness map was made on the initial base plate before the EDM cut was made (Figure 5.8). The variation of hardness ranged from 476HVN to 631HVN. The optical macrograph (Figure 5.9) shows the
correlation between etching characteristics and the through thickness hardness seen in the hardness map of Figure 5.8. The white etching band in the middle of the plate also correlates with the maximum level of hardness seen in the middle of the hardness map. As mentioned in the background, this white banding through the thickness at the centerline is a poster child for improper continuous casting by insufficient pressure by the rollers leading to macrosegregation [91, 92]. The hardness of these bands suggests there is a fine, high carbon martensite in these regions with segregated alloying editions of Cr and Mn. EDS analysis of similar banding will be presented later in this chapter.

Figure 5.8: Flash process as-received plate before EDM cut. The y-axis is through thickness of the plate.

Figure 5.9: Optical Macrograph of hardness map in Figure 5.8
5.1.2. High Hard Armor

The high hard armor is as-received and the prior processing steps are unknown. However, analysis of the hardness and microstructure leads to the conclusions that this steel is quenched and tempered similar to that reported in the literature [5, 37, 38]. The hardness map (Figure 5.10) and histogram (Figure 5.11) have a range of hardness from 497HVN-573HVN with an average of 531HVN. However, the microstructures (Figure 5.12) between these different hardness areas appear to be similar when analyzed. As discussed in the background literature, this may be due to the macrosegregation and subsequent rolling operations of the steel. However, it is not readily apparent why. The macrosegregation could have changed the level of carbon and alloying content in these regions, allowing the lath martensite to be strengthened in these regions, but more work is needed to evaluate this phenomenon.
Figure 5.10: High Hard Armor base metal hardness map

Figure 5.11: High hard armor histogram
5.1.3. Mechanical Properties

The stress-strain plot of the base material is seen in Figure 5.13. The UTS of FP is 260ksi and 245ksi for high hard. The calculated 0.02% offset yield strength [122] is 210ksi and 220ksi for high hard and flash process, respectively. The elongation at fracture for both materials is 16.0% for high hard and 14.8% for FP. The uniform elongation (elongation at UTS) is 7.5% and 9.0% for HH and FP, respectively. The failure of both base metals occurred by a mixed ductile/brittle failure as seen in Figure 5.14. The greater reduction in area of the HH sample seen in Figure 5.14d compared to
the FP sample in Figure 5.14a corresponds well with the extensometer data that the HH did have better ductility. The fracture modes present in FP include ductile and brittle behavior. As seen in Figure 5.14c, there are small ductile dimples on the order of 2-5 microns and cleavage shear. The interior of the sample showed quasi-cleavage fracture with equiaxed dimples concurrent with failure due to overload. The HH sample showed similar behavior, but as noticed in both Figure 5.14e and f the dimples are deeper and have larger size. This fracture behavior correlates well with the increased ductility of the HH compared to FP.

Figure 5.13: Stress-Strain curve for subsize base metal strength for flash process and high hard steel
Figure 5.14: Fractography of base metal (a-c) FP (d-f) HH. Both (b) and (e) are from the brittle area near the center of the cross sections while (c) and (f) are from the edge areas that showed more ductile dimples and shear surfaces.
5.2. **Baseline GMAW Evaluation**

As discussed in the experimental approach, both steels were welded with low heat input conditions with six passes in a v-groove joint on a backing bar. The hardness, microstructure and mechanical properties are discussed below.

5.2.1. **Hardness**

The resultant Vickers hardness map and macrograph of the GMAW welding for both the FP and HH steels are shown in Figure 5.15. Both share the same hardness scale of 200-550HVN. Recall that the base metal hardness for each FP and HH is 540 HVN and 530 HVN, respectively. It is apparent that significant softening is present in both the HAZs and the reheated weld metal. The softening in FP appears to be present throughout most of the HAZ, while some regions of HH are actually hardened above the base 530 HVN to around 550 HVN. The hardness for the as-deposited weld metal is reasonable for the composition (Table 4.2) and to be expected to avoid martensite that is susceptible to hydrogen cracking in the weld metal [106]. However, the weld metal is not of primary concern in this work due to the fact there is limited option in this regard to match the properties with a filler metal [123, 124].
5.2.2. Microstructure Characterization

Analysis of the two harder regions of FP and HH steel shows large prior austenite grain size with martensitic structure as seen in Figure 5.16a-b, respectively. The maximum extent of softening seen in these steels is 180 HVN for FP steel in the intercritical HAZ (Figure 5.16c) and 296 HVN for HH in the intercritical HAZ (Figure 5.16d). The microstructure present in the intercritically heated region of FP contains ferrite and tempered martensite with carbides present from the initial material. There are M-A constituents that have formed along the prior austenite grain boundaries in the FP sample that appear similar to those presented in Figure 2.34. The intercritically heated region of HH has a mixed microstructure containing tempered martensite and fresh martensite as seen in Figure 5.16d. The tempered martensite is the light etched region that does not form austenite on heating. The fresh martensite is the darker etching regions that transformed to austenite on heating and transform to lath martensite on cooling.
It is hypothesized that the chemical composition difference between the two materials (Table 4.1) explains the difference in microstructure evolution between the two. HH has more austenite stabilizing elements such as nickel that prevent the formation of ferrite for this low-heat input condition. In contrast, when austenite is formed on heating in the fine grain and intercritical HAZ of the FP steel, the grain size and composition favors the transformation of austenite to ferrite. Further analysis using physical simulations of these regions will be presented later to evaluate this hypothesis.
The impact of this softening in the HAZ impacts the ballistic worthiness. Mohandas and Reddy showed that the softened region was most susceptible to ballistic penetration [17, 18, 60]. However, the threshold for FP steel is not known, and further analysis is needed. However, the combinations of the softened HAZ and weld metal will likely result in inferior performance based on the literature.

5.2.3. Mechanical Properties

5.2.3.1. Tensile Testing

Using the ASTM E8 subsize standard, tensile testing was performed on both steels. Tests were done in triplicate to see repeatability of the data. The overall result for the tensile tests can be seen in Figure 5.17. The UTS and yield strength for high hard was 20ksi harder for FP for both measurements at 140ksi and 115ksi, respectively. Total elongation was consistently higher for HH steel (12% compared to 11%), but the uniform elongation was within the error of each other, with the average of FP having slightly better (4.3% compared to 3.6%). During testing, it was noticed that the failure occurred in different locations. The FP failed in the HAZ, while the HH failed in the weld metal. This is allows one to infer that the increased softening in the FP HAZ lead to the failure, while HH HAZ was hard and strong enough to prevent failure in this location. The content of potential M-A constituents in addition to the softened region may have also lead to the failure in this location.
The 115ksi yield strength of high hard is expected as the filler metal chosen was designed to have this yield strength. Besides using an austenitic filler material [123], or an autogenous welding process, this is the best strength and ductility to expect from the GMAW process.

5.2.3.2. Fractography

The failure, as seen in Figure 5.18a-c, has a unique appearance as it has a v-shaped failure ridge in the middle of the fracture surface that seems to be where the stress had concentrated. Simulations by Panda for laser welds show experimentally and with numerical finite element simulations similar failure modes (Figure 5.19) [125]. Panda states that the stress concentrates in an x-type pattern in the softened region, initiates and fails in the middle of the material. This is in agreement with the results presented here. The fracture surface shows ductile dimples and cleavage fracture mixed in both near the
plastically deformed region (Figure 5.18c) and in the v- region of high stress (Figure 5.18b). Similarly for high hard, the fracture surface (Figure 5.18d-f), shows a mixture of fine, equiaxed ductile dimples in the middle of the fracture to cleavage and shear of grains seen on the edges near the surface where most of the plastic flow has occurred.

The most important consideration from this tensile testing is the failure occurring in the HAZ for FP, while the weld metal fails for HH. This is resultant from the increased extent of softening in the HAZ that results in the weaker ferrite microstructure compared to the martensitic structure in the HAZ of high hard material.
Figure 5.18: GMAW Fracture Surfaces: (a-c) FP (d-f) HH Both (b) and (e) are from the center of the cross sections while (c) and (f) are from the edge areas
Figure 5.19: FEA modeling by Panda [125] showing concentration of stress in softened HAZ consistent with fracture seen in Figure 5.18

5.3. Single-Pass Heat-Affected Zone Simulations

After GMAW, the main differences in HAZ behavior between FP and HH were still unclear. The exact location, peak temperature, and microstructure evolution were unknown. It was seen that somewhere below the $A_{c3}$ temperature both steels had differing characteristics. In order to evaluate these conditions, thermal simulations were conducted with the Gleeble 3800 to control the specific HAZ thermal cycles via peak temperature and heating and cooling rates.

5.3.1. Thermal Cycles

The thermal cycles described in Table 4.3, where chosen to analyze each of the main HAZ regions within the two materials. For FP, there was softening seen throughout the HAZ, while the HH steel showed hardening and softening compared to the base metal. For this reason multiple temperatures were chosen from the fine grain and intercritical
HAZ to find the maximum extent of softening and microstructure present for each of these materials. While it is documented that multi-pass welds are essential for welding of these thickness materials, the single pass allows one to examine the effect of one particular thermal cycle on the microstructure and properties of the steel.

5.3.2. Phase Transformation Analysis

The dilatometry data was used to determine on-heating and cooling phase transformation temperatures. For on-heating conditions, the use of linear fit analysis was used to find the critical temperatures $A_{c1}$ and $A_{c3}$. Data from the CGHAZ and FGHAZ showed $A_{c1}$ and $A_{c3}$ temperatures of 747°C ± 8°C and 835°C ± 25°C for flash process steel. Similar data for high hard showed $A_{c1}$ and $A_{c3}$ temperatures of 733°C ± 8°C and 818°C ± 8°C. The variation in this data is higher than that of other similar experimentations on other steels [57, 126, 127] and is accounted for by the variation of through thickness microstructure and banding in the microstructure that will be discussed later in this chapter.

5.3.2.1. Flash Process

All of the dilatometry curves with the observed temperature and time are presented in Figure 5.20. Using known calculations from CCT diagrams, the Ms and Bs start temperatures [108, 109], and having data showing dilatational changes in slope via dilatometry, it can be determined which phase should be present and confirmed optically afterward. For instance, it can be seen that the CGHAZ fully transforms to austenite and continues heating until 1300°C. On cooling, there is no change until below the Bs temperature of 536°C and another change in dilation near the Ms temperature of 396°C.
allowing one to conclude the structure is a mixture of bainite and martensite. Further analysis developed by Eldis [128] allows for the calculation of how much of the transformation has completed and estimation on the volume fraction of each transformation product. In Figure 5.21a, linear curve fitting for the coefficient of thermal expansion for both austenite and ferrite are shown with purple and green lines, respectively. A tie line (line AC) is then constructed during the transformation between these extrapolated linear fits. The lever law (BC/AC) is then applied to determine the fraction transformed. This can be scripted and done with a computer to plot the complete transformation behavior as seen in Figure 5.21b. Knowing the transformation temperatures (e.g. B_s, M_s, M_f) for the material allows for complete knowledge of the microstructure. For the CGHAZ sample shown in Figure 5.21, it can be concluded that about 50% bainite and 50% martensite should have formed understanding the B_s temperature is 536°C and M_s temperature is 396°C. For the FGHAZ sample with peak temperature of 900°C, it can be seen that the sample is fully austenitized by 810°C and heats to 900°C before cooling. On cooling the dilation is observed before the A_c3 temperature, so a fully ferritic microstructure should be expected. The dilatometry curves starting at 800°C and decreasing to simulate the ICHAZ do not show full austenization as there is no dilatational change for the A_c3 temperature to show fully austenitic microstructure. In addition, there is only ferrite transformation seen on-cooling for these peak temperatures. The 740°C temperature shows no sign of transformation to austenite leading to only tempering of the initial microstructure occurring.
Figure 5.20: Flash Process simulated single pass HAZ thermal cycle dilatometry curves
Figure 5.21: Phase Transformation Analysis of FP CGHAZ 1300°C (a) Use of the technique developed by Eldis to find the percent transformed at a specific temperature (b) Plot of all tie line analysis within the temperature range
5.3.2.2. High Hard

Similar principles to those mentioned above are used to describe the phase transformations in HH (Figure 5.22). It is noticeable immediately, that there is no other transformation on-cooling for HH other than martensitic transformation below the $M_s$ temperature of 396°C. The lone exceptions are the subcritically heated areas where no phase transformations on heating or cooling occur. This is indicative of a fully martensitic microstructure throughout the HH steel. The data allows one to conclude the softening for high hard steel is predominantly due to the tempering of martensite on heating to a temperature below Ac1 (subcritically heated). It is also important to note that while the 720°C peak temperature is designated as subcritical, the dilatometry data suggests there is possible austenite formation on heating leading to fresh martensite formation on cooling. Hardness and microstructure observation must be done in order to evaluate this as will be seen in the sections below.

Another noticeable difference in dilation data for HH is the extent of dilation seen during the martensitic transformation. The $\Delta R/R$ range of $2.5 \times 10^{-3}$ for peak temperature of 770°C compared to the peak temperature of 900°C $\Delta R/R$ range of $5.5 \times 10^{-3}$ can be explained by the experimental setup with respect to some microstructural banding observed in the HH steels. The dilatometer is usually set on two opposite sides of the round sample. Since the samples are round, it is almost impossible to place the dilatometer in the same orientation in respect to the orientation of the banding in the sample. Since martensite transformation is related to the Kurdjumov–Sachs and
Nishiyama–Wassermann (KS/NW) orientation relationship, the volume change may also occur in a preferred orientation that dilates the sample in an oval fashion. A similar effect may be observed for on-heating portion of thermal cycle analysis due to preferred orientation of austenite with respect to initial ferrite microstructure. It has been shown using in-situ neutron diffraction that the austenite that forms under rapid heating conditions may also show preferred orientation [129].

Figure 5.22: High Hard simulated single pass HAZ thermal cycle dilatometry curves
5.3.3. **Microstructure Characterization**

In order to investigate the results the phase transformation analysis presented above, the characterization of the materials by hardness, optical microscopy, scanning electron microscopy, and energy dispersive spectroscopy (EDS) were used. The results of this characterization are presented in sections below.

5.3.3.1. **Hardness**

The hardness histograms for the CGHAZ, FGHAZ, ICHAZ, and SCHAZ for flash process and high hard are seen in Figure 5.23 and Figure 5.24, respectively. The levels of softening for both steels are consistent with those found in GMAW (Figure 5.15). As seen in the dilatometry data, the two steels vary differently depending on the peak temperature. It is normally expected to have a relatively uniform hardness throughout a specific temperature, but it is evident there are multiple effects occurring within both samples. It is first noticeable that the maximum softening of FP is greater with lowest hardness of 175 HVN occurring when heated intercritically to 800°C. However, a different softening mechanism is seen in the HH steel. The maximum softening of 290 HVN is realized when subcritically heated to 700°C. The other peak temperatures are all within the same hardness level of approximately 500-600HVN.
The flash process softening in Figure 5.23, is seen in all regions of the HAZ soften compare to the base metal, but the intercritical temperature of 800°C showed the
maximum extent of softening. When this study initially began, the aim was to only look at one peak temperature for each region, however, when analyzing the dilatometry curve of the 800°C sample (Figure 5.20), it was noticed that there was potential for full austenization so another peak temperature was analyzed at 770°C. As shown in Figure 5.25, the hardness of 770°C showed less softening and the dilatometry showed decreased amount of austenite formed on-heating. This increase in hardness lead to the testing of several intercritical temperatures for both steels to see the effect on hardness as seen in Figure 5.25. The result is clear evidence FP steels show maximum softening when heated intercritically to a peak temperature closest to the Ac3 temperature. This can be explained as following. When heating to the peak temperature, most of the initial microstructure transforms to austenite. The size of these austenite grains is expected to be small due to the relative low temperature and time. This fine grain austenite, in combination with low alloying content, allows for its transformation to allotriomorphic ferrite during slow cooling. This is supported by the dilatometry curves from samples heated to a peak temperature of 785°C to 800°C as shown in Figure 5.20. The dilation data on cooling quickly reaches the slope that is equivalent to ferrite, indicating rapid rate of ferrite formation. The increasing percent austenite to ferrite transformation in the samples leads to a range of hardness for intercritically heat-affected zone (i.e., an average hardness of 222 ±10 HVN and minimum of 208HVN for 755°C peak temperature and average of 195 ±13HVN and minimum of 174HVN for the 800°C peak temperature sample). It is important to notice there is still a large scatter in the hardness distribution
for all of the samples. This has to do with the heterogeneous nature of the initial microstructure that will be discussed in the next section.

![Hardness Histogram of FP ICHAZ and SCHAZ peak temperatures](image)

*Figure 5.25: Hardness Histogram of FP ICHAZ and SCHAZ peak temperatures*

The softening seen in the HH samples is clearly shown to happen below the $A_{c1}$ temperature in Figure 5.24 above. Regardless of the fraction transformed to austenite on heating, the intercritical HAZ always transformed to martensite as noted by the dilatometry curves in Figure 5.22. The chemical compositions listed in Table 4.1, shows the overall increase in alloying elements compared to FP, especially nickel which will stabilize the austenite [27]. This additional alloying content allows the retention of the austenite during fast cooling during welding until driving force for the displacive martensite transformation is achieved. Equilibrium calculations also support these differences in transformation behavior. Both FP and HH steels have similar $A_3$
temperatures near 800°C. However, the A1 temperature of HH (682°C) is significantly lower than that (735°C) of FP steel. It is also important to note that while the 720°C peak temperature is designated as subcritical, the dilatometry data suggests there is possible austenite formation on heating leading to fresh martensite formation on cooling. As a result, this sample shows slightly higher hardness seen in Figure 5.24.

The reduction in hardness for FGHAZ and CGHAZ between FP and HH was also analyzed. The CGHAZ and FGHAZ of FP steels show more variability, compared to HH steels as shown in Figure 5.23 and Figure 5.24. As mentioned earlier, HH steels are highly hardenable and are expected to transform to martensite for all cooling conditions. Therefore, the lack of variability in HH steel hardness distribution (Figure 5.24) is not surprising. However, the range of hardness for FP steels (Figure 5.23) is not readily known. The variations of peak temperatures are expected to change with respect to the prior austenite grain size. Using the kinetic models [82], software available online [130] was used to evaluate the cooling condition for different observed prior austenite grain sizes in FP and HH samples. Figure 5.26 shows the results for both FP and HH using prior austenite grain size of 20 and 50μm (i.e. example coarse grain and fine grain prior austenite grain sizes). The predictions show a large variation in microstructure and hardness for FP steels (e.g., 210 VHN for 20μm austenite grain size and 419 VHN for 50μm austenite grain size). This helps to validate the findings that with decreased prior austenite grain size, more allotriomorphic ferrite will be able to nucleate and grow and create a softer microstructure in FP with finer austenite grain size. However, due to chemical composition favoring austenite stabilization in HH, HAZ microstructure with
predominantly martensite microstructure is predicted. This leads to little variations in hardness for the HH microstructure (e.g. 509VHN for 20μm austenite grain size to 539 VHN for 50μm austenite grain size). The results presented show similar results between higher Ni content HSLA-100 and lower Ni-content 80 studied by Shome et al. [19].
Figure 5.26: a) FP 20μm prior austenite grain size (b) HH 20μm prior austenite grain size (c) FP 50μm prior austenite grain size (d) HH 50μm prior austenite grain size [150]
The hardness was also analyzed through mapping technique in 2 dimensional space. Selected hardness map plots and corresponding macrographs can be seen in Figure 5.27 and Figure 5.28 for flash process and high hard, respectively. It is first important to notice, since these samples are round; they are all oriented in random directions to the original through thickness plate from which they were machined. However, it is safe to assume that the banding is similar to that of the base plate in Figure 5.9. Therefore, the macrosegregation should be in the middle of the through thickness. For example, in the FP ICHAZ hardness map at 740°C, the sample’s thickness is along the abscissa, while the macrosegregation centerline running vertically with the ordinate. These hardness maps clearly show the presence of banding within the material and can help explain the wide distribution in hardness and variability in the measured critical temperatures $A_{c1}$, $A_{c3}$, and phase transformation start and finish temperatures. More explanation of these hardness maps will be discussed in the next section.
Figure 5.27: FP hardness maps and corresponding macrographs
Figure 5.28: HH hardness maps and corresponding macrographs
5.3.3.2. Metallographic Analysis

In order to evaluate the hardness histograms, maps and dilatometry data, microscopy imaging (both optical and scanning electron) was performed to analyze specific regions relating to the different hardness distribution in each of the samples. The layout of this section will be divided between flash process and high hard and then microstructure and discussion of important peak temperatures from Table 4.3 will be analyzed.

**Flash Process**

**Coarse Grain Heat-Affected Zone: 1300°C**

The CGHAZ dilatometry curves in Figure 5.20 showed the beginning of phase transformation at 547°C (near the Bs of 536°C) and finishing near 265°C denoting presence of both bainite and martensite in the steel. On-heating there is also a change in slope in the austenite region relating to the dissolution of the initial enriched Cr M₃C carbides around 1200°C that is in accord with another dilatometry studies involving dissolution of high temperature carbides in martensitic stainless steels [131]. Analysis similar to that of Eldis [128], showed about 50% bainite and 50% martensite should have formed based on the curve. The microstructure observed in Figure 5.29 shows both of these microstructures present in different regions. Martensite and micro-level banding is present in the harder regions (Figure 5.29c), while a mixture of bainite and martensite is seen in the softer regions (Figure 5.29d). As will be a trend for the following peak temperatures, the edges of the original plate (top right and bottom left of Figure 5.29a)
show a refined grain size that is due to the initial flash processing of the 4130 steel from the quenching cooling the surface faster than the interior of the plate.

Figure 5.29: Optical Microstructures of FP CGHAZ correlated with hardness map

Further analysis of this fine banding was investigated using energy dispersive x-ray spectroscopy (EDS). Figure 5.30 shows the location of the band within the sample and where the line scan was run with the resulting x-ray energy counts. Due to the low levels of alloying content present and the resolution of this detector will have difficulties in determining accurate results [132]. The fine nature of this banding as seen in the figure could lead to difficulties in the approximate 1 μm$^3$ excitation volume predicted by
Monte Carlo simulations for the parameters utilized [133]. Finer resolution techniques such as wavelength dispersive spectroscopy (WDS) capable of analyzing at smaller spatial resolution could achieve better results than those presented.

Figure 5.30: Qualitative EDS profile of banding in FP CGHAZ

**Fine Grain Heat-Affected Zone: 1100°C**

The dilatometry curve in Figure 5.20 for peak temperature of 1100°C shows the transformation on-cooling starting at 575°C. This is 40°C before the Bs temperature. The tie-line analysis of this result shows approximately 20% ferrite, 75% bainite and 5% martensite for the microstructure. Figure 5.31 shows the hardest (573HVN) and softest
(271 HVN) regions of the sample in (a) and (b), respectively. Once again, the hardest region has prominent banding present. The BSE image in Figure 5.31c shows the banding and coarse martensitic lath structure present. Recalling the hardness map from Figure 5.27, there are only a few areas with high hardness, correlating well with the dilatometry data for low percentage martensite and observed in the material. The BSE image also helps illustrate the numerous carbides still present after this thermal cycle. This supports the recent work to understand Cr-enriched carbides not dissolving during flash processing [29]. In Figure 5.31b and d, ferrite and carbides can be seen in addition to lath structures that are assumed to be bainite based on the hardness in this region.
Figure 5.31: Microstructure of FP FGHAZ 1100°C

Fine Grain Heat-Affected Zone: 1000°C

Dilatometry results for the FGHAZ at 1000°C (Figure 5.20) show that the transformation begins around 565°C and finishes below the \( M_s \) temperature at 347°C. This gives a predominantly ferrite and bainitic microstructure, with less than 5% martensite. The microstructures in Figure 5.32 verify these microstructures in both the hardened and soft regions. In Figure 5.32d-e the microstructure is mostly ferrite with small laths evident in the SEM image. The harder regions contain larger lath structures...
with prior carbides. As evidenced by the histogram in Figure 5.23, the hardness distributions for 1100°C and 1000°C are very similar and the microstructures and dilatometry confirm this data.
Figure 5.32: Microstructure of FP FGHAZ 1000°C
Fine Grain Heat-Affected Zone: 900°C

The 900°C peak temperature dilatometry curve is seen in Figure 5.20 showing the formation of ferrite in the temperature range of 725°C to 595°C with no other reactions. However in the hardest map and histogram the maximum level of hardness is 505HVN, indicating martensitic or bainitic microstructure should be present. When examined microscopically (Figure 5.33b-c) lathes are still found within areas of high hardness. However, the dilatometry did not observe these phase transformations and the curves showed full austenization on heating. One possible explanation is plausible, but it is uncertain without further analysis. The explanation is these are martensite-austenite (M-A) constituents. The higher hardness and segregation already present from the initial material would allow regions of highly enriched carbon in austenite allowing for the retention of austenite. M-A constituents are also known to have $M_s$ temperatures below room temperature [28, 84], allowing the dilatometry to miss the transformation as the tests were stopped after approximately 150°C because the cooling rate could not be maintained by the copper jaws in an adequate time. Advanced techniques such as in-situ time resolved x-ray diffraction and laser scanning confocal microscopy could be used to observe these changes in real time to find the real reason behind this discrepancy [134, 135].
Figure 5.33: Microstructure of FP FGHAZ 900°C
Intercritical Heat-Affected Zone: 800°C

The dilatometry curve for FP ICHAZ 800°C in Figure 5.20 shows an almost complete transformation to austenite on-heating and fully ferritic transformation on-cooling. The hardness map shows the retention of the mid-banding present in the base material and other HAZ areas already analyzed. It is in this region (Figure 5.34b-d) that we see the hardest microstructure. It is difficult to determine from optical microscopy what the constituents in the dark etching (green square in Figure 5.34b) and brown-orange (blue square in Figure 5.34b) are, so electron microscopy is needed. As seen in electron microscopy images, there is extensive carbide precipitation on the fine ferrite grain boundaries (dark etched region Figure 5.34c and f). There is a visual correlation that regions with more of these carbides, in addition to the original spheroidized carbides, have higher hardness. It is possible this is the formation of fine pearlite with the lamellar nature of the cementite forming after ferrite rejects carbon and alloying content into the austenite not yet transformed. The orange-brown etching constituent seen in the SEM (Figure 5.34d) shows a potential blocky M-A constituent that lies along the enriched banded region. If these are indeed M-A constituents, this continuous line within the band could result in detrimental failure due to their known low toughness value. As noted above, the hardness in this region is lowest of all examined HAZ peak temperatures because almost all of the initial microstructure is austenitized and subsequently transformed to ferrite, whereas the other peak temperatures will be shown to contain more tempered martensite as there will be more regions not transformed to austenite on-heating.
Figure 5.34: Microstructure of FP ICHAZ 800°C
Intercritical Heat-Affected Zones: 785°C

The dilatometry data for peak temperature 785°C shown in Figure 5.20 shows the continuing trend of decreased transformation to austenite. The on-cooling data also shows less of a plateau during the ferrite transformation as a result. Both of these observations are confirmed by the microstructure changes seen in Figure 5.35. The harder regions (Figure 5.35b-d) show the both the presence of banding (Figure 5.35b) and tempered martensitic microstructure (Figure 5.35c). As mentioned above, the hardness seems to increase consistently with the dark etching regions seen optically that contain either original carbides or those nucleated during tempering of martensite or on-cooling formation of pearlite. Figure 5.35d-e confirms this by seeing more ferrite present in the hardness region.

A qualitative EDS line scan analysis was done on the banding seen in Figure 5.35b, as shown in Figure 5.36 (white line is the location of the line scan). There are peaks of Cr and Mn present within the band, but there is also a good amount of noise from the submicron size of particles affecting the data. As mentioned in the CGHAZ section above, better analysis may be obtained with techniques like WDS.
Figure 5.35: Microstructure of FP ICHAZ 785°C
The dilatometry curve for the ICHAZ with peak temperature of 770°C (Figure 5.20) shows slightly less transformation to austenite than the previously examined 785°C peak temperature. In fact, the hardness distribution (Figure 5.23) shows about the same distribution of hardness between the two temperatures. Qualitatively, the microstructure (Figure 5.37) appears nearly the same as well with combination of tempered martensite (Figure 5.37e), numerous carbides, and banding. The main difference observed between these two temperatures (besides the large banded area in Figure 5.36) is the increased banding seen even in the lower hardness region (Figure 5.37d).

**Intercritical Heat-Affected Zone: 770°C**

The dilatometry curve for the ICHAZ with peak temperature of 770°C (Figure 5.20) shows slightly less transformation to austenite than the previously examined 785°C peak temperature. In fact, the hardness distribution (Figure 5.23) shows about the same distribution of hardness between the two temperatures. Qualitatively, the microstructure (Figure 5.37) appears nearly the same as well with combination of tempered martensite (Figure 5.37e), numerous carbides, and banding. The main difference observed between these two temperatures (besides the large banded area in Figure 5.36) is the increased banding seen even in the lower hardness region (Figure 5.37d).
Figure 5.37: Microstructure of FP ICHAZ 770°C
**Intercritical Heat-Affected Zone: 755°C**

The peak temperature of 755°C shows only slight transformation to austenite in Figure 5.20. The resultant microstructure contains highly enriched regions of austenite along the grain boundaries of the initial-prior austenite grain along with carbide precipitation within the tempered martensite grain (Figure 5.38). These highly enriched austenite regions transition to M-A constituents, aggregates of carbides, and possibly pearlite. As noted in the background, if these are indeed M-A constituents enveloping the grains, the toughness and mechanical properties of this region can be expected to be reduced.
Figure 5.38: Microstructure of FP ICHAZ 755°C

Subcritical Heat-Affected Zone: 740°C

As seen in Figure 5.20, the dilatometry for the peak temperature of 740°C shows the potential for reaching the $A_{c1}$ temperature, but almost no transformation to austenite on-heating. Since there is limited austenite transformation on heating, the predominant microstructure is tempered martensite as seen in the SEM picture (Figure 5.39e). The microstructure in Figure 5.39b-c shows banding containing numerous carbides and tempered lath structures consistent with the increased hardness of the other HAZ regions.
From the dilatometry curves in Figure 5.40, the peak temperature of 720°C shows no dilation response, denoting no observable phase transformation on-heating or on-
cooling. Since there is no chance of austenite formation and is heated to a lower temperature, the hardness is higher than its other SCHAZ counterpart. The only microstructures present are tempered martensite, bainite, and initial carbides as seen in Figure 5.47c-f. The optical micrographs show the prior austenite grain size well due to the precipitation of carbides along those grain boundaries that can also be seen in the SEM images.
For each peak temperature above $A_{c1}$, HH steel showed a transformation to martensite. The hardness levels remained relatively similar between the different peak
temperatures, with differences described below that deal with partial austenization and austenite grain size.

**Coarse Grain Heat-Affected Zone: 1300°C**

The dilatometry curve for the Figure 5.22 shows a completely martensitic transformation that begins at 396°C and finishes at 165°C. The microstructure can be seen in Figure 5.41 and is consistent with fully martensitic observation by the dilatometer. A clear image of the prior austenite grain size optically and with the SEM can be seen in Figure 5.41d and f. The predicament in this case is the hardness ranges from 631HVN to 505HVN as seen in the map below and seen within the initial base metal in Figure 5.10. The microstructure appears to have the similar prior austenite grain size when comparing hard with soft regions (Figure 5.41b and c, respectively), so the only likely explanation is a difference in alloying content due to macrosegregation mentioned earlier. It is also noticeable that harder regions contain more TiCN (Figure 5.41e) than the other areas, that may be aiding in deformation resistance when hardness measurements are made, in addition to their primary role of grain refinement [20].

Similar to the micro-level banding seen in FP, HH displays similar bands that are more numerous than those seen in FP. EDS analysis was performed and the results are presented in Figure 5.42. A qualitative trend of segregating alloying elements is increased in this region that is much more pronounced than that of FP (Figure 5.30). One possible reason is the widths of these banded regions are wider than that of FP, and therefore within the resolution of the EDS detector.
Figure 5.41: Microstructure of HH CGHAZ 1300°C
Figure 5.42: Quantitative EDS profile of banding in HH CGHAZ 1300°C

**Fine Grain Heat-Affected Zone: 1100°C**

The fine grain HAZ for HH has a decrease in hardness owing to the refined austenite grain size that is seen both compared to CGHAZ in Figure 5.41 and between (c) and (e) in Figure 5.43. The austenite grain size varies within the FGHAZ sample from 5-20μm between the softened and hard regions, respectively. The same trend of completely martensitic microstructure is observed as well.
Figure 5.43: Microstructure of HH FGHAZ 1100°C
Intercritical Heat-Affected Zone: 800°C

The dilatometry curve for ICHAZ 800°C presented in Figure 5.22 shows almost complete transformation to austenite on-heating and shows complete martensitic transformation on-cooling starting at 340°C. The microstructure seen in Figure 5.44 verifies the martensite present. The difference of hardness within this region again, is not very understandable from the microstructures as they are quite similar, and may have to do with the initial composition throughout the material. Another interesting observation is the soft region occurs where the micro-level banding occurs. This is opposite of the behavior seen in the flash processed steel.
Figure 5.44: Microstructure of HH ICHAZ 800°C
Intercritical Heat-Affected Zone: 770°C

Similar to the microstructure and hardness level of ICHAZ 800°C, ICHAZ 770°C has less transformation to austenite on-heating, but maintains approximately the same hardness. This trend is continued throughout the rest of the HH ICHAZ temperatures as the amount of tempering decreases the hardness locally, while the transformed austenite to martensite provides the higher level hardness. Figure 5.45d shows a representative tempered martensite microstructure within the softer region of the microstructure.

Figure 5.45: Microstructure of HH ICHAZ 770°C
Subcritical Heat-Affected Zone: 720°C

The main softening seen in the HH steels as mentioned previously occurs below the $A_{c1}$. The dilatometry curve for the SCHAZ 720°C in Figure 5.22 shows very minimal, if any, transformation to austenite on-heating. On-cooling, however, there is some dilation change that occurs well below the $M_s$ temperature at 230°C. A possible explanation for this dilation was initially thought an error, but 3 subsequent experiments all showed this transformation within the range of 180-230°C. There are a couple explanations for this observed change. The first involves M-A constituent formation that is potentially visible in Figure 5.46e. This figure shows a prior austenite grain with potential growth of new austenite on the grain boundaries that has only partially transformed and could be an M-A constituent. This would be a source of poor toughness and ballistic quality in addition to the decreased hardness seen in this region. The other possibility is there is originally retained austenite within the material that is being transformed on cooling, but this is highly unlikely given the material is tempered initially. The harder region shown in Figure 5.46b-c shows more TiCN in the harder regions as previously seen in the other thermal cycle treatments in their respective harder areas of the sample.
Figure 5.46: Microstructure of HH SCHAZ 720°C
Subcritical Heat-Affected Zone: 700°C

The final investigation looked into a purely subcritical thermal cycle as evidenced by the dilatometry curve in Figure 5.22. The distribution in hardness in this sample (±60HVN from min to max) was not as severe as the other samples (±100HVN) mainly due to the only transformation occurring is the further tempering of the martensite. Figure 5.47 shows both regions of hardness exhibiting precipitation of carbides at prior austenite grain boundaries, within the lathes and other regions described in Figure 2.25. It is also important to note there do not appear to be any constituents along the prior austenite grain boundaries like there was for 720°C peak temperature, furthering evidence that those were potentially reaustenitized regions.
5.3.4. Summary

This data suggest the solution to FP softening may lie in increasing the austenite
stabilizing elements to retard the ferrite formation and retain higher hardness levels. One can
hypothesize that will also be beneficial to high velocity impacts, as demonstrated in other steels [6, 18, 20]. However, with increased nickel content, the initial benefit of flash process mixed microstructure of martensite, bainite, and carbides may be lost. As a result, the property that enables it to successfully pass various ballistic requirements may also be lost. Therefore, further research is required to optimize the FP steels to regain the softening observed in the HAZ.

After examining multiple flash process and high hard steel samples subjected to various HAZ thermal cycles to investigate the softening, it was discovered that in FP steel, the critical softening occurs when heating intercritically close to the $A_{c3}$ temperature. Microstructure evaluation and dilatometry showed this was due to the formation of ferrite on cooling from austenite formed on-heating.

In contrast, high hard steel only softened when subcritically heated below the $A_{c1}$ temperature. This was attributed to an addition of austenite stabilizing elements that promoted austenite retention until the $M_s$ temperature was reached and fresh martensite was formed. This was also confirmed with dilatometry and microstructure observation.

5.4. **High Energy Density Laser Welding**

After understanding the microstructure evolution during low-heat input welding, the investigation turned to using another process that could limit the heat input further. From the background, it was shown HAZ softening was reduced in a variety of steels including armor [59], dual phase [13, 15, 74], pipeline [77], and TMCP steels [12, 79] by high energy density processes like laser welding. Secondary benefits of using a high energy density process included the ability to make full penetration welds autogenously
and fast welding times. The autogenous weld is also beneficial for armor applications because this is another area where the hardness can remain higher than filler metals can safely be before cracking tendencies would develop.

5.4.1. Bead on Plate Welding

Since there is not an initial standard for laser welding of armor type materials, bead on plate (BOP) welds were made on each material to find the best balance between fast travel speed and avoiding welding defects. Figure 5.48 and Figure 5.49 show the macrographs for the welding conditions used for the BOP welds. It is first noticeable that the flash process welds seem to have larger porosity in the weld metal than high hard steel. While this may be by chance depending on where the cross section is located, there may be another unknown reason for this behavior.

The travel speed of 120ipm is too fast as noted by the root humping in both steels. Root humping leads to a loss in weld joint performance and forms because of capillary instability and fluid flow of the weld pool [115, 116]. Therefore, the 120ipm was deemed too high for use in the actual butt welds. 100ipm is on the edge of the parameter window as well because of the undercut seen that is early warning sign of potential root humping even though it was not seen at this travel speed. Further analysis with hardness mapping and optical microscopy to examine the microstructure was performed to determine the best travel speed to make the butt welds.
5.4.1.1. Microstructure Characterization

Based on the macro-level observation that 120ipm creates root humping due to the excess travel speed, hardness maps (Figure 5.50) were made for the lowest and highest travel speed of 60ipm and 100ipm to evaluate the level of softening. As noticeable in Figure 5.50, the width of softening to base metal levels of 530 HVN is
obtainable within 2mm of the weld for 100ipm travel speed, while the softening is still effective 4mm away from the fusion line in 60ipm conditions. It is also noticeable again that FP has a wider width of softening than HH. This could be due resistance to tempering HH has with the increased nickel content [61] and the tempering of bainite already present in the FP steel.

Figure 5.50: BOP laser weld hardness maps (a) FP 60ipm (b) FP 100ipm (c) HH 60ipm (d) HH 100ipm

The microstructure of the flash process 100ipm weld is shown in Figure 5.51 below. The microstructure close to the fusion line still has the original base material
carbides present. This is in contrast to the GMAW conditions, where dissolution of carbides would occur in the CGHAZ and some remained in the FGHAZ. The softening starts at the dark etching line as seen in Figure 5.51a and Figure 5.51c. Since the heat input is lower and cooling rate is faster, the time at temperature is reduced during laser welding allowing less time for the carbides to dissolve. In addition to the carbides not dissolving, the cooling rate also prevents the austenite to ferrite transformation. This allows for the increased hardness within the HAZ to transform to completely martensite. There is coarsening of the carbides in addition to tempering of the martensite seen in the intercritical/subcritical HAZ. Figure 5.51b shows the fusion line with epitaxial growth from the HAZ that contains fresh martensite and correlates with high hardness above 500HVN. The microstructure seen in Figure 5.51c-d is that of tempered martensite and bainite from the initial base material with more tempering occurring closer to the weld metal due to higher peak temperatures present.
Figure 5.51: Microstructure of FP BOP 100ipm laser weld

Figure 5.52 shows an overview of the fusion zone and HAZ for HH. In Figure 5.52a, extensive banding was noticed in the HAZ and base metal. The microstructure in Figure 5.52b shows martensitic lath structure in the HAZ and epitaxial columnar growth into the fusion zone. The presence of TiCN is also seen throughout the initial material and is retained even after the welding has occurred. Micrographs in Figure 5.52c-d show the tempered martensitic region with lower hardness in the 340HVN range.
5.4.2. Zero-Gap Butt Welds

From the BOP experiments, the final choice of 90ipm was chosen for the zero-gap butt welds. 90ipm was chosen as the highest speed while trying to avoid the undercut that was seen at 100ipm. After welding was completed, radiography was performed to detect welding flaws and defects. The only defect found in both welds (circled in yellow in Figure 5.53) was lack of penetration at the beginning of the weld. However, this is expected due to the programming of the welding to ramp up the current within the first 1mm of travel to avoid blowout of the weld pool due to excess power on start-up.
5.4.2.1. Microstructure Characterization

The hardness and macrographs of the zero-gap welds can be seen in Figure 5.54. The hardness scale is the same from 300HVN-600HVN. Comparatively, the FP weld has more softening than the HH sample, likely due to the initial microstructure differences. Bainite will temper to a lower hardness and average in lower than the tempered martensite from HH. The maximum, minimum, and average hardness can be seen in Table 5.1. Also noticeable from the HH map is the presence of the macro-level hardness difference in the unaffected areas that is consistent with the base metal and simulated HAZ hardness distributions of softer near the middle of the thickness and harder towards the outside.
Figure 5.54: Hardness map and corresponding macrograph of zero-gap laser butt weld (a) flash process (b) high hard

Table 5.1: Hardness Values for zero-gap butt welds

<table>
<thead>
<tr>
<th>Material</th>
<th>Hardness Maximum</th>
<th>Hardness Minimum</th>
<th>Hardness Average</th>
</tr>
</thead>
<tbody>
<tr>
<td>Flash Process</td>
<td>669</td>
<td>295</td>
<td>477</td>
</tr>
<tr>
<td>High Hard</td>
<td>612</td>
<td>337</td>
<td>503</td>
</tr>
</tbody>
</table>

Weld Metal

Since the weld is autogenous and the cooling rate is in excess of 100°C/s, the hardness in and next to the fusion zone exceeds that of both materials due to complete martensitic microstructure (Figure 5.55). An interesting microstructural feature in the HH weld metal is the segregation banding that is seen in the weld metal. This helps understand that perhaps this composition inherently segregates strongly on solidification
and lead to problems that exist throughout this research in terms of compositional differences leading to widespread hardness variation throughout the material.

Figure 5.55: Weld metal microstructure of zero-gap laser butt welds

**Flash Process**

The hardest region in the FP laser weld occurs in the HAZ approximately 200μm from the fusion line. The microstructure of this area can be seen in Figure 5.56. The hardest point of 669HVN shows banding, extensive carbides, and martensitic laths within the microstructure. The point immediately to the left of this hard indent (Figure 5.56e-f) is 410HVN and shows vastly tempered martensite microstructure within the banded region. The average hardness of an indent in a similar location is around 300HVN, but since this indent lies within the banded region, it has higher hardness, in agreement with the other areas examined in the flash processed steel. The softest point in the laser weld can be seen in Figure 5.57. The maximum softening occurs in the intercritical region.
with hardness of 296HV. The microstructure is mostly tempered martensite and carbides. Due to the fast cooling rate, there is not enough time for the nucleation and growth of ferrite from the austenite formed on-heating. This allows the transformation to martensite to occur. Using the online HAZ simulation model [130], with a cooling rate of 100°C/s and a very fine prior austenite grain size of 5μm, the prediction is a hardness over 450HV, as evidenced in most of the HAZ (Figure 5.58). The reason for the softer microstructure found is most likely the cooling rate further from the fusion line is lower and this allow the transformation of bainite. This is a potential area for future work to examine laser welding thermal cycles similar to the single-pass GMAW simulations performed.
Figure 5.56: Microstructure seen in the banded region of the laser welded FP steel
Figure 5.57: Softened HAZ in FP zero-gap laser butt weld
The microstructure for high hard is seen in Figure 5.59 showing the two opposite hardness regions. The hard regions (Figure 5.59b-d) contain lath martensite that has similar hardness to that of the GMAW HAZ and single-pass simulation maximum hardness. Also similar to the GMAW and single-pass simulations, the peak softening occurs beyond the A$_{c1}$ as seen in Figure 5.59e. This again, is different from FP where the maximum softening occurs above the A$_{c1}$ in the intercritical temperature range. The softening seen in the HH is 337HVN. This is about 87HVN harder compared to the maximum softening seen in GMAW of 254HVN that is mostly due to less exposure to elevated temperatures that soften the steel during tempering. The width of the softened layer is also reduced dramatically for the same reasons.
Figure 5.59: High Hard microstructure form zero-gap laser butt weld
Summary of Microstructure Characterization

The ability to avoid the ferrite transformation in FP to increase the hardness of the HAZ from 180HVN to 300HVN in laser welding is a great benefit. A comparison of the welded regions of GMAW and laser welding is seen in Figure 5.60 that show the recovery of base metal hardness within 3mm of the laser weld, while the hardness map for the GMAW is not large enough to show the location of the recovery to base metal hardness.

While the laser welding still has softening, the extent is seen as much less, and softening will be unavoidable in any process that will heat the steel above the A1 temperature for an appreciable time. This would include solid state processes that usually rely on operating at temperatures in the softened austenitic state in order get material flow (e.g. friction stir welding [136]). Therefore, high energy density processes like laser welding are seen as the currently best available process to keep heat input to a minimum to avoid the HAZ softening. In addition, studies have suggested the best microstructure against ballistic tests is that of tempered martensite and a minimal sized HAZ [5, 6, 17, 59].
5.4.2.2. Mechanical Properties

The result of the tensile testing for the laser welds is seen in Figure 5.61. The yield strength of the two materials is very similar (205ksi), owing to the initial yielding occurring in the softened portion of the HAZ. It is noticed that the UTS of HH (225ksi) is higher than that of FP (215ksi). HH also had less softening, that may be the result of two reasons. The first is the increase in alloying content that would slow the tempering kinetics of the martensite. The second reason is the tempering of both the initial bainite and martensite results in a lower strength in flash processed 4130. This is due to the reduced constraining of bainite by the martensite that initially supports the increased strength of the mixed microstructure [33]. The elongation for both materials was too similar to assume one had better elongation compared to the other. Comparatively to the GMAW (Figure 5.17), the laser welding had much higher strength for both steels. This can be attributed to the reduction in softening and the elimination of the undermatched filler metal.
Figure 5.61: *Stress-Strain curve for laser welding of flash process and high hard*

The fracture surfaces of both steels were very similar as seen in Figure 5.62. The interior of the fracture exhibited brittle cleavage fracture that have little to no ductile behavior to it (Figure 5.62b,e). The failure at the edges near the surface showed shallow ductile dimples along with cleavage and shear deformation (Figure 5.62c,f). As mentioned above, the plastic deformation seen in the softened HAZ area started in a x-type fashion as mentioned by Panda [125]. However, because the microstructure is primarily martensite, there wasn’t much ductility present, and brittle failure occurred before the ductile ‘ridge’ could be realized as seen in Figure 5.14a. A combination of the tensile test results from Figure 5.13, Figure 5.17, and Figure 5.61 are shown in Figure 5.63.
Figure 5.62: Laser Welding Fracture Surfaces: (a-c) FP (d-f) HH Both (b) and (e) are from the center of the cross sections while (c) and (f) are from the edge areas.
5.4.3. Summary of Results

Laser welding reduced both the width of softening in both materials and maximum extent (Figure 5.60). Original tempered microstructure lead to the narrow region of softening. Since an undermatched filler material was not needed for the autogenous laser welding, the weld metal was fully martensitic. Failure during tensile testing occurred in the softened HAZ for both materials. The fracture surface was primarily brittle in nature with cleavage cracks along with some shallow ductile dimples in the plastically deformed region near the surface of the metals. Tensile testing showed reduction in strength and ductility for both welding processes compared to the base metal. Laser welding retained higher strength, but lower ductility compared to GMAW.
Chapter 6: Conclusions

The main goal of this research was to determine the weldability of flash processed steels, in particular for an armor application. By knowing the background of the material and analyzing existing knowledge of welding various steels, various welding experiments and simulations were carried out to determine the weldability of flash processed steels. The results are as follows:

1. Flash process steel softens (540HVN to 170HVN) when welded with low heat input GMAW conditions
   a. After doing physical simulations of HAZ thermal cycles on flash process and high hard armor, it was discovered that in FP steel, the critical softening occurs when heating intercritically close to the $A_{c3}$ temperature. Microstructure evaluation and dilatometry showed this was due to the formation of ferrite on cooling from austenite formation on cooling.
   b. In contrast, high hard steel only softened when subcritically heated below the $A_{c1}$ temperature. This was attributed to an addition of austenite stabilizing elements that promoted austenite retention until the $M_s$ temperature was reached and fresh martensite was formed. This was confirmed as well with dilatometry and microstructure observation.

2. Extensive banding both at macro hardness level and on a microsegregation level was seen leading to heterogeneous microstructural properties.
a. The difference throughout the material poses an issue with predictive and modeling tools to predict the behavior of the material when subjected to welding thermal cycles. With heterogeneous microstructure and hardness seen, it is difficult to predict the mechanical properties after welded (i.e. ballistic resistance, tensile properties, toughness)

3. Laser welding improved tensile strength (120ksi to 215ksi) and reduced extent and width of HAZ softening (170HVN to 300HVN and 25+mm to 6mm, respectively) of as-welded flash process steel

    a. However, failure still occurred in the HAZ and reduction in strength and ductility was 45ksi and 8% compared to the base metal.
Chapter 7: Future Work

7.1. Multi-Pass Heat-Affected Zone Simulations

The original work analyzed the effect of single pass welds, which works well for sheet thickness materials. However, for armor applications and other structural applications, multi-passes are necessary for a fully welded joint due to relative low penetration capable in fusion welding processes. As mentioned in the background for various steels, significant softening and deleterious phases such as M-A constituents are more prone to form in these multi-pass conditions and have already been observed in the single-pass simulations. Therefore, with the implication of potentially worse properties than single pass and necessity for multi-pass welding in thicker sections, multi-pass conditions should be analyzed.

7.1.1. Thermal Cycles

To evaluate the effect of multiple thermal cycles on a worst case scenario, only simulations using the first peak temperature of 1300°C were analyzed. After that, peak temperature of 900°C, 800°C, and 720°C with consistent heating rate of 100°C/s and a $\Delta t_{8/5}$ of 30s. The various experiments explained are shown in Table 7.1 with an example thermal cycle of a triple pass simulation is shown in Figure 7..
### Table 7.1: Peak Temperatures used in multiple thermal cycle simulations

<table>
<thead>
<tr>
<th>HAZ simulation Name</th>
<th>Peak Temperatures</th>
</tr>
</thead>
<tbody>
<tr>
<td>CG+FG</td>
<td>1300°C+ 900°C</td>
</tr>
<tr>
<td>CG+IC</td>
<td>1300°C+ 800°C</td>
</tr>
<tr>
<td>CG+SC</td>
<td>1300°C+ 720°C</td>
</tr>
<tr>
<td>CG+FG+IC</td>
<td>1300°C+ 900°C+ 800°C</td>
</tr>
<tr>
<td>CG+FG+SC</td>
<td>1300°C+ 900°C+ 720°C</td>
</tr>
</tbody>
</table>

**Figure 7.1: Triple-pass thermal cycle with peak temperatures at 1300°C, 900°C, and 800°C**

1.1.1. **Phase Transformation Analysis**

Using a similar approach to that used in the single-pass simulations, the microstructures formed during the thermal cycle and the final resultant microstructure can be determined. Figure 7.2 on the following page shows the dilatometry results of the various reheat peak temperatures. It is important to note the rainbow scale for the dilation data to better understand that as the color progresses from purple to red, the time that has transpired increases. So, for example in the CG+FG sample, the purple region includes the on-heating transformation to austenite with peak temperature to 1300°C and the subsequent transformation on-cooling to bainite and martensite. The green portion of
the curve is the on-heating portion of the second thermal cycle with peak temperature of 900°C. The yellow and red portions are the on-cooling of this second thermal cycle showing the ferrite and bainite transformations occurring. Further discussion of these thermal cycles will be discussed in the characterization section below.
Figure 7.2: Dilatometry curves for multiple thermal cycles simulated on flash processed steel
7.1.2. **Hardness**

Hardness mapping was also employed to analyze the multiple thermal cycles and the resultant histogram and comparative maps can be seen in Figure 7.3 and Figure 7.4, respectively. The hardness ranges for all samples are all closely related relatively to the single-pass CGHAZ and FGHAZ 900°C seen in the hardness histogram. In addition, the hardness maps still show heterogeneity present within the samples which will be discussed further in the characterization section.

*Figure 7.3: Hardness histogram of multi-thermal cycle HAZ simulation on flash process with single-pass CGHAZ and FGHAZ 900°C plotted for reference*
7.1.3. Microstructure Characterization

7.1.3.1. CG+FG Heat-Affected Zone

As seen in the dilatometry curve in Figure 7.2, the first pass transforms first at 550°C and finishes at 360°C, denoting a mixture of bainite and martensite. This first pass is fairly consistent for all simulations. The second pass reaustenitizes the entire microstructure and starts forming ferrite at 691°C and the transformation ends around 390°C. The final microstructure can be seen in Figure 7.5, corresponding to the hardest (440HVN) and softest (258HVN) regions in the microstructure. The hardest area is within a banded region and with SEM observation, evidence of fine lath structures can be
seen (Figure 7.5c). The soft region shows a predominantly ferritic microstructure with fine lath structures (Figure 7.5d-e).
7.1.3.2. CG+IC Heat-Affected Zone

The dilatometry curve in Figure 7.2 shows the incomplete austenization on-heating in the second pass of the 800°C peak temperature consistent with that of the single-pass. However, dilatational change is observed not only for ferrite in this case on-cooling. There is a change in slope around the $M_a$ temperature and again around 200°C. Further testing must be run to validate this is not an error in the measurements. The microstructure presented in Figure 7.6 shows the hard (400HVN) and soft regions (231HVN) of the sample. The harder region is again in a banded region that contains ferrite and tempered martensite consistent with the hardness (Figure 7.6b-c). The soft region is quite similar to that of the CG+FG sample with predominantly ferrite and heavily tempered martensite as detailed in the SEM image with extensive carbide precipitation within the laths and at prior austenite grain boundaries (Figure 7.6d-e).
7.1.3.3. CG+SC Heat-Affected Zone

The dilatometry in Figure 7.2 shows no austenization on the second pass, signifying only tempering of the CGHAZ first pass should occur. In Figure 7.7b, the
microstructure for this sample illustrates very well the difference in grain size that has been previously discussed from the through thickness variation in grain size from the initial microstructure of the FP plate from Figure 5.9. The hard regions transform fully to martensite and are tempered during the second pass. However, there is evidence of M-A constituents along the prior austenite grain boundaries as seen in Figure 7.6c-d that is not seen in the SCHAZ or CGHAZ of the single-pass welds. The only explanation that can be given here is the potential of retained austenite to be present or another reason that needs to be examined further. The softened microstructure in Figure 7.6e-f shows the ferrite present from the dilatometry curve in addition to lathe microstructure that is assumed to be bainite based off of the hardness within this region.
Figure 7.7: Microstructure of flash process CG+SC HAZ
7.1.3.4. CG+FG+IC Heat-Affected Zone

The dilatometry curve in Figure 7.2 shows a different microstructure evolution than expected for several reasons. First, the CGHAZ first pass shows only ferrite and bainite formation with the transformation beginning at 570°C and finishing by 420°C. There is no readily available reason for this occasion as the $t_{8/5}$ was still 30s. Secondly, the 800°C peak temperature on the third pass shows no dilation change on-cooling below 200°C. This may be a delayed martensitic transformation, but this is unknown and needs further evaluation. The microstructure shown in the hard region in Figure 7.8b-c shows banding that has some fine laths within surrounded by tempered microstructures, pearlite and ferrite. This is the first time pearlite has been observed within this sample. This may be due to the disappearance of the initial carbides and enough alloying elements allow for the lamellar formation of cementite and ferrite. The development of the pearlite is more evident in the softened regions seen in Figure 7.8d-e. The hardness is also noticeably higher on the edges of the original plate thickness that can be explained by the refined grain size that would increase the hardness with the primary microstructure of ferrite and pearlite.
Figure 7.8: Microstructure of flash process CG+FG+IC HAZ
7.1.3.5. CG+FG+SC Heat-Affected Zone

The dilatometry curve for the CG+FG+SC in Figure 7.2 is quite similar in the transformations to the CG+FG with the addition of the third peak temperature 720°C thermal cycle that only imposes tempering of the 2nd pass microstructure. The hardness in for this sample is lowest among the five examined thermal cycles. The 220HVN softened region seen in Figure 7.9d-e show coarsened pearlite along with ferrite. Similar to the CG+FG+IC HAZ, there is little evidence of the original spheroidized cementite from the base metal, but the banding is still present as seen in Figure 7.9b. These banded regions seem to be consistent with higher carbon and alloying element areas due to the presence of pearlite in this region seen in the SEM image in Figure 7.9c.
Figure 7.9: Microstructure of flash process CG+FG+SC HAZ
7.1.4. **Summary of preliminary results and need for further testing**

Currently, only flash process has been analyzed and more experimentation to correlate with high hard will give more useful data and explanation. Also, it may be advisable to examine different peak temperatures for the different multiple thermal cycle combinations. While this could lead to extensive testing, a recommendation based off of the single-pass and preliminary multi-pass results would be to examine a couple temperatures between $A_{c1}$ and $A_{c3}$ as the last pass for FP, and temperatures near the $A_{c1}$ for HH.

7.2. **Simulation of Laser Welding HAZ**

Along similar lines already investigated for arc-welding processes, thermal cycle evaluation of the laser welding heating and cooling rates would help identify the peak temperature and cooling rates that create the softened regions seen in the completed work. This would help further understand the microstructure that is present and rationalize the softening behavior seen. Work by Wang and Zhao can be used as a preliminary literature review of necessary conditions to simulate these fast heating and cooling conditions [81, 137, 138].

7.3. **Ballistic Testing**

In order to qualify the welding for ballistics protection, preliminary ballistic testing must be done to evaluate the welds fully. Preliminary ballistic tests (in accordance to NIJ Level III\(^1\) [139]) have been conducted by a third party and the

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\(^1\) NIJ level III states testing is performed with 7.62 mm FMJ, steel jacketed bullets (U.S. Military designation M80) with a specified mass of 9.6 g (147 grams) and a velocity of 847 m/s ± 9.1 m/s (2780 ft/s ± 30 ft/s)
aftermath of these tests have been analyzed with hardness testing and microscopy. Two samples were tested that were GMA and laser welded the same way presented in former chapters. The results (Figure 7.10) show that the bullet was stopped 10mm away from the weld centerline for a laser weld, while the GMAW was stopped 22mm away from the weld centerline. Both are outside of the softened regions of the HAZ, denoting the potential minimal improvement of the maximum extent of softening for the laser welding. Further precise testing must be completed before evaluating any softening as being unsuitable for ballistic testing. A benefit from the laser welding is that the ballistic window is twice as small for laser welding, making it more difficult to find the weakness in the material.

![Figure 7.10: Preliminary ballistic evaluation of GMAW and laser weld](image)

Hu et al. discussed the main methods of ballistic failure and are seen in Figure 7.11 [140]. Most brittle armors fracture according to brittle and radial fracture, while
average thickness and hardened armors fail with plugging and ductile hole growth. The fragmentation failure normally occurs in high strength, but brittle, thick materials.

Finally, petalling is common in thin body-type armors that are ductile. From Figure 7.10, we can see that the failure occurs through the plate by ductile hole growth and the bullets that are stopped have a radial bulge, but no fracture in the GMAW. Laser welding shows evidence of petalling and ductile hole growth in the areas where the bullet went through. Again, when the plate didn’t fail there was radial bulging, but no fracture.

![Figure 7.11: Penetration Modes of ballistic impacted materials [140]](image-url)
Evaluation of the laser welded sample was performed with hardness and microscopy of an area that successfully passed outside of the softened region and a region along with two propagated cracks near the HAZ, as seen in Figure 7.12 and Figure 7.13, respectively. The region from which the cracked sample is taken is shown in Figure 7.14 while the successfully tested sample is from further away in the base metal of the same sample. The hardness scales for the two hardness maps are the same showing the level of softening seen in the laser welded sample.

*Figure 7.12: Hardness map and macrograph of flash process steel after successful ballistic test (>10mm from laser weld)*
Figure 7.13: Hardness map of crack propagation through a laser welded flash process steel after ballistic test

Figure 7.14: Sectioning line for the location of the hardness map seen in Figure 7.13 with the surface shown here, being the bottom of the map
The bullet/base metal interface can be seen in Figure 7.15a where appreciable plastic flow and potential melting can be observed. It is noticeable there is a crack formation in the successful test that has been arrested within 2mm of the sample. In addition, a decreased hardness region near these crack tips can be seen in Figure 7.15b. This softened region near the crack tip to base metal levels may be due to stress relaxation provided from the crack from the surrounding hardened area. The large hardened area can be explained by the creation of dislocations from the rapid work hardening from the deformation of the high speed ballistic impact.

![Microstructure](image)

*Figure 7.15: Microstructure seen near the (a) bullet-base metal interface (b) soft region by crack*

Microstructural evaluation of the laser weld fracture that propagates through the whole thickness showed the failure occurring in the softest region of the HAZ as seen in the hardness map and optically in Figure 7.16. The material analyzed at the crack failure shows evidence of deformation that is likely induced from the shearing force of the bullet and resultant crack propagation from the ballistic test (Figure 7.17). Further study on the
fracture surface and potential controlled conditions with a Torsional Split Hopkinson Bar could reveal more information about the dynamic fracture toughness of this material.

Figure 7.16: Crack interfaces present in the softened HAZ and through the middle of the thickness of ballistic tested laser weld
The crack that goes through the centerline of the sample may be related to the centerline segregation originally seen in the flash processed steel (Figure 5.9). This is based on the location and cracking occurring along short regions throughout the middle of the thickness in accordance with the nature of these bands. In addition, banding was seen in the hardness map sample near the edge as seen in Figure 7.18. This cracking is similar to that seen in steels with high sulfur content that leads to lamellar cracking as described by Threadgill and shown in Figure 7.19 [141].

In addition to this noticeable cracking in the middle of the sample, fractography of the ballistic test hole (Figure 7.20) shows a similar anomaly where the fracture surface has a distinct ridge in the middle of the sample. Normally, ballistic failure should not have this type of feature in a petalling/ductile hole. This helps corroborate the theory put forward that banding in the microstructure is detrimental to the ballistic properties of the
steel and can lead to additional cracking in the material that was also shown by Khan et al. [103]

Figure 7.18: Banding seen near the fracture through the middle of the thickness of the ballistic tested laser weld.

Figure 7.19: Description of lamellar cracking [141] as seen in ballistic tested laser weld in Figure 7.13
Figure 7.20: Fractography of an unsuccessful ballistic test with center thickness crack

7.4. Toughness Properties

Toughness is an important mechanical property, especially in welded steels where the formation of M-A constituents and the inherent ductile-to-brittle transformation exists. The analysis of the microstructure seen in the single-pass simulations and the GMAW shows the potential formation of M-A constituents in the HAZ. Evaluation of the different peak temperatures or of as-welded conditions (laser and GMAW) could better show the weldability of flash processed steels and if laser welding has an advantage in this regard as well.
References


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