Experimental and Computational Investigation of Temper Bead Welding and Dissimilar Metal Welding for Nuclear Structures Repair

THESIS

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Abstract

Structural materials in nuclear power plants are exposed to harsh service conditions and are susceptible to environmental degradation over extended period of time. There are several weldability challenges in the component replacement and repair such as the formation of brittle martensitic microstructure in low alloy ferritic steel and the solidification/hot cracking when cladding nickel filler metal on cast stainless steels. The overarching goal of the present research is to establish the quantitative knowledge of heat transfer, molten metal flow, mixing and dilution, and microstructure evolution during arc welding process. In particular, the tempering of weld heat-affected zone (HAZ) and the weld metal dilution in dissimilar metal weld are studied.

In temper bead welding, the heat input from welding is purposefully utilized to temper the hard microstructure for improving toughness. In the present research, the effect of linear heat input on HAZ tempering was studied by a combination of experimental testing and numerical modeling. Temper bead welding experiments were performed on SA-533 Grade B Class 1 (P-No. 3) steel with filler metal 309L by cold wire Gas Tungsten Arc Welding (GTAW). Three different linear heat inputs were used while the power ratio was kept constant. The HAZ microstructure was characterized by
scanning electron microscope (SEM) and the extent of tempering in HAZ was quantified using micro-hardness mapping and thermo-mechanical simulation in Gleeble®. The extent of tempering is found to be a strong function of the peak temperature and to a much lesser extent is affected by time. 2-D and 3-D weld heat transfer models, using the double-ellipsoidal heat flux equation, were developed in Abaqus, a commercial finite element analysis code, to calculate temperature evolution. The peak temperatures experienced in the substrate’s HAZ were correlated to the hardness distribution. The results indicate that the linear heat input can have a significant influence on the extent of tempering in temper bead welding.

For the dissimilar metal weld study, high quality and repeatable experimental data was generated by a series of autogenous spot welds (i.e., no filler metal) which were made on base plates of austenitic stainless steel 304L and nickel alloy 690 using GTAW. Two heats of stainless steel 304L were used: one with low sulfur content and the other with high sulfur content. The 304L plates were butt welded together with alloy 690 plates in two shielding gases: pure argon and a mixture of argon and helium. The temperature profiles in the weldment were measured by carefully-placed surface and bottom mounted type K and type C thermocouples. Micro-hardness mapping and chemical analysis using Energy-Dispersive X-ray Spectroscopy (EDS) in SEM were used to examine the property and composition inhomogeneity in the weld metal. 2-D and 3-D weld pool models were developed based on ANSYS Fluent, a commercial computational fluid dynamics code, to understand the transport of chemical elements in the dissimilar metal weld pool. Existing constitutive equations of viscosity and surface tension in the literature were used in the
weld pool models. The molten metal fluid flow exhibits a strong outward flow driven by the Marangoni shear stress, resulting in a wide and shallow pool. The hardness distribution and chemical composition is relatively uniform in the dissimilar weld metal, indicating an extensive mixing in the molten pool. The predicted temperature by the 3-D weld pool model shows reasonable correlations with experimental data, indicating its potential in capturing the weld pool transport physics.

In summary, the present research constitutes a significant step toward an improved understanding of the two important weldability issues for the repairing of nuclear structures. First, for temper bead welding, it is shown that the extent of tempering for SA-533 HAZ is a strong function of peak temperature. Therefore, the linear heat input has an important effect on HAZ tempering and is also an important parameter that should be considered and controlled in addition to the power ratio for welding procedure development. Future work for temper bead welding includes measuring the Charpy toughness of the tempered HAZ. Second, weld metal dilution in a dissimilar metal weld of stainless steel and nickel alloy is crucial for studying solidification cracking. Experimental data has been generated that can be used to validate future molten pool physics model to understand the molten metal flow and mixing of chemical alloying elements.
For my Mom, Dad, and Heting Li
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Chapter 1: Introduction

Nuclear energy is a major resource for generating base-load electricity in the U.S. The prolonged exposure to high temperature and stress in service can result in severe component degradation especially for several important welds in the components of pressurized water reactors (PWRs). For instance, one assembly is a low alloy carbon steel SA-508 surge nozzle butt-welded to a cast CF8 stainless steel pipe using high chromium nickel filler metal (FM) 82/182 (AWS designation ERNiCrFe-3) [1]. The materials of this assembly are initially selected for preventing corrosion from the high temperature light water solution in the primary coolant loop. However after years of service, the dissimilar metal joint made of Ni-base alloy 600 and its matching filler metal 82/182 is found to be susceptible to primary water stress corrosion cracking (PWSCC). Hence advanced Ni-base alloy 690 and its matching nickel alloy FM 52M have been deployed for both repair and performance improvement of the existing components and new constructions due to their superior resistance to PWSCC. A typical structural weld overlay of alloy 52M for repairing is shown in Figure 1.1. The overlays are circumferentially welded from the low alloy carbon steel nozzle to the stainless steel pipe. In addition to increasing the wall thickness and providing resistant boundary, deposition of the overlay can also create compressive stresses in the inner diameter (ID)
of the butt weld to further prevent crack propagation [2]. However, the solution has also created two new weldability issues, namely (1) the formation of brittle martensitic microstructure in low alloy ferritic steel’s heat-affected zone (HAZ) and (2) solidification cracking when depositing alloy FM 52M on cast stainless steel substrate (even when one 308L buffer layer is used), as shown in Figure 1.1.

Figure 1.1: PWR surge nozzle weld geometry with 52M structural weld overlay [1]

The present research aims at improving the understanding of the HAZ tempering and weld metal dilution, which is essential to address the aforementioned weldability issues in nuclear component repair and replacement.

First, the tendency to form martensitic microstructure in low alloy ferritic steel welds usually requires post weld heat treatment (PWHT) to improve toughness. However, PWHT for the entire pressure vessel is not practical in field. Therefore, temper bead welding is purposefully utilized to temper the hard microstructure for improving toughness. ASME Boiler and Pressure Vessel Code Section IX defines the heat input used for temper bead welding based on the power ratio [3]. For the same power ratio, the linear heat input per unit length can vary significantly. Thus an unanswered question in the literature is whether the same extent of tempering can be attained if the linear heat input per unit length is varied while the power ratio is kept constant. In Chapter 3, temper
bead welding using different linear heat inputs but a constant power ratio is investigated. Second, a critical factor for solidification cracking in dissimilar metal weld has been attributed to the weld metal dilution, which in turn is affected by the weld pool transport phenomena including heat transfer, fluid flow, and mass diffusion. In Chapter 4, the effect of molten metal fluid flow on dilution in dissimilar metal weld is discussed. Finally, the key findings of this research are summarized in Chapter 5 and the future work is also discussed.
Chapter 2: Literature Review

2.1 Post Weld Heat Treatment

During welding, the rapid and localized heating and cooling can affect the properties of the base metal being joined. For example, many low alloy ferritic steels are produced via quench and temper heat treatment to achieve a combination of high strength and toughness. The base metal microstructure can be significantly altered as it is heated to different peak temperatures in the weldment, and can subsequently form brittle martensite upon rapid weld cooling. Hence, Post Weld Heat Treatment (PWHT) is oftentimes used to temper the brittle microstructure formed in the heat-affected zone (HAZ) of low alloy steels for improved toughness. In addition, it is used as a stress-relieving process to reduce the weld residual stresses [4].

Repair welding, especially in nuclear power generation components where the ferritic steels are heavily used, requires PWHT. There are several limitations in applying PWHT for repair welding of nuclear structural components. First, it is impractical to apply PWHT when the repair welding is done for giant pressure vessels or when there are mechanical loads carried on the structure. Second, impurity elements such as phosphorus present in the FMs can result in segregation in the weld metal during welding and subsequently embrittlement during PWHT. Finally, the PWHT process can be very expensive and time consuming for field welding repairs. Thus, as an alternative technique
for PWHT, temper bead welding (TBW) is developed and widely used in nuclear power generation industry for repair welding of ferritic steels [5].

2.2 Temper Bead Welding

TBW was initially developed for special weld components in nuclear power generation industry where heat treatment is required after welding but not feasible by traditional PWHT. The basic approach of TBW is to make a weld deposit at specific location with controlled heat input in such a way that the welding heat will affect the undesired microstructure in the HAZ or weld metal beneath that bead, thereby achieving the purpose of having a PWHT [5]. In ASME code Section IX, TBW is defined as “a weld bead placed at a specific location in or at the surface of the weld for the purpose of affecting the metallurgical properties of the heat-affected zone or previously deposited weld metal [3].”

The mechanism of temper bead welding is discussed in the following. Figure 2.1 schematically shows the microstructure formation in a single bead on plate weld of ferritic steel, where several sub-regions in the weld HAZ are correlated to the iron carbon equilibrium phase diagram. As shown in this figure, there are four distinct sub-regions in the HAZ based on different temperature profiles: Coarse Grain Heat-affected Zone (CGHAZ), fine Grain Heat-affected Zone (FGHAZ), Intercritical Heat-affected Zone (ICHAZ), and Subcritical Heat-affected Zone (SCHAZ). The chemical composition of the base metal and cooling rate are two important factors that determines the microstructure.
formed in the substrate’s HAZ. When heated above the lower transformation temperature ($A_1$), ferrite/martensite transforms into austenite forms, resulting in formation of a new microstructure. On the other hand, tempering occurs at temperatures close to but below $A_1$ where no ferrite to austenite transformation takes place [4].

![Iron-Carbon Equilibrium Diagram](image)

Figure 2.1: Various parts of HAZ as related to the iron-carbon equilibrium diagram [4]

Tempering, as shown in Figure 2.1, takes place at the SCHAZ. Figure 2.2 shows the extent of tempering of the subsequent beads to pervious beads’ HAZ. As shown in the figure, bead 2 overlaps bead 1 by about 50% and bead 3 overlaps bead 2 by about 10% of the bead width. Much of the weld metal and HAZ of bead 1 is affected when compare with bead 2, where only a small portion of the weld metal and HAZ is altered [4] [5]. Thereby, the temper bead placement is essential to the success of temper bead welding.
2.2.1 Welding heat input

Other than bead placement, careful control of the heat input is another important aspect for the success of temper bead welding. For an early study by B&W [6], the heat input used for specification of Gas Tungsten Arc Welding (GTAW) was defined using the conventional heat input shown below:

\[ H_{in} = \frac{I \times V}{TS} \]  

(2.1)

where \( H_{in} \) is the linear heat input per unit length (J/mm), \( I \) is the welding current (A), \( V \) is the voltage (Volt), and \( TS \) is the travel speed (mm/s).

2.2.2 Welding power ratio

It is noted that Equation (2.1) does not consider the filler metal addition during welding. Therefore, another formula for calculating heat input including the filler metal
was introduced in a study by EPRI Repair & Replacement Applications Center (RRAC) [7]. That heat input is referred to as “Power Ratio” as shown below:

\[
P_R = \frac{I \times V}{WFS \times A_{FW}}
\]  

(2.2)

where \(P_R\) is the power ratio (W/mm\(^2\) or J/s-mm\(^2\)), \(WFS\) is the wire feed speed (mm/s), and \(A_{FW}\) is the cross-sectional area of the filler wire (mm\(^2\)). One advantage of using the power ratio is that it can be directly correlated to the weld metal dilution. For instance, Smartt and Key showed that if the linear heat input per unit length was held constant, the weld metal dilution was reduced as the power ratio was decreased [8]. It is also noted that both the linear heat input defined in Equation (2.1) and the power ratio in Equation (2.2) are “gross” heat input and do not include the arc efficiency factor whose value is about 80% for GTAW.

ASME Boiler and Pressure Vessel Code Section IX defines the heat input used for temper bead welding based on the power ratio [3]. For the same power ratio, the linear heat input per unit length can vary significantly. This can be illustrated by combing Equation (2.1) and (2.2), as shown below:

\[
P_R = \frac{(TS)^2}{WFS \times A_{FW}} H_{in}
\]  

(2.3)

In Equation (2.3), the travel speed and wire feed speed can vary independently of each other. Therefore, there is no direct dependence of \(H_{in}\) on \(P_R\). For instance, the power ratio can be kept constant while the linear heat input is reduced to \(1/4\) of its original value by simply doubling \(TS\), reducing \(WFS\) to \(1/4\) of its original value, or a combination of changes in both \(TS\) and \(WFS\).
2.3 Heat-affected Zone Tempering Kinetics of Low Alloy Steels

When welding onto low alloy steels, the weld metal and the HAZ can form predominately martensitic microstructure due to rapid cooling. Such microstructure exhibits very high hardness but poor fracture toughness and low ductility [9]. As it is a thermally unstable phase, martensite can decompose into ferrite and cementite when heated close to the critical transformation temperature ($A_{c1}$). Such tempering heat treatment is commonly utilized to decrease the hardness and improve the fracture toughness. Published studies of isothermal tempering show that the tempering process of martensite proceeds in a series of overlapped stages. In general, the tempering of martensite starts with diffusion of excess carbon to precipitate carbides resulting in a dispersion of carbides in a ferritic matrix. The tempering kinetics, controlled mainly by carbon diffusion, depends on both temperature and time. At the elevated temperatures over extended period, cementite starts to coarsen and forms spheroids above 523 K (250 °C), and martensite begins to recrystallize above 873 K (600 °C) [10]. Tempering in welding is highly non-isothermal. The extent of tempering is significantly affected by the local peak temperature due to the short time resulting from the rapid heating and cooling during welding.

For low alloy steels, elements such as Mn, Cr, Mo, etc. are added to increase the steels’ hardenability, and these types of steel often require PWHT [9]. However, most of the added elements are strong carbide formers, and past studies have shown that with additions of these elements, the softening reaction during tempering is delayed by retarding either cementite precipitation or growth within the tempered martensite.
Because the carbides formed by those alloying elements are thermodynamically more stable than cementite, and when there is sufficient time, these carbides will be formed in preference to cementite [11]. It is noted that alloy carbides will not form until the temperature is in the range of 500-600 °C or above and the formation of alloy carbides are often accompanied with an increase in hardness [12]. This phenomenon is called secondary hardening and it happens most likely in steels containing Mo, V, W, Ti, and Cr at higher alloy content. An example of secondary hardening effect of increasing Mo content in a series of steels containing 0.1 wt% carbon is shown in Figure 2.3.

![Image of Figure 2.3]  

**Figure 2.3:** The secondary hardening effect of Mo during the tempering of quenched steels [13]
2.3.1 Prediction of martensite decomposition

For some special steel welds such as welding of the dual-phase steels in automobile industry, the material’s mechanical properties will degrade when the martensite is tempered or fully decomposed at weld’s subcritical HAZ. Several studies have then been done to characterize this HAZ softening phenomenon [13]. Xia et al. had shown that the measured softening inside the steel’s HAZ was a sigmoidal function of welding heat input and the softening was also found to be a linear function of martensite fraction after welding [14]. Baltazar Hernandez et al. showed that the softening behavior was primarily contributed to martensite decomposition and ferrite phase itself did not experience a significant reduction on hardness [15]. The evidences of these studies implied that the HAZ softening could be solely due to martensite decomposition. Based on other literatures, the martensite is only partially decomposed when tempering occurs in a short time, and the decomposition will continue as the tempering temperature and time are increased [16]. Thus, this behavior of transformation is well described by the Johnson-Mehl-Avrami-Kolmogorov equation [11], as shown below:

$$\phi = 1 - \exp(-kt^n)$$  \hspace{1cm} (2.4)

where $\phi$ is the volume fraction of martensite that has been transformed, $k$ is a material parameter representing the energy barrier for the tempering transformation (and is a strong function of tempering temperature), $t$ is the tempering time, and $n$ is another parameter representing the rate of the transformation. By applying a simple normalizing relationship to the hardness data, the volume fraction of martensite can be determined by the following equation [16]:

---

11
\[ \phi = \frac{H_{BM} - H}{H_{BM} - H_\infty} \quad (2.5) \]

where \( H_{BM} \) is the as-received base metal (martensitic) hardness, \( H \) is the measured hardness of the tempered sample, and \( H_\infty \) is the minimum alloy hardness. Therefore, there is no tempering when \( \phi \) is 0, and martensite has fully decomposed when \( \phi \) is 1.

### 2.4 Dissimilar Metal Welding

Dissimilar metal fusion welding in general can create five different regions in the joint [17], as shown in Figure 2.4. The composite zone is the region where both liquid of filler metal and base metal are mixed, and this region has relatively uniform mixing and composition. The unmixed zone is the region where the base metal is melted but has not yet mixed with the filler metal, and this zone is usually very narrow and hard to distinguish. The weld interface (transition zone) is the region where the base metal composition is gradually changed to filler metal composition in a short distance, and within this region, all microstructural constituents can exist. The partially melted zone is the region in which the based metal has temperature spanning between its solidus and liquidus temperatures, and thus the base metal is not fully melted. Finally, the heat-affected zone is the region that is affected by welding temperature below base metal’s solidus temperature [17] [18].
Dissimilar metal welding of nickel-base FMs onto both austenitic stainless steels and low alloy steels is extensively applied on components repair in nuclear power industry. Nickel-base FMs usually have high percentages of chromium, iron, molybdenum and niobium. The Ni-Cr-Fe families of alloys used in repairing of nuclear components are solid-solution strengthened alloys which exhibit excellent corrosion resistance and moderate strength after welding [19]. To realize the full benefits of nickel-base FMs, their weldability issues need to be addressed. Those issues include solidification cracking and ductility-dip cracking (DDC). Particularly, nickel alloy 52M structural weld overlays (shown in Figure 1.1) have been identified strongly susceptible to solidification cracks. Weld metal dilution has been shown to be a critical factor affecting solidification cracking.

### 2.4.1 Solidification cracking

Solidification cracking is also known as hot cracking and generally associated with brittle failures, as shown in Figure 2.5. It usually happens when the last liquid to
solidify is not strong enough to counteract the tensile stresses formed by thermal contraction and solidification shrinkage of the weld metal upon cooling, and this effect is especially pronounced when the last liquid to solidify exists along the grain boundary of two different grains [20]. Solidification cracking is a complex phenomenon, and theories to explain the mechanisms of solidification cracking are still evolving.

Figure 2.5: Solidification cracking in the composition transition zone between the filler metal 52M overlay and the 308L buffer layer [20]

The solidification temperature range (STR) is an important indicator for solidification cracking susceptibility. In general, as the STR of weld metal increases, the susceptibility of solidification cracking increases. Impurity elements especially sulfur, phosphorous, boron, and silicon can play an important role in increasing STR due to the reason that these elements segregate to the liquid upon cooling and thus form low melting eutectic constituents which will depress the solidus temperature [17] [20]. A typical diagram illustrates both the effect of impurity elements and STR to susceptibility of
solidification cracking is shown in Figure 2.6. In some alloys, the amount of low melting eutectic liquid can be controlled by changing the content of the alloying element such as Nb. Crack healing can occur when a significant amount of eutectic liquid forms at the end of solidification stage and backfills the crack [17].

![Diagram of solidification process](image)

Figure 2.6: Effect of impurity elements and STR to solidification cracking susceptibility:
(a) macro of weld pool (b) pure iron with no STR (c) iron with silicon and (d) iron with sulfur [21]

2.4.2 Weld metal dilution

A large body of research concludes that the primary factor in solidification cracking of repair weld overlays is contributed to excess weld metal dilution of the nickel-base FM 52M. Dilution is used to describe the average degree of mixing between the base metal and the filler metal, and is calculated using the following equation [21]:

\[
\text{Dilution (\%)} = 100 \times \frac{B}{A+B}
\]  

(2.6)
where $A$ is the cross sectional area of the filler metal deposited and $B$ is the cross sectional area of the melted base metal, which is shown in Figure 2.7.

![Figure 2.7: Areas of dilution calculated in Equation (2.6) [22]](image)

McCracken and Smith found that the dilution of 52M overlays on 304L correlated well with power ratio which defined in Equation (2.2). They also found that by controlling power ratio below $0.186 \text{ kW/mm}^2$ (120 kW/in$^2$), the solidification cracking can be avoided [22]. The relationship is plotted in Figure 2.8. Therefore, control of weld metal dilution is a key factor to overcome solidification cracking in dissimilar metal welding of nickel-base FM to stainless steel substrate. An understanding of molten pool fluid flow is essential to study the weld metal dilution.
2.5 Weld Pool Fluid Flow

The weld pool fluid flow behavior is mainly governed by four driving forces including buoyancy force, the electromagnetic force (Lorentz force), the surface tension gradient (Marangoni shear stress) at the weld pool surface, and the impinging force (shear stress) of the arc plasma [23], as shown in Figure 2.9.

Figure 2.8: Relationship of dilution and GTAW power ratio of filler metal 52M and base metal stainless steel 304L [23]
Figure 2.9: Driving forces for weld pool fluid flow behavior: (a, b) buoyancy force; (c, d) electromagnetic force; (e, f) shear stress caused by surface tension gradient; (g, h) impinging force caused by arc plasma [21]

The buoyancy force, as shown in Figure 2.9 (a), represents the sink of cooler liquid metal at point b where the temperature is the lowest at the melting point, and the raise of the hotter liquid metal at point a where the heat source is more concentrated above. This is because the density of liquid metal decreases with increasing temperature. The electromagnetic force, also called the Lorentz force, is induced by combining the
converging current field and the magnetic field, as shown in Figure 2.9 (c). This force results in an inward flow by pushing the liquid metal downward at the weld pool center. For pure metals without the surface-active elements, the liquid metal surface tension ($\gamma$) decreases with increasing temperature. Thus the hotter liquid metal that has a lower surface tension at point a will be pulled out by the cooler liquid metal at point b which has a higher surface tension. This effect will result in an outward flow of the liquid metal from the center of the weld pool to the edge and returning below the pool surface as shown in Figure 2.9 (e). When the material contains a small amount of surface-active elements (e.g., S), the surface tension can increase with temperature. As a result, the direction of the flow is reversed. Finally, the impinging force by the arc plasma, as shown in Figure 2.9 (g), results in an outward flow of the liquid metal from the center to the edge of the weld pool in the same direction of the outward flowing arc plasma [20]. Numerous numerical models have been developed for weld pool fluid flow incorporating the above four individual driving forces [24].

2.5.1 Marangoni convection

Since a liquid with a high surface tension pulls more strongly on the surrounding liquid than one with a low surface tension, the presence of a gradient in surface tension will naturally cause the liquid to flow from regions of low surface tension to high surface tension. Such surface tension driven convection, called Marangoni convection, is commonly induced by temperature gradients tangential to the interface [25]. Surface-active elements such as O and S can have a marked effect on the surface tension even
though they are only present in parts per million (ppm) levels. Particularly, the temperature coefficient of surface tension ($\partial \gamma/\partial T$) can be changed from negative to positive with addition of surface-active elements. For instance, pure liquid iron has a negative surface tension coefficient at all temperatures. However, with addition of S, $\partial \gamma/\partial T$ can be changed to positive [20]. Based on Heiple’s model shown in Figure 2.10, with low sulfur content in the weld pool, the hotter liquid metal (thus with lower surface tension) in the center of the weld pool is pulled out by the cooler liquid metal (with higher surface tension) in the weld pool edge due to negative surface tension coefficient. This results in an outward flow of the liquid metal and subsequently a wide but shallow weld pool shape. With high sulfur content in the weld pool, the cooler liquid metal (with lower surface tension) at the weld pool edge is pulled in by the hotter liquid metal (with higher surface tension) at the center of the weld pool due to positive surface tension coefficient. This produces an inward flow of the liquid metal and subsequently a narrow but deep weld pool shape [20] [26].
Some experiments done by Limmaneevichitr and Kou also proved Heiple’s model. As shown in Figure 2.11, the stainless steel welds with higher sulfur content exhibited inward flow resulting in a relatively narrower but deeper weld pool when compared to the welds with lower sulfur content. Such distinct weld pool shapes can be explained using the surface tension data shown in Figure 2.12, which shows that the low sulfur stainless steel has a negative $\frac{\partial \gamma}{\partial T}$ while the high sulfur stainless steel has a positive $\frac{\partial \gamma}{\partial T}$ up to 2100 °C [26].
Figure 2.11: YAG laser welds in 304 stainless steels with (a) 40 ppm sulfur and (b) 140 ppm sulfur [21]

Figure 2.12: Surface tension data of two different heats of stainless steel 316. High sulfur has 160 ppm more sulfur than the low sulfur [21]

2.5.2 Laminar and turbulence flow

Laminar flow by definition is a type of fluid flow in which the fluid travels smoothly or in regular paths. Laminar flow occurs in cases where the fluid is moving slowly and its viscosity is relatively high. On the other hand, turbulent flow is a type of fluid flow in which the fluid undergoes irregular fluctuations. In other words, the velocity of the fluid at a point is continuously undergoing changes in both magnitude and
direction in turbulent flow [27]. In modeling of weld pool fluid flow, both laminar and turbulent flow modes have been used. Although the weld pools are often visualized to have unsteady and chaotic behaviors, it is still not completely clear whether weld pool flow is turbulent or laminar [28]. For a simple consideration of the effect of turbulence, many flat surface laminar models were reported to use adjustable factors for the thermal conductivity and viscosity to enhance the momentum and the energy transfers. This method does not directly consider the turbulence physics. In some cases, the weld pool depths are over-predicted than the measured one from actual welds. Weld pool modeling based on the turbulence is also developed [28]. Choo and Szekely used Reynolds-Averaged Navier-Stokes turbulence model in their simulation. They suggested that weld pool flows were more likely to be turbulent on the basis that the numerical simulations with turbulent models provided a better agreement of weld pool shapes with experimental results [29]. Hong et al. employed $k-\varepsilon$ turbulence model in the simulation of GTA welding of stainless steel 304. Their simulation results also showed a good agreement of weld pool depth, whereas the depth obtained using a laminar model was found to over-predict experimental results [20] [30]. The simulation results of both laminar and turbulent models are shown in Figure 2.13.
Figure 2.13: Comparison of predicted and measured weld pool shape: (a) laminar flow and (b) turbulent flow from Hong et al. [30]

The literature review shows a large body of past studies addressing the two weldability issues (i.e., temper bead welding and solidification cracking). However, the following technical gaps still exist. First, the tempering kinetics for pressure vessel steels such as SA-533 Grade B Class 1 is not well established for heating rates and peak temperatures relevant to welding. Particularly, the quantitative data of extent of tempering as a function of temperature and time is limited. Second, for solidification cracking, there is a lack of quantitative understanding of weld pool transport physics (i.e., heat transfer, fluid flow and diffusion) that are essential to study and control weld metal dilution. Although transport phenomena based models have been successfully used for studying weld pool physics in various alloy systems, there have been limited applications of such models to dissimilar metal welds between nickel alloy and stainless steel.
Addressing these two technical gaps, the present research aims at improving the understanding of the HAZ tempering and weld metal dilution.
Chapter 3: Temper Bead Welding

3.1 Materials

The steel used for temper bead welding was a low alloy carbon steel SA-533 Grade (Gr.) B Class (Cl.) 1, an ASME P-No.3 pressure vessel steel. The as-received block was sectioned into several flat plates with dimensions of 210 mm long, 160 mm wide and 19.4 mm thick. The filler metal used was stainless steel ER 309L with a diameter of 0.89 mm. Chemical compositions of the two materials are listed in Table 3.1.

Table 3.1: Chemical composition of SA-533 base metal and ER 309L filler metal

<table>
<thead>
<tr>
<th>Element</th>
<th>SA-533</th>
<th>ER 309L</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>0.19</td>
<td>0.017</td>
</tr>
<tr>
<td>Mn</td>
<td>1.27</td>
<td>2.06</td>
</tr>
<tr>
<td>Ni</td>
<td>0.65</td>
<td>13.6</td>
</tr>
<tr>
<td>Cr</td>
<td>0</td>
<td>23.3</td>
</tr>
<tr>
<td>Mo</td>
<td>0.57</td>
<td>0.07</td>
</tr>
<tr>
<td>Cu</td>
<td>0.14</td>
<td>0.04</td>
</tr>
<tr>
<td>Si</td>
<td>0.19</td>
<td>0.47</td>
</tr>
<tr>
<td>P</td>
<td>0.009</td>
<td>0.016</td>
</tr>
<tr>
<td>S</td>
<td>0.009</td>
<td>0.001</td>
</tr>
<tr>
<td>Co</td>
<td>0.016</td>
<td>0.031</td>
</tr>
<tr>
<td>Al</td>
<td>0.023</td>
<td>0</td>
</tr>
<tr>
<td>Ti</td>
<td>0</td>
<td>0.004</td>
</tr>
<tr>
<td>V</td>
<td>0</td>
<td>0.063</td>
</tr>
<tr>
<td>Fe</td>
<td>Bal.</td>
<td>Bal.</td>
</tr>
</tbody>
</table>
3.2 Design of Experiment

The welding experiments were divided into two groups: (1) single bead on plate welds, and (2) multiple bead cladding on plates. The first group, i.e., single bead on plate welds, was used to quantify the steel hardenability of the weld heat-affected zone (HAZ). Each of the individual beads was designed with a welding length of 160 mm long, and the spacing between passes was 50 mm wide. In the second group of welding experiments, multiple bead cladding on SA-533 steel plate was deposited using ER 309L filler metal to evaluate the extent of tempering in the substrate’s HAZ. The designed bead placement was schematically shown in Figure 3.1. There were three layers of beads for a clad. The very first bead was positioned at a distance of about 12 mm to the left of the plate centerline. For the remaining passes in the first layer, the tungsten electrode was first lined up at the right edge of the previous bead and then was offset by 1.02 to 1.27 mm (0.040 to 0.050 in) toward the right side to deposit the next bead. For depositing the first bead of the second and third layers, the tungsten electrode was positioned at the bottom of the valley between the first and second beads made at the layer below. The remaining beads in those two layers were then deposited following the same procedure as that for the first layer. Each cladding bead was designed with a welding length of 150 mm long. No pre-heating was used for all the cladding processes, and the plate was allowed to cool down to the room temperature before the next weld bead was deposited.
3.3 Experimental Procedures

3.3.1 Welding processes

Cold wire Gas Tungsten Arc Welding (GTAW) process was utilized for the temper bead welding experiments. All the cladding beads were made by Jetline series 9500 welding control system with the power source Miller Dynasty 380 TIG welder shown in Figure 3.2. The shielding gas used was pure argon with a gas flow rate of 35 Cubic Feet per Hour (CFH), and the electrode that used for producing all the welds was 2% ceriated tungsten electrode with 20 degree tip angle, 0.508 mm (0.02 in) tip blunt, and 3.175 mm (0.125 in) diameter. Three sets of welding parameters were evaluated, as summarized in Table 3.2. These parameters yielded three levels of linear heat input: low (685 J/mm), medium (1060 J/mm), and high (1432 J/mm) for the same power ratio about 215 W/mm².
Figure 3.2: Welding control system and power source

Table 3.2: Welding parameter used for temper bead welding experiments

<table>
<thead>
<tr>
<th>Welding parameters</th>
<th>Low (L)</th>
<th>Medium (M)</th>
<th>High (H)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Voltage (V)</td>
<td>9.5</td>
<td>10.5</td>
<td>10.5</td>
</tr>
<tr>
<td>Current (A)</td>
<td>145</td>
<td>171</td>
<td>231</td>
</tr>
<tr>
<td>Wire feed speed (mm/s)</td>
<td>18.4</td>
<td>22.9</td>
<td>30.9</td>
</tr>
<tr>
<td>Wire diameter (mm)</td>
<td></td>
<td>0.89</td>
<td></td>
</tr>
<tr>
<td>Travel speed (mm/s)</td>
<td>1.91</td>
<td>1.69</td>
<td>169</td>
</tr>
<tr>
<td>Heat input (J/mm)</td>
<td>685</td>
<td>1060</td>
<td>1432</td>
</tr>
<tr>
<td>Power ratio (W/mm²)</td>
<td>217.5</td>
<td>214.3</td>
<td>214.1</td>
</tr>
</tbody>
</table>

For single bead on plate welds, each individual bead was deposited using the three levels of linear heat input in Table 3.2. For multiple bead welds, three sets of cladding experiments were performed to evaluate the effect of linear heat input on tempering of the substrates’ HAZ. The first set, referred to as “M-M-M”, utilized the medium heat input (1060 J/mm) for all three layers. The second set, referred to as “H-L-L”, applied the high heat input (1432 J/mm) for the first layer, and then the low heat input (685 J/mm)
for both the second and third layers. The last set, referred to as “H-H-H”, used the high heat input (1432 J/mm) for all three layers.

3.3.2 Thermal cycle measurements

To track the thermal history during welding, type-K thermocouples (TCs) were placed to every plate. For single bead on plate welds, three thermocouples were attached onto the plate surface in between passes, as shown in Figure 3.3. In order to measure higher peak temperatures in the substrate’s HAZ in the multiple bead cladding experiments, small blind holes were drilled from the backside of the plate to about 2 mm underneath the top surface. Type-C TCs were spot welded to the bottom of the drilled holes and the holes were then sealed with ceramic paste to secure the TC attachment. The placement of TCs for M-M-M and H-L-L is schematically illustrated in Figure 3.4 (a) where, one pair of TCs was positioned below the very first bead, and the other pair was below the centerline of the clad. Figure 3.4 (b) shows the thermocouple positions on the back side of the plate for those two cladding experiments. For H-H-H cladding experiment, three thermocouples were placed from the back side of the plate; all three were located along the centerline of the clad: one in the front, another on the middle, and the third in the back, as shown in Figure 3.5.
Figure 3.3: Thermocouple positions for single bead on plate welds with (1) low (2) medium, and (3) high heat inputs form left to right.

Figure 3.4: (a) Schematics of thermocouple positions and (b) back side of the plate showing sealed thermocouples for M-M-M and H-L-L cladding experiments.

Figure 3.5: Thermocouple positions for H-H-H cladding experiment.
3.3.3 **Metallurgical characterization**

The welds for both single bead on plate and multi-bead clad had cross-sections extracted and mounted in 2 inch Bakelite molds in a Leco PR-36 mounting press. Samples were polished through the steps of 240, 320, 400, 600, and 800 grit silicon carbide sand papers. Water was used as lubricant for all steps except for the 800 grit where Leco Diamond Extender compound was used to prevent pitting. Polishing followed a progress of 9 µm, 3 µm, and 1 µm diamond paste on Leco LeCloth with Leco Diamond Extender compound used as lubricant. Etching of the welds entailed a two-step process of swab-etching with 2% Nital to reveal the low alloy carbon steel’s microstructure, followed by 10% chromic acid electrolytic etching at 5 volts to reveal the stainless steel filler metal passes. Light optical micrographs were taken on an Olympus GX51 with up to 1000X magnification. Scanning electron microscopy was conducted in a FEI Quanta 200 scanning electron microscope to examine the formation of tempered and untempered martensite in welds’ HAZ.

3.3.4 **Hardness measurement**

Heat-affected zone hardness for single bead on plate welds and multi-bead clads were conducted by Leco LM 100AT micro-hardness tester. Line hardness measurements were done through width and depth directions of single bead on plate welds. Hardness maps with micro-hardness indents of load 300 g or 500 g and spacing from 150 µm or 250 µm were taken for multi-bead clads. The hardness values in Vickers (HVN) were plotted in Excel.
3.3.5 Gleeble thermal simulation

As the HAZ experiences high temperature gradient during welding, it is difficult to correlate the time and temperature to the extent of tempering using the hardness measured in the actual weld. Hence, to establish the HAZ tempering kinetics for SA-533 steel, controlled tempering experiments were performed in Gleeble 3800 a thermo-mechanical physical simulator. The Gleeble, as shown in Figure 3.6, was equipped with low force jaws and water quench system to prevent inducing stress and to rapid cool samples, respectively. All the Gleeble samples were machined from the as-received SA-533 plates, and were flat bars with dimensions of 110 mm long, 10 mm wide, and 4 mm thick. The view of the sample can be seen in Figure 3.7.

Figure 3.6: Gleeble 3800 thermo-mechanical simulator with water quench system
Each sample was welded with a type K thermocouple in the middle, which was used both to control power from the Gleeble for achieving the specified temperature profile and to record the actual sample temperature. Several samples were prepared for different tempering conditions. The tempering temperatures tested were 700, 600, 500, and 400 °C, and the tempering times used were 0.5, 1, 5, 60, and 600 s. Prior to tempering, each sample was heated to 1300 °C with a heating rate of 450 °C/s, held at 1300 °C for 2 s, and followed by water quenching to room temperature. This process was used to form fully martensite for subsequent tempering. The Gleeble thermal cycle used for this “pre-processing” was shown in Figure 3.8.
3.4 Modeling of Temper Bead Welding

3.4.1 Modeling of thermal cycle

To a first approximation, a 2-D heat transfer model was developed based on Abaqus® to calculate the temperature evolution on a weld transverse cross section. The weld bead profiles were extracted as a series of XY coordinates from the actual weld macrographs for claddings M-M-M and H-L-L. Those coordinates were then imported into Abaqus CAE, a FEA pre-processor, to create the geometry and then the mesh for the two claddings.

The 2-D model only considers the heat conduction on the weld transverse section. To improve the accuracy of thermal modeling, a preliminary 3-D heat transfer model was
developed for the cladding H-H-H based on the same procedures as the 2-D model for M-M-M and H-L-L.

It is not readily available the temperature-dependent thermo-physical property data for SA-533 steel especially at high temperatures. The data for A-516 Gr. 70 steel was thus used due to its similar composition and properties [33]. There was some weld metal dilution especially at higher heat inputs, but for simplicity, the effect of weld metal dilution was not considered and the thermo-physical property data for ER 309L [33] was used for the weld metal.

The heat input from the arc was modeled using the moving heat source approach. It was based on the widely used double-ellipsoidal heat flux equation shown below [34]:

\[
q(x, y, z) = \frac{6\sqrt{3} Q}{abc \pi^{3/2}} e^{-\frac{x^2}{a^2}} e^{-\frac{y^2}{b^2}} e^{-\frac{z^2}{c^2}}
\]

where \( q \) is the volumetric heat flux at a location (i.e., a mesh node), and \( Q \) is the total arc heat. In the model, the center of the heat source is positioned at the center of a given weld bead. \( x, y, z \) are distances to the heat source center along the transverse, thickness and longitudinal (or traveling) directions, respectively. \( a, b, c \) are heat source parameters, which values are taken as weld bead half width, weld bead depth/height, and weld pool front/rear length, respectively [34].

3.4.2 Modeling of heat-affected zone hardness

The HAZ hardness prediction is based on the results of both Gleeble simulation and calculated thermal history that are introduced previously. Particularly, the hardness for each Gleeble simulated sample is measured, and the hardness values at each
tempering time and temperature are then used in Equation (2.5) to calculate the extent of tempering (φ). For simplicity, φ is assumed to be dependent only on temperature and not on time. This simplification is valid since φ exhibits a much stronger dependence on temperature than time as discussed in details later. Taking advantage of this simplification, φ value at 0.5 s as a function of tempering temperature is plotted and a correlation is then developed utilizing the least square root fitting.

Using the correlation between the extent of tempering (at 0.5 s) and the tempering temperature, the final hardness in the substrate’s HAZ can be assessed. This is done by first extract thermal cycles at selected monitoring locations in the CGHAZ and FGHAZ from the weld heat transfer models. The peak temperature experienced due to deposition of subsequent, adjacent beads at those locations are then inputted into the developed correlation to calculate φ, which in turn is used to calculate the hardness. The calculated hardness is compared to that from the hardness map measured on the actual weld.

3.5 Experimental Results and Discussion

3.5.1 Thermal cycle measurements

Figure 3.9 (a) and (b) showed the thermal cycles measured by thermocouples for the first and second layers of clad M-M-M. This pair of thermocouples was located to the left of the entire clad below the center of the very first pass. Post-weld sectioning revealed that the thermocouples were placed at 4.5 mm beneath the top surface, short of the 2 mm depth planned. As shown in Figure 3.9 (a), for the first layer, the peak temperature was highest (close to 1000 K) for the first pass and then decreased for the
subsequent passes in this layer, which were located further away from the thermocouples. As shown in Figure 3.9 (b), the peak temperature measured by the thermocouples for the first pass of the 2\textsuperscript{nd} layer was only about 200 K lower than that for the first pass of the 1\textsuperscript{st} layer. This indicated the heat from depositing the second layer is significantly diffused into the substrate’s HAZ beneath the first layer. The peak temperature for the third layer dropped further down to 600 K. Hence, the tempering on the substrate’s HAZ resulted mostly from the deposition of the second layer.

![Figure 3.9: Welding thermal cycles for clad M-M-M (a) 1\textsuperscript{st} layer (b) 2\textsuperscript{nd} layer](image)

For clad H-L-L, the thermal cycles measured by another pair of thermocouples for the first and second layers are shown in Figure 3.10 (a) and (b). This pair of thermocouples was positioned near the back of the clad (close to the second pass of the 1\textsuperscript{st} layer) and at a depth of 2 mm below the top surface. Due to the high heat input used for depositing the first layer and the close proximity of thermocouples to the surface, the peak temperature measured at the thermocouples location was quite high around 1580 K. However, the peak temperature experienced at this location dropped drastically to only 850 K when depositing the second layer.
Finally, the measured thermal cycles for Clad H-H-H are plotted in Figure 3.11 (a) and (b) for depositing the 1\textsuperscript{st} and 2\textsuperscript{nd} layers, respectively. This pair of thermocouples was located at the center of the entire clad (in-between the third and fourth passes of the 1\textsuperscript{st} layer) and also at a depth of 2 mm below the top surface. Similar to clad H-L-L, the peak temperature for the 1\textsuperscript{st} layer was also quite high around 1600 K. But unlike H-L-L, the peak temperature dropped to 1150 K for the 2\textsuperscript{nd} layer. This is expected as the 2\textsuperscript{nd} layer beads in H-H-H were deposited with high heat input while those in H-L-L were deposited with low heat input.
The percentage of decrease in peak temperature measured by the TCs from the first layer to the second layer was 20% for clad M-M-M, 46% for H-L-L, and 28% for H-H-H. It is noted that the TCs were placed at about 2 mm beneath the surface for H-H-H and H-L-L but further away from the surface (4.5 mm) for M-M-M. A comparison of thermal cycles in H-H-H and H-L-L shows there is a much higher drop in the peak temperature in H-L-L. It occurred because it was more difficult for the heat from the second layer with low heat input to diffuse through the large weld beads in the first layer deposited using the high heat input. As discussed later, the higher drop in the peak temperature correlates to the weaker tempering for clad H-L-L.

3.5.2 Metallurgical characterization

The weld macrographs of the three single bead on plate welds, deposited using low, medium and high heat inputs, are shown in Figures 3.12 (a), (b), and (c), respectively. The power ratio used was the same for those three welds. As the linear heat input rises, both the weld bead depth and width increased. An important factor for such
behavior was the welding current. The Lorentz force, an essential driving force for molten pool convection, was proportional to the square of current [22]. For instance, as the current rises by 160% from low (145 A) to high (231 A) heat input, the Lorentz force can increase by 254%. The higher Lorentz force results in a stronger downward “digging” flow at the pool center and a more intense molten pool convection; both in turn contribute to the deeper and wider bead profile as the heat input becomes higher.

![Single bead profiles with different heat input](image)

Figure 3.12: Single bead profiles with different heat input (a) low, (b) medium and (c) high

A consequence of the increased weld depth and width is the rise in weld metal dilution, which is defined as the ratio of the melted area on the substrate to the total area of the weld bead, as shown in Equation (2.6). The weld metal dilution as a function of heat input is plotted in Figure 3.13. A jump in dilution was observed when the heat input increased from 685 to 1060 J/mm. As the heat input further rised beyond 1060 J/mm, the increase in dilution was tapered off.
The single bead on plate welds’ HAZ was characterized to examine the hardenability of SA-533 steel during welding. Particularly, welds made by medium and high heat inputs were selected for characterization of the HAZ microstructure in the scanning electron microscope (SEM). SEM images were taken for each weld’s coarse grain heat-affected zone (CGHAZ), fine grain heat-affected zone (FGHAZ) and base metal (BM). These microstructures are shown in Figure 3.14. The distinct lath-shaped gain structures oriented with the prior austenite grain boundary in CGHAZ suggest a fully martensitic microstructure. The FGHAZ microstructure is somewhat finer and also appears to be martensite. The weld made by the low heat input was not examined. Since it had the fasted cooling rate (due to the lowest heat input), it is expected to also form a fully martensitic microstructure in the CGHAZ and FGHAZ.

Figure 3.13: Weld metal dilution as a function of linear heat input
To understand the formation of martensite during welding, the Continuous Cooling Transformation (CCT) diagram is examined, as shown in Figure 3.15. In this diagram, the critical (slowest) cooling rate (avoiding the nose of bainitic transformation) to form fully martensitic structure is 25 °C/s (77 °F/s); this curve appears as line b in the diagram. The typical cooling rate experienced in HAZ at the high heat input was estimated to be about 75 °C/s based on the temperature profiles recorded by the thermocouple shown in Figures 3.10 and 3.11. As the weld cooling rate is much higher than the critical cooling rate required for forming martensite, the HAZ microstructure of the as-welded beads made by all three heat inputs is expected to be fully martensitic.
Figure 3.15: Measured CCT diagram of SA-533 steel

The weld macrographs of the three layer, multiple bead claddings, deposited using combinations of low, medium and high heat inputs, are shown in Figure 3.16 (a), (b) and (c), respectively. As described previously, the power ratio used for all three claddings was the same. However, due to different linear heat inputs, the beads’ profiles as well as the HAZ sizes appeared quite different.
Figure 3.16: Cross-sectional images of multiple layer temper beads: (a) clad M-M-M, (b) clad H-L-L, and (c) clad H-H-H

Tempering of HAZ in the base metal occurred from heating both by neighboring beads in the first layer and by the upper beads in the second and to a much lesser extent the third layer. For the first bead at the first layer, as discussed earlier in this section, martensite formed around the HAZ of the bead after welding. When the second bead deposited next to the first bead, in transverse direction, the HAZ that was created by the first bead was changed into re-martensitized, tempered, and untempered regions shown in Figure 3.16 (b) by tracing the HAZ edges. The re-martensitized region experienced temperature that was re-heated above Ac1 by the second bead, and upon cooling,
martensite formed again. The tempered region experienced temperature that was below Ac1 but still high enough to decompose martensite, so tempering occurred at this region. For the untempered region, the temperature due to the second bead was not high enough to fully decompose martensite. Therefore, there was consisted of mostly untempered martensite in this region.

The important role of the second layer played in tempering first layer’s HAZ can be illustrated using the temperature profiles. Except for the rightmost and leftmost weld toes, the re-martensitized and untempered regions were tempered primarily due to depositing the second layer. This is due to the relatively high peak temperature (as recorded by the thermocouples) when depositing the second layer (see Figures 3.10 and 3.11). It is noted that the extent of tempering depended strongly on linear heat inputs used to deposit beads of the second layer. In Figure 3.16 (a), part of the HAZ for beads applied by medium heat input at the second layer appeared to be “superimposed” at the first layer’s HAZ. In Figure 3.16 (c), due to high heat input used to deposit beads of the second and third layers, part of the HAZ for beads even at the third layer appeared at the first layer’s HAZ. However in Figure 3.16 (b), the beads made by low heat input at the second layer did not seem to have strong temperature profiles that could affect the first layer’s HAZ.

To more clearly examine the effect of the second layer on tempering, a transverse macrograph of two layers is plotted, as shown in Figure 3.17. This section was produced by staggering the cladding length. In other words, when depositing the 3rd layer, it did not cover the full length of the 2nd layer, thus leaving a section of the 2nd layer not covered by the 3rd layer. As shown in Figure 3.17 (a), the beads at the second layer made by low heat
input did not show significant tempering effect on the first layer’s HAZ. However as shown in Figure 3.17 (b), the beads at the second layer made by high heat input appeared to superimpose part of their HAZ in the first layer’s HAZ.

To further examine the extent of tempering in H-H versus H-L, SEM images at two positions located at approximately the same position to the fusion boundary of the first layer were taken. These two positions are labeled as points 1 and 2 in Figure 3.17. Position 1s were located right underneath the third bead’s center at the first layer, and position 2s were located right underneath the re-martensitized regions in-between the second and third beads at the first layer. The SEM images for the two positions are shown in Figure 3.18. Based on the discontinuity of the lath-shaped microstructure, the martensite at all four positions was partially tempered. However, by comparing the images for H-L in Figures 3.18 (a) and (b) with those for H-H in 3.18 (c) and (d), more martensite had decomposed in the case where higher heat input was used in the 2nd layer.
Particularly, Figure 3.18 (d) shows coarser carbides (shown as bright rods) in the high heat input case (H-H) than the lower heat input case (H-L). In summary, the microstructure results observed in SEM clearly show that the extent of tempering in the 1st layer’s HAZ was higher when the high heat input was used for depositing the beads of the 2nd layer.

Figure 3.18: SEM images for: (a) clad H-L position 1, (b) clad H-L position 2, (c) clad H-H position 1 and (d) clad H-H position 2
3.5.3 Hardness measurement

The micro-hardness map for the single bead on plate deposited using the medium heat input is shown in Figure 3.19. The substrate’s HAZ, i.e., the region adjacent to the fusion line exhibited very high hardness about 500 Vickers hardness number (HVN). For comparison, the base metal hardness of SA-533 steel was only 220 HVN. Combining with the SEM results shown in Figure 3.14 (a) and (b), it is further confirmed that such hard microstructure in the substrate’s HAZ was martensite as expected.

Figure 3.19: Hardness map for single bead on plate weld deposited with medium heat input

To examine the effect of heat input on the size of hardened HAZ, the hardness profiles along the weld centerline were extracted from the micro-hardness maps and plotted together in Figure 3.20 for the three levels of heat input. Despite the large difference in the heat input, the peak hardness was about 500 HVN for all three welds. This again confirmed that the microstructure was martensite, whose hardness was found to be a strong function of steel composition (especially the carbon concentration) and a much weaker function of cooling rate [35]. Outside the high hardness region, the
hardness dropped gradually to the base metal hardness as the distance to the fusion line increased.

Figure 3.20: Hardness distribution along the depth direction at the weld centerline for the three single bead on plate welds shown in Figure 3.12

Figure 3.21 compared the hardness distributions of multiple bead clads made with three different sets of welding parameter: (a) M-M-M, (b) H-L-L and (c) H-H-H. For all three clads, the micro-hardness maps show some high hardness spots near the start and end weld toes of the first layer, possible due to the “edge” effect as those outside HAZs were exposed to less weld thermal cycles compared to the interior HAZs. As shown in Figure 3.21 (a), the hardness of the interior HAZ for M-M-M was significantly reduced to about 330 HVN when compared to the peak hardness of 500 HVN in the single bead on plate weld deposited using the medium heat input (shown in Figure 3.19). This was in contrast to the hardness distribution of clad H-L-L. Figure 3.21 (b) shows that only a narrow band near the substrate fusion boundary had a hardness dropped to 330 HVN. The hardness in a large portion of the substrate’s HAZ remained high. Particularly, the
regions which hardness was above 500 HVN were likely to contain the untempered martensite. Finally, for clad H-H-H shown in Figure 3.21 (c), the high hardness spots were almost completely disappeared in the interior HAZ. A large portion of clad H-H-H substrate’s HAZ had hardness below 300 HVN, and the small remaining portion had hardness around 330 HVN. These regions were likely to have tempered martensite. The hardness distribution further illustrated the increasing tempering effect by increased linear heat input while the power ratio was kept constant.

![Hardness Maps](image)

Figure 3.21: Hardness maps for (a) M-M-M clad, (b) H-L-L clad, and (c) H-H-H clad

The significant difference in tempering resulted from the three sets of heat input was further observed in Figure 3.22, where the hardness profiles along the weld centerline were plotted. The two hardness profiles for single bead on plate with medium and high heat inputs were used to represent those prior to tempering corresponding to M-
M-M, H-L-L, and H-H-H, respectively. As shown in this figure, there was a much lower hardness decrease and thus a less extent of tempering in the substrate’s HAZ for H-L-L than M-M-M and H-H-H.

![Hardness profiles along the depth direction at the weld centerline for single bead on plate welds and multi-bead clads](image)

Figure 3.22: Hardness profiles along the depth direction at the weld centerline for single bead on plate welds and multi-bead clads

It is noted that the micro-hardness measurement can be more sensitive to the local microstructure variation due to the smaller sampling area compared to the conventional hardness testing. For example, the micro-hardness reading can appear high for a tempered martensite microstructure composed of a few hard carbide precipitates surrounded by soft ferrite. However, the significantly higher hardness over a much wider region in H-L-L compared to M-M-M and H-H-H indicated that the hardness difference shown in Figures 3.21 and 3.22 was not due to the local scatter of microstructure.
With the alloying addition of Mo in SA-533, this type of steel was hard to temper especially down to base metal hardness. This situation could be seen from Figure 3.21 where all three heat inputs were not able to temper the substrate’s heat-affected zones to somewhere close to 220 HVN (base metal hardness), and this was due to the reason that Mo is a strong carbide former which can retard cementite precipitation as well as the tempering reaction [12].

In summary, the hardness results clearly indicated that the linear heat input could have a significant effect on the extent of tempering when the power ratio was kept constant.

### 3.5.4 Gleeble thermal simulation

The Ac1 and Ac3 temperatures were first measured in Gleeble by a dilatometer to make sure the highest tempering temperature was below the steels’ lower transformation temperature (Ac1). In Figure 3.23, the testing results show that the as-quenched steel SA-533 had its Ac1 and Ac3 of 747 °C and 836 °C, respectively.

Next, tempering heat treatment was done in the Gleeble at different temperatures and times. The Gleeble simulated samples were metallographically prepared, and their hardness after tempering was measured. The results are summarized in Table 3.3. It is interesting to note that the lowest hardness obtained at 700 °C for 60 s was still 282 HVN, fairly comparable to that observed in the hardness maps from the actual multiple bead clads. Also, there is an increase in hardness after tempering at 700 °C for 600 s, possibly due to the secondary hardness effect discussed previously in Figure 2.3. Particularly, for
the prolonged tempering at 700 °C, M₆C type of carbide (e.g., molybdenum carbides) could form along the grains boundaries, replacing the cementite precipitates [13].

Using the hardness values, the extent of tempering (Φ) is calculated using Equation (2.5). The as-quenched hardness used in the calculation is around 480 HVN. This data is plotted as a function of tempering time for different temperature in Figure 3.24. It is found that Φ is a strong function of temperature for the SA-533 steel. For example, for a tempering time of 0.5 s, Φ increased from 0.22 to 0.7 as the temperature increased from 400 to 700 °C. On the other hand, Φ did not show a strong time dependence for this steel. For instance, at 700 °C, Φ was raised only very slightly from 0.7 to 0.74 as the tempering time increased from 0.5 to 60 s. This suggests that the extent of tempering in temper bead welding of SA-533 steel is mainly dedicated by the local peak temperature experienced.

![Figure 3.23: Lower (Ac1) and higher (Ac3) transformation temperature measurement of as-quenched SA-533 steel](image)

Figure 3.23: Lower (Ac1) and higher (Ac3) transformation temperature measurement of as-quenched SA-533 steel
Table 3.3: Hardness for Gleeble simulated samples at different tempering temperatures and times

<table>
<thead>
<tr>
<th>Tempering time (s)</th>
<th>Tempering temperature (°C)</th>
<th>Hardness (HVN)</th>
</tr>
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<tbody>
<tr>
<td>0.5</td>
<td>700</td>
<td>297</td>
</tr>
<tr>
<td>1</td>
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<td>296</td>
</tr>
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<td>282</td>
</tr>
<tr>
<td>600</td>
<td>700</td>
<td>361</td>
</tr>
<tr>
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<td>340</td>
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<tr>
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<td>600</td>
<td>316</td>
</tr>
<tr>
<td>0.5</td>
<td>500</td>
<td>380</td>
</tr>
<tr>
<td>5</td>
<td>500</td>
<td>379</td>
</tr>
<tr>
<td>600</td>
<td>500</td>
<td>344</td>
</tr>
<tr>
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<td>422</td>
</tr>
<tr>
<td>5</td>
<td>400</td>
<td>398</td>
</tr>
<tr>
<td>600</td>
<td>400</td>
<td>384</td>
</tr>
</tbody>
</table>

Figure 3.24: Plot of the extent of tempering as a function of tempering time
3.5.5 **Modeling of thermal cycle**

As discussed in the previous section, the third layer of cladding beads imposed a much lower tempering effect on the substrate’s HAZ than the second layer due to the much lower peak temperature. Hence, only the first and second layers were modeled to a first approximation. Moreover, as shown in Fig 3.21, the hardness maps for all clads exhibit periodicity due to the repeated deposition of weld beads. By taking advantages of such periodicity, only the first couple of weld beads in the first and second layers were considered in the current weld heat transfer model.

For clad M-M-M, five beads were modeled: the first three beads in the first layer and the first two beads in the second layer, as shown in Figure 3.25 (a). The distribution of peak temperature caused by deposition of the two beads in the second layer was plotted in Figure 3.25 (b). From the calculated temperature evolution on the weld cross section, the thermal cycles at the thermocouples’ location were extracted. The calculated thermal cycles and the experimental data measured by the thermocouples for bead beads #1 (in the first layer) and #6 (in the second layer) were shown in Figure 3.26 (c) and (d), respectively. The calculated peak temperature, heating and cooling rates were consistent with the experimental data.
Figure 3.25: (a) Mesh for clad M-M-M, and (b) distribution of peak temperature caused by deposition beads on the second layer.

Figure 3.26: Welding thermal cycles for clad M-M-M. (a) and (c) are for the first layer, and (b) and (d) are for the second layer.
Figure 3.27 (a) showed the mesh for clad H-L-L. The mesh included the first three beads in both the first and second layers. Figure 3.27 (b) plotted the distribution of peak temperature caused by depositing the three beads in the second layer. By comparing to that shown in Figure 3.25 (b), the peak temperature caused by depositing the second layer for H-L-L diffused to a shallower depth than that for M-M-M. The comparison of calculated and measured thermal cycles for H-L-L is shown in Figure 3.28 (c) and (d) for weld beads #2 (in the first layer) and #7 (in the second layer), respectively. The calculated temperature results were also consistent with the experimental data.

Figure 3.27: (a) Mesh for clad H-L-L, and (b) distribution of peak temperature caused by depositing beads on the second layer
Figure 3.28: Welding thermal cycles for clad H-L-L. (a) and (c) are for the first layer, and (b) and (d) are for the second layer.

The calculated temperature distribution for the two clads can be used to explain the effect of linear heat input on tempering of the substrate’s HAZ. For this purpose, a monitoring location was selected to position along the centerline of the second bead of the first layer at a distance of 2 mm beneath the fusion line in the HAZ. In other words, this monitoring location was placed at the same relative position (depth) for both clads M-M-M and H-L-L. The thermal cycles calculated at this monitoring location for two clads are shown in Figure 3.29. As shown in Figure 3.29 (a), the temperature profiles experienced at the monitoring location were quite similar when depositing the first layer in both clads. This suggested that the hardened region in the substrate’s HAZ after depositing the first layer was likely to be similar in size for the medium and high heat inputs. Indeed, a similar hardness profile along the weld centerline was observed in the
single bead on plate welds made using the medium and high heat inputs, as shown in Figure 3.20. On the other hand, the peak temperatures experienced at the monitoring location when depositing the second layer for H-L-L were much lower than those for M-M-M, as shown in Figure 3.29 (b). As discussed earlier, the extent of tempering in the HAZ was a strong function of peak temperature as tempering was a thermally activated process. The lower peak temperatures in the substrate’s heat-affected zone by depositing the second layer for H-L-L thus resulted in a lower extent of tempering than M-M-M. Since power ratio was kept constant for H-L-L and M-M-M, the linear heat input used in temper bead welding can have a significant influence on the extent of tempering.

Figure 3.29: Comparison of calculated thermal cycles at a monitoring location positioned in the HAZ at the same relative position to the fusion line of M-M-M and H-L-L: (a) the first layer and (b) the second layer
Although widely used as a quick screening tool, 2-D models do not consider the heat conduction along the travel (or longitudinal) direction. The accuracy of calculated temperature profiles may deteriorate as the travel speed is reduced and the heat input is increased. To improve the simulation accuracy, a preliminary 3-D model in Abaqus was developed for the clad H-H-H. As shown in Figure 3.16 (c), part of the third layer’s HAZ was clearly imprinted in the first layer’s HAZ. This indicated that the high heat input beads on the third layer also had a strong effect on tempering the substrate’s HAZ. Hence, the 3-D model considered the first three beads in the 1st layer, the first two beads in the 2nd layer, and the first one bead in the 3rd layer to better account for the high heat input used, as shown in Figure 3.30. The distribution of peak temperatures is plotted in Figure 3.31. Due to the time constraint, the 3-D model was yet to be validated using the experimental TC data at the time of writing this dissertation. In a future work, the 3-D model will be validated and then used to calculate the extent of tempering.

Figure 3.30: 3-D mesh for clad H-H-H with six beads
3.5.6  **Modeling of heat-affected zone hardness**

Two pieces of information is needed to model the HAZ hardness in temper bead welding: thermal cycles and tempering kinetics. The former (i.e., thermal cycles) can be calculated using the weld heat transfer models. For the latter, as discussed in Section 3.5.4, the extent of tempering is a strong function of tempering temperature and less dependent on tempering time. As a first approximation, the extent of tempering measured at 0.5 s as a function of temperature is established to describe the tempering kinetics. As shown in Figure 3.32, the function of the extent of tempering ($\phi$) at 0.5 s to tempering temperature exhibits a strong linear relationship. Using a linear fit, the equation for tempering kinetics is given as:

$$\phi = 0.0015T - 0.3888$$  (3.2)
where $T$ is the tempering temperature. Using this equation, the $\phi$ can be calculated for a given peak temperature and thereafter the hardness for certain area at the substrate’s HAZ after tempering.

![Graph showing the relationship between the extent of tempering ($\phi$) and tempering temperature](image)

**Figure 3.32:** Predicated relationship between the extent of tempering ($\phi$) at 0.5 s and tempering temperature

In order to validate the applicability of this linear equation developed by Gleeble isothermal tests to an actual temper bead weld, the thermal profiles are extracted from two locations of each 2-D model for clads H-L-L and M-M-M, as shown in Figure 3.33. Both locations in each model are lined up in the depth (vertical) direction in between the first and second beads. Point 1s are located at the CGHAZ of both models and are 0.7 mm underneath the fusion boundary, and point 2s are located at the FGHAZ of both models and are 1.2 mm underneath the fusion boundary. For all four locations, the peak temperatures for the first and second beads are above Ac1, indicating that martensite is
formed in the first bead, and re-formed in the second bead. Subsequently, these locations are not reheated above Ac1 by other beads, and the peak temperature (below Ac1) occurs at the fourth bead in M-M-M model and the fifth bead in H-L-L model, respectively. Using this peak temperature, the extent of tempering is calculated using Equation (3.2) and the results are summarized in Table 3.4. Points 1 and 2 experience more tempering in M-M-M than those in H-L-L.

Experimental data for those locations are extracted from the hardness maps of clads M-M-M and H-L-L shown in Figure 3.21 (a) and (b). As shown in Table 3.4, the predicted hardness values are fairly consistent with the measured values especially for H-L-L. The hardness values also correlate well with the large hardness differences at the CGHAZ and FGHAZ of clads M-M-M versus H-L-L shown in Figure 3.22.

The simple equation shows a good potential to capture the extent of tempering in the HAZ of SA-533 steel. In the future, both time- and temperature-dependent tempering kinetics will be included for better predicting the HAZ hardness profile combined with the thermal cycles calculated by 3-D weld heat transfer model.
Figure 3.33: Extractions of calculated temperature profiles from: (a) M-M-M and (b) H-L-L thermal models

Table 3.4: Comparisons of calculated and measured HAZ hardness

<table>
<thead>
<tr>
<th>Model</th>
<th>Location</th>
<th>Peak Temperature (°C)</th>
<th>( \phi )</th>
<th>Calculated/Measured Hardness (HV)</th>
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<tr>
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<td>746</td>
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<td>290/308</td>
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<td>M-M-M</td>
<td>2</td>
<td>673</td>
<td>0.62</td>
<td>318/293</td>
</tr>
<tr>
<td>H-L-L</td>
<td>1</td>
<td>480</td>
<td>0.33</td>
<td>395/391</td>
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<tr>
<td>H-L-L</td>
<td>2</td>
<td>483</td>
<td>0.34</td>
<td>393/410</td>
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</table>
Chapter 4: Dissimilar Metal Welding

4.1 Materials

As shown in Figure 1.1, the structural weld overlay on pressurizer surge nozzle-to-safe end joint involves dissimilar metal weld of nickel FM 52M on austenitic stainless steel, where dilution plays an important role in affecting the propensity of solidification cracking. To obtain high quality experimental data for studying the weld pool transport phenomena and their effect on dilution, a butt joint configuration was used. The butt joint consisted of a half plate of nickel base alloy 690 and austenitic stainless steel 304L as a surrogate of the DMW in the structural overlay. The base metal nickel alloy 690 was selected and substituted for Alloy 52M (AWS designation ERNiCrFe-7A) in dissimilar metal welds, as Alloy 52M is only produced in the form of filler wire. Since small amount of sulfur can have a significant impact on the weld pool geometry and dilution as shown in Figure 4.1, two heats of stainless steel 304L were evaluated: one with low sulfur (LS) content (0.001 wt%), and the other with high sulfur (HS) content (0.028 wt%). The HS 304L represented the material for cast austenitic stainless pipe or elbow which typically contains a high level of impurities. The chemical composition for both heats of stainless steel 304L and nickel alloy 690 is provided in Table 4.1.
Figure 4.1: Effect of sulfur on weld metal dilution and weld pool shape [12]

Table 4.1: Chemical composition of stainless steel 304L and nickel alloy 690

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<thead>
<tr>
<th>Element</th>
<th>Chemical Composition (wt-%)</th>
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<td></td>
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<tr>
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<td>18.23</td>
</tr>
<tr>
<td>Ni</td>
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<tr>
<td>Mn</td>
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<td>P</td>
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<td>Si</td>
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<td>Mo</td>
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<tr>
<td>Ti</td>
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</tr>
<tr>
<td>Fe</td>
<td>Bal.</td>
</tr>
</tbody>
</table>
4.2 Overview of Experimental Design

To generate high-quality and repeatable experimental data, the welding experiments were focused on making a series of autogenous spot welds (i.e., no filler metal).

Particularly, autogenous spot welds were made on single metals of low and high sulfur stainless steel plates with dimensions of 5 cm long, 5 cm wide, and 1.2 cm thick for initial testing and characterization of weld pool shapes, as shown in Figure 4.2 (a). Next, dissimilar metal welds of nickel alloy 690 with different heats of stainless steel plates were produced, as shown in Figure 4.2 (b). The half plates were cut into 5 cm long, 2.5 cm wide, and 1.2 cm thick, and two dissimilar pieces were tightly clamped together by a clamper shown in Figure 4.3.
4.3 Experimental Procedures

4.3.1 Welding processes

Both single metal and dissimilar metal welds were made using a GTAW process. The single metal spot welds were made by Jetline beam welding control system with the power source of Miller Dynasty 300LX TIG welder shown in Figure 4.4 (a). The dissimilar metal welding was made by a fixed torch with the power source of Miller Dynasty 350 TIG welder shown in Figure 4.4 (b). The welding parameters that applied to all welds were listed in Table 4.2. For single metal welding, all welds were shielded by pure argon. For dissimilar metal welding, two kinds of shielding gases were evaluated: pure argon, and a mixture of argon and helium.
Table 4.2: Welding parameters for single metal and dissimilar metal welds

<table>
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<th>Welding parameters</th>
<th>Single metal welding</th>
<th>Dissimilar metal welding</th>
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<tr>
<td>Current (A)</td>
<td>250 or 300</td>
<td>200</td>
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<tr>
<td>Voltage (V)</td>
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<tr>
<td>Welding time (s)</td>
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<td>Electrode diameter (mm)</td>
<td>2.4</td>
<td>2.4</td>
</tr>
<tr>
<td>Gas flow rate (CFH)</td>
<td>20</td>
<td>20</td>
</tr>
</tbody>
</table>

To obtain accurate temperature profiles around the weld pool for model development, two types of thermocouples were used for thermal cycle measurements: type K and type C thermocouples. The type K thermocouples were attached on the surface of the plate a distance of 10 mm to the weld pool edge; the type C thermocouples were inserted from the back side of the plate to location that was 2 mm below the top surface, as schematically illustrated in Figure 4.5 (a). For DMWs, a total of four pairs of TCs were used: two type K thermocouples on the surfaces of both pieces, and two type C thermocouples inserted through drilled holes for each plate, as schematically illustrated in Figure 4.5 (b). After insertion of type C thermocouples, the holes were sealed by ceramic
paste, which was a type of high temperature chemical set cement made by OMEGA. The backside of an assembled plate with sealed TCs prior to welding is shown in Figure 4.6.

The TC temperature recording was conducted with an instruNet Model 100 analog/digital input/output data acquisition system shown in Figure 4.7. Each pair of thermocouples was connected to a separate channel on instruNet, and temperature was measured at 100 Hz sampling rate with a 4 kHz low pass filter to reduce noise.

Figure 4.5: Thermocouple positions for (a) single metal (b) dissimilar metal welding

Figure 4.6: Ceramic paste for sealing thermocouples
4.3.2 Metallurgical characterization

Each weld was cross-sectioned and mounted in 1.5 inch Bakelite molds in a Leco PR-36 mounting press. The sample was then grinded and polished following the standard procedures as described in Section 3.3.3. Etching of the dissimilar metal welds entailed a two-step process of electrolytic etching in 10% oxalic acid at 5 volts and 0.5 amps to reveal stainless steel microstructure, followed by 10% chromic acid electrolytic etching at 6 volts and 0.5 amps to reveal nickel alloy microstructure. Light optical micrographs were taken in an Olympus GX51 with up to 1000X magnification.

4.3.3 Hardness mapping

After polishing, hardness measurements were conducted on the dissimilar metal welded samples by Leco LM 100AT micro-hardness tester. The hardness for weld metal, mixed zone, heat-affected zone, and base metal was mapped using a load of 100 g and a spacing of 150 μm.
4.3.4 Calculation of weld metal dilution and composition

For the particular autogenous spot welding of Ni alloy 690 to stainless steel 304L, the dilution is the amount of melted Ni alloy 690 into the amount of melted stainless steel 304L during welding. The calculation is shown in Equation (4.1):

\[
\text{Dilution} = \frac{V_{304L}}{V_{304L} + V_{690}}
\]  

(4.1)

where \( V_{690} \) is the volume (\( \text{mm}^3 \)) of melted Ni alloy 690, and \( V_{304L} \) is the volume of melted stainless steel 304L. The volume measurement is done by first taking macrographs of the cross section of dissimilar metal weld. Then the fusion line is digitized as a series of points and reproduced in Abaqus CAE. The cross section is revolved by 180 degrees to create a 3-D geometry of the half pool whose volume is then measured in Abaqus CAE, as shown in Figure 4.8.

![Meshed half weld pool 3-D geometry for (a) Ni 690 side and (b) low sulfur stainless steel 304L side with pure argon.](image)

Figure 4.8: Meshed half weld pool 3-D geometry for (a) Ni 690 side and (b) low sulfur stainless steel 304L side with pure argon.
The weld dilution determined by Equation (4.1) can be further used to calculate the weld metal composition assuming a uniform mixing in the weld pool. EDS scans in SEM are made to measure the actual chemistry compositions at the mixed weld metal.

4.4  Modeling of Weld Pool Fluid Flow

The weld pool model was developed based on ANSYS Fluent, a commercial computational fluid dynamics (CFD) solver, to understand the transport of chemical elements in the dissimilar metal weld pool. A unique characteristic of the dissimilar metal weld pool is the complex chemistry due to the mixing of nickel alloy with stainless steel. Therefore, a pre-requisite for the weld pool model is the dependency of viscosity and surface tension on local chemistry. Existing constitutive equations of viscosity and surface tension in the literature were used and implemented as user defined subroutines in Fluent for the weld pool model. The surface tension was defined in Equation 4.2 [36] [37] [38].

\[
\gamma = \gamma_m^0 - A(T - T_m) - RT \Gamma_s \ln \left[ 1 + k_1 Y_s e^\left(-\frac{\Delta H^0}{RT}\right) \right]
\]

(4.2)

where \(\gamma_m^0\) is the surface tension of pure metal; \(A\) is a constant which expresses the variation of surface tension of pure metal at temperature above the melting point; \(\Gamma_s\) is the saturated surface excess concentration; \(R\) is the universal gas constant; \(Y_s\) is the activity if sulfur in solution; \(k_1\) is the absorption coefficient; and \(\Delta H^0\) is the enthalpy of segregation.

The viscosity was defined in Equation 4.3 [37] [38].
\[
\lg \eta^{\text{alloy}} = \frac{2570}{T} - 0.8224 + 1.75 \times 10^{-3} Y_{\text{Cr}} + 1.1 Y_{\text{Fe}} + 10.2 \times 10^{-3} Y_{\text{heavy}} \quad (4.3)
\]

where \(Y_{\text{Cr}}\) is the mass fraction of Cr; \(Y_{\text{Fe}}\) is the mass fraction of Fe; and \(Y_{\text{Heavy}}\) is the total mass fraction of heavy elements (i.e., \(W + \text{Re} + \text{Nb} + \text{Ta} + \text{Mo} + \text{Hf}\)).

The local composition of dissimilar metal weld is not known a priori and is calculated by solving highly non-linear and coupled transport equations of mass continuity, momentum conservation, energy conservation, and species transport (i.e., convection and diffusion of alloying elements). The model was first built in 2-D to simplify the modeling developing and testing work. A preliminary 3-D model was conducted with parallel computing in Fluent.

### 4.5 Experimental Results and Discussion

#### 4.5.1 Thermal history measurements

For single metal welding, the temperature history on the surface of both low and high sulfur stainless steel plates is shown in Figure 4.9 (a); and the temperature history on the pool bottom of both low and high sulfur stainless steel plates is shown in Figure 4.9 (b).
(a) Surface temperature history measured by type K thermocouples

(b) Weld pool bottom temperature history measured by type C thermocouples

Figure 4.9: Temperature history of single metal welds

For dissimilar metal welding, the temperature history of the pool bottom for both nickel and stainless steel under pure argon shielding is shown in Figure 4.10. The temperature history of the pool bottom under mixed argon and helium shielding is shown in Figure 4.11.
(a) Pool thermal cycle of high sulfur stainless steel and nickel alloy under 100% argon

(b) Pool thermal cycle of low sulfur stainless steel and nickel alloy under 100% argon

Figure 4.10: Dissimilar metal welding under pure argon shielding
(a) Pool thermal cycle of high sulfur stainless steel and nickel alloy under 50% argon and 50% helium

(b) Pool thermal cycle of low sulfur stainless steel and nickel alloy under 50% argon and 50% helium

Figure 4.11: Dissimilar metal welding under mixed shielding gas

The collected temperature data at locations in or close to the weld pool shows peak temperature as high as the melting temperature of the metal, providing high quality data to validate the weld pool model. All temperature curves show rapid heating and cooling during spot welding. Some temperature curves such as the high sulfur stainless steel plot in Figure 4.10 (a) and nickel alloy plot in Figure 4.11 (a), had some small
fluctuation in the peak temperature possibly due to touching of the thermocouples to the molten pool during welding.

4.5.2 Metallurgical characterization

The as-welded single metal welds for high and low sulfur stainless steel are shown in Figure 4.12 (a) and (b), respectively, and their weld bead shapes are shown in Figure 4.13 (a) and (b) for HS and LS, respectively. Both spot welds were made by 300 amps with the same tungsten electrode and argon shielding, and the welds had a diameter of 1.1 cm. Due to different sulfur content, the high sulfur stainless steel welds resulted in a deeper penetration than the low sulfur stainless steel welds, indicating more inward flows driven by Marangoni shear stress during welding. The low sulfur stainless steel welds, were likely to undergo outward flows during welding, resulted relative shallow weld pool shape.

Figure 4.12: As-welded single metal welds: (a) single metal spot welding of high sulfur stainless steel and (b) single metal spot welding of low sulfur stainless steel
For dissimilar metal welding, the as-welded samples of nickel alloy 690 to high and low sulfur stainless steels with argon shielding are shown in Figure 4.14 (a) and (b), respectively. The polished and etched weld cross-section for HS and LS dissimilar metal welds are shown in Figure 4.15 (a) and (b), respectively. Both weld nugget shapes exhibited some asymmetry about the butt interface, as the fluid flow patterns in stainless steels and nickel alloys were likely to be different due to different sulfur content. The molten pool shape suggested that the molten metal fluid flow likely exhibited strong outward flow driven by the Marangoni shear stress, resulting in a wide and shallow pool. In particular, the half nugget on stainless steel plates was wider but shallower than the other half on alloy 690 plates. The weld pool was skewed somewhat toward nickel plates side (with less sulfur content) when welded with high sulfur stainless steel plate.

Figure 4.13: Cross-sectional images of (a) high sulfur (b) low sulfur stainless steels

Figure 4.14: As-welded dissimilar metal welds: (a) nickel alloy and high sulfur stainless steel and (b) nickel alloy and low sulfur stainless steel
Figure 4.15: Cross-sectional images of dissimilar metal welds with pure argon shielding

The cross-sectional microstructures of HS and LS dissimilar metal welds made with mixed shielding gas (argon and helium) are shown in Figure 4.16 (a) and (b), respectively. The ‘hotter’ arc, due to higher thermal conductivity of additional helium, likely induced a stronger inward flow of the molten metal at the pool center, exhibiting both a wider and deeper weld pool shape in dissimilar metal joints.
Figure 4.16: Cross-sectional images of dissimilar metal welds with mixed shielding gas

4.5.3 Hardness mapping

The hardness for two dissimilar metal welds made by pure argon was mapped. The results are shown in Figure 4.17 (a) and (b) for HS and LS welds, respectively. The hardness distribution is relatively uniform in the dissimilar welds especially for the LS
weld, indicating an extensive mixing and thus relatively homogenous chemistry and microstructure in the weld molten pool.

Figure 4.17: Hardness maps for dissimilar metal welds made with pure argon

4.5.4 Calculation of dilution and mixing chemistry

The calculated volumes and thereafter the predicted dilutions are summarized in Table 4.3 for dissimilar metal welds made under different conditions
Table 4.3: Calculated half weld volume and dilution

<table>
<thead>
<tr>
<th></th>
<th>Ni and low S stainless steel</th>
<th>Ni and high S stainless steel</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Ni side</td>
<td>stainless side</td>
</tr>
<tr>
<td>Volume (mm$^3$)</td>
<td>77.23</td>
<td>62.34</td>
</tr>
<tr>
<td>Dilution</td>
<td>0.447</td>
<td></td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th></th>
<th>Ni and low S stainless steel</th>
<th>Ni and high S stainless steel</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Ni side</td>
<td>stainless side</td>
</tr>
<tr>
<td>Volume (mm$^3$)</td>
<td>106.08</td>
<td>49.71</td>
</tr>
<tr>
<td>Dilution</td>
<td>0.319</td>
<td></td>
</tr>
</tbody>
</table>

From Table 4.3, it can be seen that for both Ar and Ar+He shielding gases, as the sulfur content increased in the weld, the amount of melted alloy 690 was reduced while 304L raised. In other words, with a higher sulfur, the alloy 690 is more diluted with 304L, a trend also observed in Figure 4.1. Interestingly, the addition of helium in the shielding gas reduced the extent of dilution of nickel alloy 690 with stainless steel 304L for the same sulfur content.

The EDS results of two point scans at the dissimilar weld metal between nickel alloy 690 and LS and HS stainless steel 304L with pure argon are shown in Figure 4.18. Particularly, the Cr and Ni contents after welding are compared with the predicted results based on the calculated dilution, as shown in Table 4.4 and 4.5. The comparison shows reasonable correlation of the predicted chemistry to the measured extent of mixing, indicating that the weld pool was extensively mixed.
Figure 4.18: EDS point scan of fusion zone at (a) LS stainless steel and nickel alloy weld and (b) fusion zone at HS stainless steel and nickel alloy weld with pure argon

Table 4.4: Comparison of predicted and measured weld chemistry for HS weld

<table>
<thead>
<tr>
<th>Elements</th>
<th>304L HS composition (wt%)</th>
<th>Ni 690 composition (wt%)</th>
<th>Predicted composition after mixing (wt%)</th>
<th>Measured composition after mixing (wt%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cr</td>
<td>18.28</td>
<td>27</td>
<td>22.62</td>
<td>24.38</td>
</tr>
<tr>
<td>Ni</td>
<td>8.02</td>
<td>58</td>
<td>32.86</td>
<td>37.05</td>
</tr>
</tbody>
</table>

Table 4.5: Comparison of predicted and measured weld chemistry for LS weld

<table>
<thead>
<tr>
<th>Elements</th>
<th>304L LS composition (wt%)</th>
<th>Ni 690 composition (wt%)</th>
<th>Predicted composition after mixing (wt%)</th>
<th>Measured composition after mixing (wt%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cr</td>
<td>18.23</td>
<td>27</td>
<td>23.08</td>
<td>25.89</td>
</tr>
<tr>
<td>Ni</td>
<td>8.07</td>
<td>58</td>
<td>35.68</td>
<td>41.61</td>
</tr>
</tbody>
</table>
4.5.5 Modeling of weld pool fluid flow

The 2-D model was conducted first and the calculated results are shown in Figure 4.19 (a) and (b), respectively. In this case, the weld pool surface was assumed to be flat because of autogenous welding without filler metal and under moderate arc current. Since the surface position was known a priori, the model did not consider the free surface flow. The simulation results obtained using the simplified 2-D model showed different fluid flow patterns in 304L and Alloy 690 due to different sulfur content and viscosity, which in turn, resulted in an asymmetry in weld pool shape. In Figure 4.19 (a), the simulation results also indicated a strong outward fluid flow behavior driven by the Marangoni shear stress, resulting in a wide and shallow pool. The distribution of sulfur concentration is plotted in Figure 4.19 (b).

![Figure 4.19: 2-D modeling results of (a) temperature and molten metal flow, and (b) mixing of sulfur in the weld pool](image)

Figure 4.19: 2-D modeling results of (a) temperature and molten metal flow, and (b) mixing of sulfur in the weld pool
Due to the time constraint, the 3-D model was developed only for a single metal and did not consider the species transport at the time of writing this dissertation. For the 3-D model, by taking the advantage of symmetry, the geometry of the model was only half of the weld. The 3-D model dimensions were exactly the same as the experimental plates, and the material considered was the low sulfur stainless steel 304L. Specially coded user defined heat flux functions were applied in Fluent for parallel computing of the weld pool temperature and velocity fields which are shown in Figure 4.20. In the model, the predicted temperature was extracted from the region where thermocouples were placed, and both the simulated and measured temperature profiles during welding are plotted in Figure 4.21. The predicted temperature by the 3-D weld pool model showed reasonable correlations with experimental data, indicating its potential in capturing the transport physics in the weld pool.

Future work will include expanding the 3-D model to include species transport to model the mixing of chemical elements in the molten pool. The 3-D model will be validated using the experimental data of molten pool shape and temperature profiles for both single metal welds and DMWs.
Figure 4.20: Zoomed-in view of weld pool temperature and velocity fields

Figure 4.21: Computational and experimentally measured weld temperature profiles
Chapter 5: Summary, Conclusions and Future Work

5.1 Summary and Conclusions

In summary, two important weldability issues for nuclear component repair were investigated in the present research: (i) HAZ tempering in temper bead welding, and (ii) weld metal dilution in dissimilar metal welding between nickel alloy and austenitic stainless steel.

For temper bead welding, experiments were performed on a SA-533 Gr. B Cl. 1 steel using cold wire GTAW with three different linear heat inputs while keeping the power ratio constant. The metallurgical characterization of the substrate’s HAZ was done by SEM. The extent of tempering in the HAZ was quantified using micro-hardness mapping and Gleeble thermal simulation. 2-D and 3-D weld heat transfer models using the double-ellipsoidal heat flux equation were developed to calculate the peak temperatures and to assess the final hardness in the substrates’ HAZ.

Dissimilar metal welding experiments were performed on stainless steel 304L and nickel alloy 690 using autogenous GTA spot welding. Two heats of stainless steel 304L were used: low (0.001 wt%) and high (0.028 wt%) sulfur content with two shielding gases: pure argon and a mixture of argon and helium (50% Ar-50% He). The welding
temperature profiles were captured by thermocouples placed closely underneath the molten pool. The extent of mixing was characterized using micro-hardness mapping. 2-D and 3-D weld pool fluid flow models using specially coded user defined functions of Marangoni shear stress and composition-dependent viscosity were developed to simulate the dissimilar weld pool molten metal flow behavior.

Through the experiments and numerical simulations, the following summaries can be drawn:

(i) HAZ tempering in temper bead welding:

- The HAZ of a single bead on plate weld on SA-533 steel contains similar hardened microstructure for all three levels of heat input considered (i.e., 685, 1060 and 1432 J/mm), and the hard microstructure is confirmed to be martensite by SEM. This indicates the SA-533 steel has a high hardenability and can easily forms martensite in HAZ during welding.

- The higher the heat input used for depositing the second and third layers, the larger the extent of tempering in the substrate’s HAZ (beneath the first layer). Particularly, the three layer clad deposited by high heat input only (i.e., H-H-H) has the lowest hardness in the substrate’s HAZ than the other two clads (i.e., H-L-L and M-M-M) and thus experiences the largest extent of tempering.

- The Gleeble simulated results show that the extent of tempering in SA-533 steel HAZ is a strong function of tempering temperature and to a much lesser extent a function of tempering time. It is difficult to reduce the HAZ hardness to that of the base metal considering the short time at temperature in welding. Finally, the
tempering at high temperature (700 °C) over an extended period of time (600 s) shows an increase in hardness due to the secondary hardening effect.

- The thermal cycles calculated by the weld heat transfer model are consistent with those measured by thermocouples, indicating the validity of the calculated temperature distribution.

- As the tempering is a strong function of temperature, the peak temperature experienced in the substrate’s HAZ is a useful indicator for the extent of tempering. Particularly, a higher heat input in the 2nd and 3rd layers exposes a larger portion of the substrate’s HAZ to elevated temperatures, thus resulting in a higher extent of tempering. For clads H-L-L and M-M-M, the deposition of the 2nd layer plays an important role on tempering the substrate’s HAZ generated by the 1st layer. For clad H-H-H, due to the high heat input, the deposition of the 3rd layer also played some role on tempering. This distribution of peak temperature due to reheating of substrate’s HAZ by depositing the 2nd and 3rd is consistent with the measured hardness maps.

- Since the power ratio is the same for clads H-L-L, M-M-M, and H-H-H, the substantial difference in the HAZ hardness indicates that the linear heat input can have a significant impact on the extent of tempering in temper bead welding. Hence, both the linear heat input and power ratio should be controlled for welding procedure development of temper bead welding.

(ii) Welding metal dilution in dissimilar metal weld between nickel alloy and austenitic stainless steel:
• For dissimilar metal welding, the weld nugget shape exhibits some asymmetry about the butt interface. In particular, the half nugget on 304L stainless steel plate is wider but shallower than the other half on alloy 690 plate.

• Higher sulfur content changes the surface tension gradient to positive which promotes an inward flow at the pool center, resulting in relatively narrower but deeper weld pool shape.

• The “hotter” arc due to the addition of helium results in both a wider and deeper weld pool shape in the dissimilar metal joint.

• The hardness distribution in the dissimilar metal joint is relatively uniform, indicating an extensive mixing and thus relatively homogenous chemistry and microstructure in the dissimilar metal weld pool.

• The preliminary weld pool simulation results indicate a strong outward fluid flow driven by the Marangoni shear stress, resulting in a wide and shallow pool. The fluid flow patterns in 304L high sulfur stainless steel and alloy 690 are different due to the different sulfur content and viscosity.

• In the 3-D model, the calculated temperature profiles show reasonable correlations with the experimental measured data, indicating its potential in capturing the transport physics in the weld molten pool.

The present research constitutes a significant step toward an improved understanding of the two important weldability issues for the repairing of nuclear structures. Particularly, the high-quality, comprehensive experimental data generated will be valuable to develop and validate computational weld models in the future. Such
validated models in turn can be used for process optimization for temper bead and dissimilar metal welding.

5.2 Future Work

Future work for temper bead welding will include 3-D modeling of the substrate’s HAZ microstructure evolution by combining the calculated thermal cycles and hardness equations generated via isothermal Gleeble test. Charpy impact toughness testing will be important to evaluate the toughness of tempered martensite and be correlated to the hardness measured in the tempered HAZ.

Future work for dissimilar metal welding will include the additional metallurgical characterization in SEM and EDS, and weld pool fluid flow modeling with calculation of Lorenz force and species transport in 3-D.
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