WELDING WITH LOW ALLOY STEEL FILLER METAL OF X65 PIPES INTERNALLY CLAD WITH ALLOY 625: APPLICATION IN PRE-SALT OIL EXTRACTION

THESIS

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By

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Abstract

Oil extraction from traditional subsea reserves has been successfully accomplished in the past using internally clad X65 pipelines that are joined with Alloy 625 filler material, which matches the clad material. With the discovery of pre-salt oil fields off the coast of Brazil, which are located at greater depths below the sea floor than most oil fields, there is interest to use a pipe reeling technique to place the pipelines into service. With the reeling technique, there is a need for the welds joining the internally clad X65 pipes to overmatch the yield strength of the pipe material by 100 MPa. This ensures that deformation during reeling and straightening occurs in the pipe, rather than the welds. Also, this process will not allow for post weld heat treatments and acceptable weldability needs to be achieved. Therefore, Alloy 625 filler material does not meet the requirements for the pipe reeling, namely the yield strength overmatch. Instead, low alloy steels are being considered as fill passes to join these internally clad pipes that have a root pass of Alloy 625 to fuse the clad material. There is uncertainty as to the weldability between these materials due to concerns of solidification cracking related to using a higher solidification temperature range filler metal over substrate with lower solidification range.

In this study, low alloy steel welding consumables, ER100S-G and ER70S-6, were considered as potential fill pass consumables and an additional Ni-based alloy, Alloy 686, was considered aside from Alloy 625 as root pass material. Finally, in the event that low alloy steel couldn't easily be welded directly over these Ni-based alloys, buffer materials (UTP A 80 Ni and Alloy 625 LNb) were considered to isolate the root and fill passes. To study these new materials, computational modeling, bead-on-plate welding trials, and groove welding trials were performed. Thermodynamic simulations were used to determine solidification temperature ranges and phase formation in the range of dilutions between two materials. Welding design of experiments in the bead-onplate (BOP) position was performed to test favorable material combinations and attempt to optimize parameters to assess if metallurgical defects could be eliminated. Welding trials on a flat, narrow groove geometry were performed for compatible materials to verify that the materials were compatible in a higher restraint scenario.

It was found by these techniques that ER100S-G was more compatible with both Ni-based alloys than ER70S-G, specifically with Alloy 686. In BOP welds of ER100S-G over Alloy 625, solidification cracking and liquation cracking was encountered, but no cracks were experienced when welding over Alloy 686. Also, a defect previously thought to be solely related to castings, called shrinkage porosity, was found in all welds of ER100S-G over both Alloy 625 and 686. In addition, the buffer layer material UTP A 80 Ni showed the best compatibility with ER100S-G of all materials that were tested.

In groove welding, the best solution with no buffer, ER100S-G welded directly over Alloy 686, showed centerline solidification cracking that was related to travel speed and was able to be eliminated with parameter optimization. Limited shrinkage porosity defects were encountered in groove welding. Lack of fusion (LOF) defects where the welds contact the sidewall and previous weld (triple-point) were also encountered due to a low heat input welding process, Cold Metal Transfer, being used. The weave amplitude of these welds greatly affected the LOF defect formation.

Although cracking was eliminated in the BOP welding trials of ER100S-G over Alloy 625, the Ni-based substrate was eliminated from the study due to 3 potential metallurgical defects that can occur. Instead, testing with Alloy 686 was more successful and ER100S-G has been successfully welded over this alloy without cracking in a narrow groove geometry. UTP A 80 Ni as a buffer layer material shows promising results in computational modeling simulations as well as BOP welding trials, but has not yet been attempted in groove welding trials.

This study determined that welding low alloy steel over Ni-based alloy can be performed using optimized parameters without solidification or liquation cracking, however, shrinkage porosity defects may occur as a result of the LAS consumable having a higher melting temperature than the Ni-based substrate.

Dedication

To my parents, Daniel and Connie, for their unconditional love, endless support, and

outstanding guidance.

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Chapter 1: Introduction & Motivation

With the ever increasing demand for oil in the current global energy climate, there has been a recent push to find oil reserves large enough that can answer that demand. Recent exploration in 2006 has resulted in the discovery of pre-salt subsea oil reserves in the Santos Basin that are located up to 300 kilometers off the coast of Rio de Janeiro (Figure 1), spanning 800 km along the coast of Brazil [1]. The depths of the reserves are roughly 7 km below sea level, which requires special considerations due to harsher conditions at these extreme depths. The layers under which the pre-salt layer is located are shown in Figure 2. Finally, the economic viability of the pre-salt reserves has a good outlook, with predictions of 4 million barrels/day to be collected from these reserves at the peak extraction period [2].



Figure 1: Location of pre-salt reservoirs [3]



Figure 2: Schematic of pre-salt depths [4]

While oil extraction from this location is very attractive, special considerations must be observed. Pre-salt oil reservoirs have higher H₂S and CO₂ content than traditional oil reserves, higher temperatures, pressures, and depths are experienced, and this particular reserve is located far from the coast. The location of the pre-salt reserves makes reeling an attractive option due to cost considerations. Reeling involves spooling the welded pipeline on a large reel and allows welding and inspection to be performed onshore. The reeling vessel transports the spooled pipe to the final service location to be straightened and installed. In order to ensure that deformation during the reeling and straightening operations occurs in the pipeline rather than the joints, a strength overmatch is required for the welded joints. In addition, the X65 pipelines that are used to extract the product need to be internally clad with a corrosion resistant alloy (Alloy 625) to protect the steel from the high H₂S and CO₂ contents in these areas. As such, the girth welds that join these internally clad pipelines that will subsequently undergo reeling need to meet three criteria:

- 1. Overmatch the pipeline material yield strength by 100 MPa (DNV OS F101 [5])
- 2. Similar service properties as pipeline material without utilizing post weld heat treatment
- 3. Have acceptable weldability

These girth welds were previously performed with Alloy 625, but the strength requirement cannot be achieved with this consumable. Instead, low alloy steels are being considered as fill passes welded over a nickel-based root pass of Alloy 625 (Figure 3). This dissimilar combination could be susceptible to solidification cracking from the dilution of the low alloy steel weld metal by Alloy 625.



Figure 3: Schematic of welding configuration in groove geometry

Over the course of this study, potentially viable low alloy steel consumables will be explored with the utilization of low heat input weld processes. These materials will undergo computational modeling, bead-on-plate weld testing, and narrow groove weld testing. The overarching goal is to find a combination of Ni-based root pass material and low alloy steel fill pass material that will be free of defects, meet the strength requirement, and require no post weld heat treatment. Additionally, if a metallurgically compatible combination is not found, a buffer layer will be considered to isolate the root and fill passes.

Chapter 2: Background

2.1 Materials for Dissimilar Metal Welds in Oil & Gas Industry

To construct pipelines for use in subsea oil extraction, the oil and gas industry utilizes a variety of materials. The materials need to be corrosion resistant since they are exposed to crude oil, which contains hydrogen sulfide (H_2S). For the pipeline material, steel is often used that can be internally clad with a corrosion resistant alloy for corrosion protection and cost effectiveness. The materials that are used to join the pipelines need to have acceptable weldability, high strength, and high toughness to avoid potential failures. These materials used in the O&G industry will be explained in the following sections.

2.1.1 Steel Pipe Material

In the O&G industry, X65 pipeline grade steel is commonly used for oil extraction. These materials act as the pipelines that transport oil from its original location under the sea floor, as such, these materials need a good combination of strength, toughness, and weldability [6]. These advantageous properties are gained through specific alloying additions and thermomechanical processing. The American Petroleum Institute (API) creates the standards for strength for pipeline steels and these steels are named according to their minimum yield strength [7]. For example, an X65 grade steel has a minimum yield strength of 65 ksi. Due to the strength overmatch requirement of 100 MPa, the mechanical properties of the material of interest are presented in the units of MPa rather than ksi (Table 1). It should also be noted that these steel pipelines can be internally clad with a corrosion resistant alloy to protect the steel from corrosion from the product that is being transported. Internal cladding is also performed for the cost saving aspect, since the entire pipeline does not need to be fabricated from a corrosion resistant alloy.

API Grade	Yield Stre	ngth (MPa)	Tensile Strength (MPa)				
	Minimum	Maximum	Minimum	Maximum			
X65	450	600	535	760			

Table 1: Mechanical Properties of API Grade Pipeline Material X65 [7]

As previously stated, the beneficial combination of strength, toughness, and weldability is gained through alloying and thermomechanical processing. Examining the alloying additions, the carbon content is limited due to weldability and toughness concerns with detrimental microstructure formations [8]. Strong carbide formers (Nb, V, Ti, etc.) are crucial to the properties of these steels as they provide precipitation strengthening, which impedes grain growth by limiting grain boundary motion [6, 9, 10]. Solid solution strengthening is also achieved with additions of Mn, Mo, and Ni. A study on line pipe steels summarized the effects of each alloying addition, however, an in depth explanation will not be provided here [11]. Instead, the composition of the X65 grade of interest is presented in Table 2.

API		Material Compositions (wt. %) (maximum values)											
Grade	Fe	С	Si	Mn	Р	S	V	Nb	Ti	Cu	Ni	Cr	Mo
X65	Rem.	0.18	0.45	1.7	0.025	0.015	*	*	*	0.5	0.5	0.5	0.5

 Table 2: Material Composition for X65 Steel [7]

*Unless otherwise agreed, the sum of the Nb, V, and Ti concentrations shall be ≤ 0.15 %

The other method to achieve the desired properties of steel pipe material is thermomechanical processing. This process involves heating to austenitization temperatures and soaking at this temperature for a length of time, followed by straining during cooling [12, 13]. During straining, the austenite grain size is reduced, which simultaneously increases strength and toughness [6]. This grain size reduction is a result of the fine precipitates, such as NbC, that hinder grain boundary movement as well as recrystallization of the microstructure. The fine austenitic grain structure transforms to a fine ferrite structure upon reaching the transformation temperatures. A schematic of the process is provided in Figure 4.



Figure 4: Thermomechanical processing schedule for a microalloyed steel, T_S = soaking temperature, T_{nr} = non recrystallization temperature, A_{R3} and A_{R1} = ferrite start and finish temperatures [12]

2.1.2 Filler Materials

There are numerous filler materials used in the O&G industry for the joining of steel pipelines, however, this study focuses on solid solution strengthened nickel-based alloys and low alloy steels. These materials can be used for dissimilar metal welds, which requires an understanding of each material to ascertain properties, microstructure, and potential defects when one material is diluted by the other. Each material will be explained in the following sections.

2.1.2.1 Solid-Solution Strengthened Nickel-based Alloys

Solid-solution strengthened (SSS) nickel-based alloys derive their strength from the namesake, alloying additions that cause solid solution strengthening. Due to their exceptional corrosion resistance and moderate strength level, SSS nickel-based alloys are used in a wide array of applications. The yield strength levels achievable by this family of nickel-based alloys is between 345-480 MPa [14, 15], which is the driving force for finding a new alloy to weld the fill passes for the internally clad pipes in order to meet the 550 MPa yield strength requirement. Nickel-based alloys can achieve higher strength levels through precipitation of γ ' and γ '' in precipitation-strengthened nickel-based alloys, but this class of alloys will not be discussed.

As the name implies, SSS nickel-based alloys gain strength through alloying additions that act as substitutional atoms in the matrix. All SSS nickel-based alloys solidify as a face centered cubic (FCC) austenite with secondary phases and/or carbides at the grain boundaries depending on the alloy. These alloying atoms have the effect of expanding the FCC lattice, which strengthens the material. Cr, Fe, Mo, W, and Cu may act as these substitutional atoms to provide solid-solution strengthening, and Co, Nb, and Ta can also be used in some alloy systems [14]. Carbide formation can be promoted with the addition of Nb, Ti, W, Mo, Ta, and Cr. Carbon additions will preferentially segregate to the grain boundaries and form carbides with the aforementioned alloying additions. Cr and Mo additions also greatly improve corrosion resistance [14]. Most SSS nickel-based alloys have good weldability and will ideally be resistant to solidification cracking due to control of solidification temperature range and some alloys having the ability to backfill and heal cracks that form during solidification.

A large amount of research has been performed on Alloy 625 (ERNiCrMo-3) to study its solidification and segregation characteristics [16, 17]. For this purposes of this study, Alloy 625 solidifies as austenite with NbC and Laves constituents in the microstructure. In addition, when a weld is placed over Alloy 625, the HAZ created in the 625 will remain relatively unchanged from the original weld metal aside from limited grain growth [14]. Alloy 686 (ERNiCrMo-14) is another SSS nickel-based alloy that contains a tungsten addition rather than the niobium addition of Alloy 625, and also contains a higher molybdenum content. Studies have been performed to compare the properties of Alloy 625 and 686 [18-21] and the main trends of these studies conclude that Alloy 686 has increased pitting and crevice corrosion resistance attributed to the W addition and the increased Mo content. A comparison between Alloy 625 and 686 microstructures is shown in Figure 5.



Figure 5: Microstructural comparison of (A) Alloy 625 (B) and Alloy 686 [20]

2.1.2.2 Low Alloy Steels

Low alloy steels are widely used in industry and account for 95% of metals used in construction and fabrication [22]. They are readily weldable, relatively cheap, and have good mechanical properties, both high strength and toughness [23]. Low alloy steels can achieve yield strengths between 400 and 730 MPa [24] and are strengthened by one or more of the following methods [25]:

- 1. Grain refinement
- 2. Solid solution strengthening
- 3. Precipitate strengthening
- 4. Transformation strengthening
- 5. Dislocation strengthening

To understand the metallurgy of low alloy steels, the Fe-C phase diagram (Figure 6) can be used along with knowledge of non-equilibrium transformations. For the range of carbon contents in low alloy steel (<0.2%), austenite exists at high temperatures and will transform to ferrite and pearlite, which is a mixture of ferrite and cementite. Ferrite is a relatively soft phase while pearlite is brittle. When analyzing non-equilibrium cooling conditions such as welding, martensite and bainite may form [23]. Carbon content controls the hardness of martensite, which is one of the hardest phases that is possible in low alloy steels, but can be tempered to reduce hardness and improve toughness. Low alloy steels depend on solid solution strengthening, precipitation strengthening, and grain refinement for their properties.

Traditionally, low alloy steels typically contain less than 8 wt. % alloying additions and virtually all of the types of low alloy steels contain carbon, manganese, and silicon [23]. Carbon and manganese additions cause the transformation from austenite to

ferrite to occur at a lower temperature, which will control the grain size to avoid a loss in strength from grain growth. Additionally, very small additions of vanadium, niobium, and titanium can be used to form carbides and nitrides to prevent grain growth and encourage precipitation hardening in some cases [23, 26, 27]. Chromium, nickel, molybdenum, copper, nitrogen, and zirconium are also used in small amounts depending on the alloy and property requirements.



Figure 6: Fe-C phase diagram [23]

The shielding gases used with low alloy steel are often argon or an $Ar+CO_2$ mix. The CO₂ addition increases the penetration of the weld, however, spatter also increases with the addition [28]. Finally, low alloy steel, under certain conditions, can be susceptible to certain forms of cracking, such as solidification cracking and hydrogen induced cracking. These cracking mechanisms are discussed in the Weldability Issues section 2.2.

Finally, when attempting to perform dissimilar welds where low alloy steel will be interacting with nickel-based alloy, constitution diagrams can be used to roughly predict the microstructure based on the chemical composition of the resultant weld metal. The Schaeffler diagram is one such diagram that can be used with this combination, though it is not designed for it (see Solidification Cracking section 2.2.1). Another constitution diagram developed by Gould was specifically designed for microstructural predictions with the low alloy steel/nickel-based alloy combination [29]. Unfortunately, the materials that were used in the creation of the constitution diagram were not matching with the materials in this study, so the applicability is limited.

2.2 Weldability Issues

When welding low alloy steel over a nickel-based alloy, there are a variety of potential defects that could plague the resulting weld. Besides process related defects, such as lack of fusion, metallurgical defects can also occur. Some defects will arise during welding and solidification, but some may require an incubation period and not transpire until days after welding is completed. Normally, low alloy steels are relatively resistant to solidification cracking due to their crack resistant microstructure. However, dilution of a low alloy steel weld by a nickel-based alloy can render the microstructure susceptible by creating an austenitic structure. Similarly, liquation cracking is also a possibility during the welding of subsequent passes with a low alloy steel filler if grain boundary liquid forms in a prior pass that has an austenitic microstructure.

A defect previously thought to be related only to castings, known as shrinkage porosity, is also a possibility with this material combination. The main reason for this defect is the large difference in liquidus/solidus temperatures between low alloy steel and nickel-based alloy. Additionally, these welds may also be susceptible to hydrogen induced cracking if a susceptible microstructure, threshold level of hydrogen, restraint, and ambient temperatures are simultaneously experienced.

2.2.1 Solidification Cracking

When welding low alloy steel and nickel-based alloy in a dissimilar weld application, solidification cracking may be a concern due to dilution of the low alloy steel weld by the nickel-based substrate. Solidification cracking is a hot cracking defect that occurs along solidification grain boundaries (SGB) during solidification. In order for cracking to occur, the weld must have an imposed restraint (thermal and/or mechanical) and a susceptible microstructure (austenite) [30]. This defect occurs during the terminal stages of solidification as intrinsic and extrinsic stresses acting on the grains exceed the strength of the "almost completely solidified weld metal" [27]. It is the liquid remaining at the grain boundaries that cannot accommodate the strain, and a crack forms at the
SGB. Some examples of solidification cracking are shown in Figure 7. Fracture surfaces of solidification cracks can be described as "eggcrate" in appearance (Figure 8), having a morphology that is cellular or dendritic [30]. There are multiple theories to explain the mechanism of solidification cracking: the shrinkage brittleness theory by Borchvar et al. [31-33], the strain theory by Pellini [34, 35], the generalized theory of supersolidus cracking by Borland [36], the modified generalized theory by Matsuda et al. [37, 38], and the technological strength theory by Prokhorov [39]. While multiple theories have been developed, they all require that a liquid film forms at the grain boundary towards the end of solidification and the grains are pulled apart by tensile stresses. Many weldability tests have been developed to study solidification cracking response from various alloys [40].



Figure 7: Solidification cracking along solidification grain boundaries, (A) 7075 aluminum [41], (B) Ni-based Alloy 718 [30]



Figure 8: "Eggcrate" appearance of solidification crack fracture surface, (A) aluminum alloy, (B) Ni-30Cr alloy [30]

The main factors that contribute to solidification cracking are primary solidification phase, solidification temperature range (STR), amount and distribution of liquid in last stage of solidification, surface tension of the GB liquid, and grain structure [27]. Materials that solidify as austenite are typically more prone to solidification cracking than materials that solidify as ferrite. As such, austenitic stainless steels and nickel-based alloys are particularly susceptible [30]. Increasing the STR, defined as the difference of the liquidus temperature and the solidus temperature, results in an increased susceptibility to solidification cracking due to the presence of a larger, weaker "mushy" zone where both solid and liquid exist. STR normally increases with the addition of alloying elements as well as the presence of impurities such as sulfur and phosphorus. In low alloy steels as well as nickel-based alloys, it is known that impurity elements preferentially segregate to the grain boundaries and lower the melting temperature, which increases the overall STR [27, 42]. In austenitic microstructures, sulfur and phosphorus

have a low solubility in austenite, which leads to the grain boundaries being enriched with the impurity content. Varestraint testing was used to compare experimental iron and nickel-based alloys and a trend was found that directly correlated a larger STR to a worsening cracking response (Figure 9) [14]. Studies using austenitic stainless steels have found that decreasing impurity content (S+P) narrows the STR of the weld and lowers solidification crack susceptibility [43-45].



Figure 9: Varestraint results showing effect of STR on solidification cracking susceptibility [14, 46]

Extensive research has been accomplished to create composition-based equations to calculate a cracking susceptibility factors (CSF) for carbon/low alloy steels, which were later summarized by Matsuda [47]. Many of the equations predict an increase in CSF with additions of carbon, sulfur, and phosphorus, with their additions lowering the

STR of the steel and potentially increasing grain boundary wetting. Manganese is a notable alloying addition that can lower the CSF due to interacting with sulfur to form MnS, rather than allow sulfur to enrich the GB liquid or form FeS, which is known to lower solidification cracking resistance [30]. It should be mentioned that these relationships apply to undiluted weld metal.

In the case of low alloy steel (ER100S-G), it solidifies with a BCC microstructure and the STR is relatively small. These characteristics render the resulting weld metal relatively resistant to solidification cracking. However, the nickel-based substrate (Alloy 625 or 686) has an austenitic (FCC) structure and a large STR [14]. When welding with a low alloy steel filler material over a nickel-based alloy substrate, dilution of the weld metal by the substrate occurs, according to Equation 1 and Figure 10, altering the STR and potentially the microstructure. The dilution can be easily measured using a geometric calculation (Figure 10) or a scanning electron microscope.

Dilution =
$$\frac{B}{A+B} * 100 (\%)$$

Equation 1: Geometric dilution determination, A is area of weld reinforcement, B is area of weld penetration (see Figure 10)



Figure 10: Geometric dilution measurement with important areas shown

Once the low alloy steel is diluted by the nickel-based alloy, the resultant weld metal may be susceptible to solidification cracking. The microstructure can be predicted using the Schaeffler diagram [48], but it is not extremely accurate and initially was developed for stainless steels [14]. Although the chromium and nickel equivalents of nickel based alloys are outside of the range of the diagram, tie lines can be constructed between the two materials of interest to evaluate the microstructure prediction according to dilution (Figure 11). The materials used in this study are predicted to be fully austenitic above ~23% dilution, which may increase susceptibility to solidification cracking.



Figure 11: Schaeffler diagram with two different base metal/filler metal combinations 19

In addition to impurity elements, niobium also expands the STR and increases susceptibility to solidification cracking in the presence of iron. Studies by DuPont et al. with nickel-based superalloys and dissimilar metal welds with Alloy 625 and super austenitic stainless steel showed that the partition coefficient (κ) of niobium decreased as iron content was increased [46, 49, 50]. The partition coefficient (Equation 2) is a way to quantify segregation behavior in alloying additions, with a value of 1 indicating that segregation will not occur. As the k value decreases below 1, preferential segregation occurs towards the liquid, meaning that alloying addition will remain in the grain boundary liquid towards the end of solidification. Thus, the conclusions of the studies by DuPont et al. were that niobium would preferentially segregate to the grain boundaries in the presence of iron, which would lead to a depression of the melting temperature of the grain boundary from the formation of a niobium-rich eutectic constituent [49]. The increased segregation of Nb in the presence of Fe was the result of Fe decreasing the solubility of Nb in austenite (γ), thus it partitions to the liquid and will be pushed to the grain boundaries [50]. Schematics of the relationship found in these studies are shown in Figure 12.

$$\kappa = \frac{C_S}{C_L}$$

Equation 2: Partition coefficient (κ), composition of solid at solid/liquid interface (C_S), composition of liquid at solid/liquid interface (C_L)



Figure 12: Partition coefficient (κ) of niobium as a function of iron content of Ni-19Cr alloy, (A) Ni-19Cr alloy [49], (B) Alloy 625 and super austenitic SS dissimilar weld [50]

While niobium plays a role in solidification cracking in nickel-based systems containing iron, certain alloys exhibit a "backfilling" effect that can cause solidification cracks to be refilled by remaining liquid before freezing is completed, effectively healing the cracks [30, 42]. A widely used nickel-based alloy, Alloy 625, exhibits this behavior by creating a large amount of NbC eutectic liquid or Laves phase towards the end of solidification that can heal any cracks that may form during welding (Figure 13). This backfilling effect can render alloys resistant to solidification cracking that were initially susceptible to solidification cracking due to a large STR.



Figure 13: Crack healing effect of Alloy 625 [30]

Weld restraint also plays a large role in solidification cracking susceptibility. A large intrinsic restraint is experienced due to solidification shrinkage, which ranges from 3 to 8% for most weld materials [30]. Additionally, when considering groove welding, weld bead profile greatly affects the restraint level that the weld experiences. Bead concavity causes a large restraint if the weld bead bridges the entire joint, so a flat or slightly convex bead can be used to decrease the restraint experienced by the last material to solidify and reduce solidification cracking susceptibility. Depth/width ratio also greatly affects restraint levels and should be considered when creating welding schedules [30]. Bead geometry effects are shown in Figure 14. Bead solidification shape can also affect solidification cracking response, as a tear drop shape weld pool can encourage centerline solidification cracking, while an elliptical pool shape will be more resistant [27]. A study of girth welds of low carbon iron joining X80 pipes concluded that solidification cracking

response was enhanced at higher travel speeds and that cracking was avoided at low travel speeds [51].



Figure 14: Effect of geometry, placement, and depth/width ratio on solidification cracking [52]

Mitigation techniques have been used to avoid solidification cracking, such as controlling the primary solidification phase, controlling weld restraint, and reducing impurity contents. In dissimilar welding applications, a wide array of dilution levels are possible, so McCracken et al. studied the solidification cracking response of GTAW welds of 52M nickel-based alloy over 304L stainless steel by comparing power ratio to dilution. The power ratio (Equation 3) value takes arc power and cross sectional weld area into account and was developed by EPRI to control dilution for welding processes where "wire feed is independent of weld energy" [53]. In their study, they found that power ratio held a positive relationship with dilution and that above 45% weld dilution, solidification cracking occurred (Figure 15). While this study involved welding of nickelbased alloy over stainless steel, the importance of power ratio was proved and can be applied to find a relation to dilution where normal heat input relations are not found.

Power ratio = $\frac{Power}{Cross \ sectional \ area} = \frac{Amperage * Voltage}{\frac{WFS}{TS} * Cross \ sectional \ area \ of \ filler}$

Equation 3: Power ratio equation, WFS = wire feed speed, TS = travel speed [53]



Figure 15: Weld dilution vs. power ratio for 52M welds over 304L base material in 2G position [53]

2.2.2 Liquation Cracking

When welding low alloy steel to join pipeline that has a root pass of nickel-based alloy, subsequent passes of LAS after the initial LAS pass could cause weld metal liquation cracking. The initial LAS pass may be susceptible to this defect due to dilution from the root pass in the weld metal that causes a susceptible microstructure, austenite, to form. Liquation cracking is a hot cracking defect that can occur in the HAZ of a weld (HAZ liquation cracking) or in reheated weld metal (WM liquation cracking) adjacent to the fusion zone. Liquation cracking occurs above the effective solidus of the material during cooling and develops at grain boundaries [30]. The mechanism of liquation cracking involves the melting of grain boundaries in the HAZ or reheated weld metal as well as a restraint from solidifying weld metal and shrinkage that opens the liquated grain boundaries (Figure 16) [27].



Figure 16: HAZ liquation cracking mechanism [27]

In this study, WM liquation cracking will be considered since it is more likely to occur in this application than HAZ liquation cracking. Austenitic stainless steels and

nickel-based alloys are particularly susceptible to WM liquation cracking if they contain a single phase structure (Figure 17). The defect can occur at both solidification grain boundaries (SGBs) and migrated grain boundaries (MGBs) in reheated weld metal and can form when a shallow temperature gradient develops adjacent to the fusion boundary in the substrate. If a shallow temperature gradient exists, the melting temperature of the grain boundary can be exceeded due to the melting point depression caused by impurity and alloy segregation to these locations (Figure 18) [30].



Figure 17: WM liquation cracking; (A) austenitic stainless steel [30], (B) Alloy 625 [14]



Figure 18: Effect of temperature gradient on local melting of grain boundaries [30]

To mitigate liquation cracking, composition, grain size, and heat input must be considered. Since WM liquation cracking requires melting of grain boundaries, impurity elements (P, S, and B) that segregate to the grain boundaries and lower the melting temperature should be kept to a minimum [30, 54]. Another mitigation technique is to ensure that there is some second phase in the austenitic structure, such as ferrite. This second phase will affect the wetting characteristics of the liquated GB [30]. In addition, small grain sizes in the material will decrease the susceptibility to liquation cracking due to having more grain boundary area that prevents full wetting of the GB liquid [14, 55, 56]. The small grain size also creates a stronger structure that can better account for the stress buildup. Related to grain size, heat input also plays a role in liquation cracking mitigation. Low heat input will create a steeper temperature gradient adjacent to the fusion boundary, which will decrease the extent of liquation (Figure 18) [57]. Finally, reducing restraint can reduce the likelihood of liquation cracking.

2.2.3 Shrinkage Porosity

Shrinkage porosity is a new defect in welding that was previously encountered in casting processes. Shrinkage porosity is different from gas porosity sometimes found in welding and could potentially be found at the interface of welds involving a higher melting point (T_m) consumable welded over a lower T_m substrate. There is an important distinction between gas and shrinkage porosity in that gas porosity is formed by entrapped gases in a casting or weld and shrinkage porosity forms due to the contraction of the casting during solidification and does not require entrapped gases to form, though a combination of the two is possible. Also, the shape of the two different forms of porosity are determined by the timing of formation and provide a method to determine which defect is encountered. Gas porosity normally occurs as round bubbles that occur towards the beginning of solidification and shrinkage porosity develops later and is often interdendritic (Figure 19) [58].



Figure 19: Microscopic view of pores in 7Si-0.4Mg alloy; (A) round pore (gas porosity), (B) interdendritic pore (shrinkage porosity) [58]

Before discussing the theory of why shrinkage porosity forms in dissimilar welds of a high T_m consumable over a low T_m substrate, it is useful to understand the general mechanisms of shrinkage. During solidification of a liquid, three distinct shrinkages occur, illustrated in Figure 20. The three types of shrinkages are liquid contraction, solidification contraction, and solid contraction [58].



Figure 20: Shrinkage regimes: liquid, solidification, and solid [58]

When considering the mechanism of shrinkage porosity it relates to welding, the first two shrinkage regimes, liquid and solidification shrinkage, are of particular interest. Also, the microstructure of the resulting solid is another important factor, as it will dictate the amount of shrinkage that occurs, with face centered cubic (FCC) structures experiencing a greater degree of volume change than body centered cubic (BCC). FCC structures experience a greater contraction due to having a closer packed structure than BCC. As a comparison, pure nickel (FCC structure) experiences a volume decrease of 5.11% and iron (BCC) a decrease of 3.16% [58].

There are multiple types of shrinkage porosity that can be encountered in castings, but the type that closely relates to welding is shrinkage porosity that occurs in thick sections (Figure 21). This type is known as internally nucleated porosity and is normally experienced by short-freezing-range alloys like aluminum-bronze. During casting, the outside skin freezes quickly and the shrinkage pores can develop when the pressure from feeding is insufficient. Thick sections in castings are normally prone to this type of defect. A pore can develop as shrinkage is occurring from solidification when the remaining liquid in the system becomes elastically stretched until a critical value is reached. This critical value is known as the fracture pressure of the liquid. Once the fracture pressure is exceeded, a pore forms to relieve the stress buildup, similar to a crack forming under high stress that provides stress relief to the system. After the pore develops, further growth is possible as solidification progresses. The defect has been documented in a variety of steels, nickel alloys, and aluminum alloys (Figure 22).



Figure 21: Internally nucleated shrinkage porosity [58]



Figure 22: Shrinkage porosity in various alloy systems, (A) CG8M steel [59], (B) A356 aluminum alloy [60], (C) M35-1 nickel alloy [61], and (D) 319-type aluminum alloy [62]

Significant effort has been spent attempting to find a solution to shrinkage porosity because it can be a serious defect. Niyama et al. created a criteria to determine if shrinkage porosity will occur in castings depending on temperature gradient (G) and cooling rate (T) (Equation 4) at the end of solidification [63]. The initial studies showed how, regardless of solidification time, the criterion could be used to determine porosity formation (Figure 23). Upon its establishment as an important factor in casting, it was added to casting software packages to model processes and predetermine defect formation [64].

$$Ny = \frac{G}{\sqrt{\dot{T}}}$$

Equation 4: Niyama criterion that depends on temperature gradient (G) and cooling rate (\dot{T}) [63]



Figure 23: Effect of solidification time (t_f) on the Niyama Criterion [63]

Multiple studies by Carlson et al. have used the Niyama criterion to predict porosity formation in high nickel steel and nickel alloys [61, 65]. A dimensionless form of the Niyama criterion was also developed to better account for the properties and solidification characteristics of the alloy of interest [64, 66]. The applicability of this criteria has not been considered for welding related shrinkage porosity, but it may benefit from the examination of the methods of defect elimination in castings.

Once shrinkage porosity has formed, hot isostatic pressing (HIP) has been used to decrease porosity size. Yong et al. [67] noted how HIP could be used, but adds an additional processing step, so any method that eliminates the defect from forming is a better solution. Their study found that creating a temperature gradient in TiAl alloys that

encourages solidification from the bottom to the top of the mold, rather than transverse solidification from the mold walls to the center of the mold (mold-temperature-gradient method), improves shrinkage porosity resistance.

Some concerns for the shrinkage porosity defects is whether they can be identified with inspection techniques and how they degrade material properties. A study by Ghaffari et al. explored ultrasonic characterization of aluminum castings due to the ultrasonic attenuation that occurs as the pores scatter the waves [62]. Reference specimens were used (pore free) as comparison with the parts with porosity to accurately measure the ultrasonic attenuation and use density derived volume fractions of shrinkage porosity to characterize the presence of the defect.

Another study executed by Hardin and Beckerman [68] evaluated the fatigue life of cast AISI 8630 steel components containing shrinkage porosity. After determining the porosity distribution in each sample, Finite Element Analysis (FEA) software was used to model the stress fields developed by the defects and the local elastic mechanical properties were decreased as the volume of pores in a given area increased. As a result, the specimens containing shrinkage porosity had a greatly reduced fatigue life than reference data of the same material with no defects (Figure 24).

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Figure 24: Fatigue life data comparison between AISI 8630 steel with and without shrinkage porosity [68]

2.2.3.1 Welding Higher Melting Point Consumables over Lower Melting Point Substrates

When welding dissimilar materials, the resulting weld contains many different zones or regions (Figure 25). Near the fusion boundary, a transition region must exist between the weld metal composition and base metal and the potential to form an unmixed zone of base metal exists which was first proposed by Savage et al. [69, 70]. It is in these zones where shrinkage porosity has been found in weldments when welding a high melting point consumable over a lower melting point substrate. Multiple studies that will be discussed below are helpful in explaining the mechanism for shrinkage porosity formation, since it has not been documented in welding to date.



Figure 25: Regions of a fusion weld [30]

During welding, segregation can occur on micro- and macroscopic scales. Microsegregation refers to the redistribution of solute during solidification, which leads to the grain boundaries in the weld having a different composition than the interior of the grains [27, 71]. Alternatively, macrosegregation occurs on a much larger scale, with incomplete mixing of a weld pool being an example of the phenomenon [27]. For the purpose of this study, it is helpful to examine work done with welding of a higher T_m consumable over a lower T_m substrate. This topic has been studied in depth by Kou and Yang [72-77]. The primary reason for the experiments was to examine macrosegregation in dissimilar welds using Cu-Ni and Al-Si alloys. Multiple mechanisms were presented to describe the phenomenon of macrosegregation in welds with a weld liquidus temperature (T_{LW}) below and above the base metal liquidus temperature (T_{LB}). Mechanism 1 is used to describe macrosegregation when T_{LW} is greater than T_{LB} and directly relates to welding low alloy steel over a Ni-based alloy. The mechanism deems complete mixing throughout the entirety of the weld metal impossible. In addition to incomplete mixing, there theoretically must exist a stagnant or laminar-flow layer of liquid called an unmixed zone, referred to in the study as a filler deficient beach, which consists of melted base material that does not mix with the weld. This zone must form because at the interface of the bulk weld material and the base material, the temperature is equal to T_{LW} . This temperature is higher than T_{LB} , so melting of the base material must occur. The zone thickness is a function of liquidus temperature difference ($T_{LW} - T_{LB}$), so a larger difference will lead to a larger unmixed zone.



Figure 26: Fusion boundary macrosegregation when using $T_{LW} > T_{LB}$. (A) – filler deficient beach formation; (B) – weld metal intrusions, beach, peninsulas, and islands formed according to Mechanism 2 [72]

This unmixed zone is susceptible to random intrusions of weld metal, which will freeze quickly upon entering the colder unmixed zone (Figure 27). As a result of weld

metal intrusions into the unmixed zone, filler deficient peninsulas and islands can form, which can remain liquid after the surrounding weld metal intrusions have solidified. It is in and around these regions (unmixed zone, filler deficient peninsulas, and filler deficient islands) where shrinkage porosity has been encountered. Some examples of unmixed zones and weld metal intrusions found in these studies are presented in Figure 28.



Figure 27: Weld metal intrusions in unmixed zone during and after welding [76]



Figure 28: Mechanism 2, unmixed zone (filler deficient beach) and weld metal intrusions in Cu-Ni alloy dissimilar weld [76, 78]

2.2.4 Hydrogen-Induced Cracking

Hydrogen-induced cracking (HIC) is a type of cold cracking that can occur in the weld metal or heat affected zone of weldments in the presence of hydrogen. In order for HIC to occur, four factors are required simultaneously: hydrogen content, high restraint, a susceptible microstructure (martensite), and low temperatures (-100 to 200°C) [27]. However, if one of these factors is absent, HIC can be avoided. Cracking can occur immediately after welding, but can also occur after an incubation period. The hydrogen content is typically introduced into the weldment during the welding process through the dissociation of water vapor, hydrogen gas, and hydrogen bearing compounds (grease, oil). After dissociation, the atomic hydrogen can be absorbed into the weldment [30]. The process of hydrogen diffusion in a weld is shown in Figure 29, with T_F representing the transformation temperature of austenite to ferrite/pearlite and T_B the transformation temperature of austenite to martensite [79]. As the weld metal transforms to ferrite/pearlite from austenite, hydrogen is rejected from the structure due to the decreased solubility of hydrogen in ferrite/pearlite than austenite. The hydrogen enters the austenitic HAZ from the weld metal due to the high diffusion coefficient of hydrogen in ferrite. However, hydrogen has a low diffusion coefficient in austenite and is trapped in the structure, which then transforms to martensite as cooling progresses.



Figure 29: Hydrogen diffusion from weld metal to HAZ during welding process [79]

Restraint is provided by the shrinkage experienced during solidification as well as thermal contraction and is usually hard to control in production scenarios. The microstructure is an important factor for HIC, which is ultimately dictated by the composition and cooling rate experienced. High hardness microstructures are most susceptible to the defect, with martensite being particularly vulnerable, being both hard and brittle. Near ambient temperatures are needed because high temperatures allow the hydrogen to diffuse and avoid congregation at susceptible areas in the weld. Alternatively, low temperatures render the hydrogen immobile [27, 30].

Although a single theory cannot explain all aspects of HIC, multiple theories have been proposed to explain various mechanisms that occur. Although they will not be discussed in detail, they are as follows: planar pressure theory by Zapffe and Sims [80], surface adsorption theory by Petch [81, 82], decohesion theory by Troiano [83], hydrogen-enhanced localized plasticity theory by Sofronis et al. [84-88], and Beachem's stress intensity model by Beachem [89]. To avoid HIC, multiple studies have shown that preheating and applying interpass temperatures (Figure 30) to the materials being welded is an effective method [90]. Preheating and interpass temperatures slow the cooling rate of the weldment, which is beneficial to reduce the cooling rate and hinder martensite formation as well as to remain above the temperature range in which HIC is likely to occur. Additionally, in multi-pass welding where PWHT is not applicable, a weld schedule may be developed that effectively tempers underlying passes and the HAZ to reduce susceptibility of HIC [30]. To quantify susceptibility to this defect, the delayed hydrogen cracking test can be used, which involves charging a sample with hydrogen and subjecting the material to a tensile load [91]. The amount of time for failure at a particular stress level is used to quantify and compare material susceptibility to HIC.



Figure 30: Effect of preheat on HIC of a high strength steel [90]

2.3 Welding Processes

When welding low alloy steel over a nickel-based alloy, dilution of the weld

metal by the substrate is a serious concern due to the possibility of solidification

cracking. In order to decrease the dilution of the weld metal by the substrate, a lower heat input welding process should be used. The low heat welding processes that were available for this study were pulsed gas metal arc welding (GMAW-P) and Cold Metal Transfer (CMT). Both are low heat input variations of standard gas metal arc welding. Of the two low heat input processes, it was been found that CMT has a lower current requirement to deposit the same amount of material as in GMAW-P [92]. This leads to the conclusion that CMT will have a lower heat input than GMAW-P. Subsequently, CMT will also have less penetration into the substrate material, which could assist in keeping the dilution to a minimum. When considering other low heat input variants of GMAW, CMT has lower heat input than Surface Tension Transfer (STTTM), Regulated Metal Deposition (RMDTM), and short arc GMAW [93]. CMT also boasts a ~50% increase in welding speeds compared to standard GMAW [94]. Finally, research has been done to predict weld profiles for both CMT and GMAW-P using a neural network and interpolation. These techniques were not employed in this study, but could become useful in determining weld dilution in future studies [95].

2.3.1 Gas Metal Arc Welding

The Gas Metal Arc Welding (GMAW) process utilizes an electric arc to melt continuously fed filler material and base material to create a weld. The filler wire acts as an electrode that conducts the current and melts into the weld pool and Argon and/or CO₂ are typically used as shielding gases to prevent atmospheric contamination. GMAW is a popular welding process due to its ability to be use in any position and its ease of automation [96]. The process can be used with various metal transfer types depending on the welding parameters. At low currents, the transfer type will likely be globular, which consists of the formation of large droplets at the end of the filler wire that are transferred to the weld by gravitational forces. Low heat input can be achieved, but this transfer type is difficult to control due to the variation in droplet size and arc instability. At the other end of the spectrum, spray mode is achieved at high currents. Spray transfer mode is more easily controlled but suffers the drawback of high heat input, which will cause excessive dilution.

In order to overcome the drawbacks of each transfer type, pulse transfer can be used. The transfer works by pulsing between high and low current. A balance is achieved with the pulse so the mean current does not exceed the threshold level to allow for spray transfer. The high peak current serves to detach a molten droplet at the end of the wire, which enters the molten weld pool, and the low background serves to maintain the welding arc (Figure 31). If using the proper pulse parameters, a single droplet is transferred to the weld pool during each pulse.



Figure 31: Pulse cycle of GMAW-P [97]

A comparison of transfer modes is displayed in Figure 32. The user has the control to program the peak and background current as well as the corresponding time spent at each level. Additionally, pulse frequency can be adjusted. These additional variables compared to GMAW adds complexity to parameter optimization to create stable welds, but in the case of the Fronius power supply used in the project, these values are controlled synergically based on wire feed speed. The use of synergic GMAW-P simplifies the job of the operator by automatically selecting appropriate pulsing parameters using pre-programmed algorithms based on wire feed speed, electrode size, material, and shielding gas [97].



Figure 32: Comparison of transfer modes of GMAW: (a) globular, (b) spray, and (c) pulse [96]

2.3.2 Cold Metal Transfer

Cold Metal Transfer is a low heat input variant of gas metal arc welding that was patented by Fronius in 2004 [98]. The process works by first initiating an arcing phase, where a weld pool is established and a molten droplet begins to form at the end of the filler wire. The wire is fed towards the weld pool and a short circuit is established once contact is made. Once the short circuit is detected, the arc current is reduced to a low level and after a set period of time, the wire begins retracting. The mechanical retraction of the wire assists in the detachment of the molten droplet. After detaching the droplet, the arc reignites and the cycle is repeated. This cycle operates at approximately 60 Hz. To ensure low heat input, the process works with the electrode remaining positive (EP) throughout the weld. A typical CMT cycle, containing arcing and short circuit components, as well droplet formation and detachment is demonstrated in Figure 33 and Figure 34 respectively.

The decrease in current during the short circuit period decreases the overall heat input, which can lead to lower dilution levels compared to GMAW. The low heat input

capabilities of CMT make it an attractive choice to avoid high dilution of the low alloy steel welds by the nickel-based root pass. Additionally, the electrical characteristics are controlled synergically by the wire feed speed, so the desired wire feed speed and selected synergic line program determines the current and voltage. The complex electrical characteristics during a cycle are separated into an arcing phase, where a droplet forms on the electrode, and short circuit phase, where the wire makes contact with the weld pool and is retracted. Due to the fact that CMT is a waveform controlled process with large fluctuations in current and voltage over a short period of time, it was necessary to implement more advanced methods to measure heat input as described in the next section.



Figure 33: CMT cycle, containing arcing phase to create molten drop on the wire and a short circuit phase when the electrode makes contact with the weld pool. The decrease in current at the end of the arcing phase ensures that the molten droplet remains on the end of the wire to be deposited during the S/C phase [98]



Figure 34: Cold Metal Transfer droplet detachment over one cycle [99]

2.3.3 Heat Input

It is often useful to measure the heat input of welds using Equation 5 to identify the energy per unit length of the bead. To accomplish this, the current and voltage must be measured and the travel speed known. For traditional constant voltage GMAW, the method of measuring current and voltage does not greatly affect the calculation because their waveforms remain relatively constant during welding. RMS voltage and current can be used to calculate heat input in these situations [97].

Heat Input =
$$\frac{V * A}{TS}$$

Equation 5: Heat input equation for non-waveform controlled welding processes; V = voltage, A = current, and TS = travel speed

However, when using a waveform controlled process, such as GMAW-P or CMT, there is a large fluctuation in electrical characteristics in very short periods of time. Previous studies utilizing GMAW-P [97] and CMT [100, 101] processes identified the importance of measuring waveform controlled process electrical characteristics. With GMAW-P, using RMS power, the heat input was roughly 10% higher than the actual heat input of welding. Also, using average power, the heat input underestimated the actual heat input of welding by roughly 12%. Rather than use the RMS or average power values, instantaneous values were obtained and an average instantaneous heat input could be calculated. Average instantaneous current and voltage values can be obtained by using a data acquisition system and sampling at a rate at 10 times the frequency of interest, to prevent aliasing [97]. In order to measure these instantaneous values, a data acquisition system can be used in conjunction with a voltage divider and current shunt to step down the electrical characteristics to protect the sensitive data acquisition system [100, 101]. After gathering the instantaneous electrical data, an average can be found and the values can be plugged into Equation 5 to find the heat input.

Chapter 3: Objectives

The primary objective of this study is to evaluate the applicability of low alloy steel filler metals as fill pass material for welding of internally clad X65 pipes that have a Ni-based alloy root pass. The higher strength requirements of welds on pipes that undergo reeling are not met by existing welding methods of Alloy 625 fill passes. Low alloy steel meets the strength requirement, however, solidification cracking is a concern with this metallurgical combination when welding low alloy steel over Alloy 625. Additionally, a different root pass material (Alloy 686) and buffer layer materials (UTP A 80 Ni and Alloy 625 LNb) are also of interest if welding low alloy steel over Alloy 625 is not possible. Three phases of this project will be performed to accomplish this objective:

3.1 Computational Modeling

Thermo-Calc[™], a thermodynamic simulation tool, will be used to predict nonequilibrium solidification temperature ranges, phases that occur during solidification, and partitioning effects of certain alloying additions. The objective is to identify favorable material combinations that can be tested in the next phase of the study, bead-on-plate welding.

1. Create pseudo-binary phase diagrams between each material that is being tested for use in this dissimilar weld configuration to ascertain solidification temperature ranges and expected phases. This involves simulations between low alloy steel and Ni-based alloys, both root pass material and buffer material. If a particular combination is predicted to be susceptible to a certain defect, it will be eliminated from analysis.

 Predict partitioning effects of alloying additions in combinations of low alloy steels and Ni-based alloys.

3.2 Bead-on-Plate Welding Experiments

Welding performed on a flat, horizontal plate with low alloy steel over a layer of Ni-based alloy. Use Cold Metal Transfer and Pulsed Gas Metal Arc Welding to assess low heat input welding processes. The objective is to find viable metallurgical combinations that may work in the next phase of the project, groove welding.

- 1. Perform preliminary welding experiments with a wide range of parameters to identify successful parameters.
- Design of experiment studies to find operational windows for defect avoidance.
 Welding of low alloy steel over different Ni-based alloys, both root pass materials and buffer materials.

3.3 Groove Welding Experiments

Narrow groove welding will be carried out in the flat, horizontal position to transition into narrow groove welding of pipe sections. The objective is to study defect formation in narrow groove geometries and strive to eliminate all metallurgical defects.
- 1. For successful combinations in the bead-on-plate welding experiments, perform narrow groove welding experiments in the flat, horizontal position.
- 2. Refine welding parameters such that all metallurgical defects are eliminated from the low alloy steel fill passes.

Chapter 4: Materials & Procedures

4.1 Materials

In this study, solid-solution strengthened nickel-based alloys, low alloy steel, and nearly pure nickel consumables were used. For the base material, two different types of steel were used in place of the X65 base material that will be used in production. X65 was not available in plate form, only pipe, so other steels were used: P11 and 1018. To simulate the root pass of the groove welds, Alloy 625 and Alloy 686 were used. The main difference between the two is that Alloy 625 has a ~3.5% niobium addition and Alloy 686 contains no niobium. Alloy 686 also contains a small tungsten addition and higher chromium content than Alloy 625. The interest in attempting both was to ascertain if the niobium addition affected cracking response. For the low alloy steel consumables, ER100S-G and ER70S-6 were used to simulate fill passes over the nickel-based alloy root pass. To evaluate the potential use of a buffer layer between the root passes and fill passes, UTP A 80 Ni and Alloy 625 LNb were used. UTP A 80 Ni is nearly pure nickel with a small titanium addition. Alloy 625 LNb is similar to Alloy 625, but with no niobium addition. The compositions of each material used in this study are summarized in Table 3. The mechanical properties of the welding consumables and base materials are also summarized in Table 4.

т	Carada	щ								Materia	al Comp	osition (v	wt. %)							
гуре	Grade	Ħ	Fe	С	Si	Mn	Р	S	Cr	Мо	Ni	V	W	Al	Со	Cu	Ν	Nb	Ti	Other
	AU (25	1	0.30	0.01	0.04	0.03	0.01	< 0.01	21.90	9.00	64.90	0	0	0.12	0.003	<0.01	0	3.50	0.20	0
Root	Alloy 625	2	0.457	0.011	0.10	0.021	0.002	0.001	21.552	8.742	65.30	0.005	0.007	0.108	0.025	0.025	0.012	3.425	0.173	0
Pass	AU (0(1	0.28	0.004	0.032	1.5	0.001	0.001	20.41	16.39	58.44	0	3.72	0.33	0	0.01	0	0	0.08	0
	Alloy 686	2	0.31	0.005	0.04	0.29	0.003	0	22.60	15.79	56.70	0.032	3.77	0.227	0.014	0.06	0	0	0.018	0
	ER100S-	1	95.97	0.11	0.58	1.50	0.008	< 0.001	0.25	0.49	0.89	0	0	< 0.01	0	0.19	0	0	0	0
Fill Pass	G	2	95.93	0.108	0.57	1.62	0.0073	0.004	0.278	0.478	0.926	0.0014	0.004	0.0129	0.0122	0.032	0.007	0.0026	0.003	0.0005 (Zr)
	ER70S-6	N/A	97.24	0.085	0.84	1.54	0.007	0.008	0.03	< 0.01	< 0.04	0	0	0	0	0.20	0	0	0	0
Buffer Laver	UTP A 80 Ni	N/A	0.09	0.02	0.02	0.033	0.003	0.001	0.002	0.003	96.50	0	0	0.06	0	0.003	0	0.004	2.955	0
Luger	Alloy 625 LNb	N/A	0.296	0.004	0.065	0.007	0.005	0.003	22.22	9.92	64.27	0.003	2.929	0.092	0.002	0.012	0	0.007	0.087	0
	X65	N/A	97.95	0.08	0.26	1.40	0.009	0.002	0.08	0.07	0.03	0.04	0	0.036	0	0.01	0.006	0.027	0	0.004 (B)
Base	P11 Steel	N/A	96.30- 97.66	0.05- 0.15	0.50- 1.00	0.30- 0.60	0.025	0.025	1.00- 1.25	0.44- 0.65	0	0	0	0	0	0	0	0	0	0
	1018 Steel	N/A	98.81- 96.26	0.14- 0.20	0	0.60- 0.90	< 0.040	< 0.050	0	0	0	0	0	0	0	0	0	0	0	0

 Table 3: Compositions of Welding Consumables and Base Materials

Туре	Grade	Shielding Gas	Yield Strength	YieldTensileStrengthStrength		Elongation	Testing
			Ν	lpa	Noten (J)	(70)	Condition
Root	Alloy 625	Ar+30%He+0.5%CO2	480	760	130 @ - 196C	42	As-Welded
	Alloy 686	Ar+30%He+2%H2+0.1%CO2	550	790	55 @ +20C	30	As-Welded
D :11	ER100S-G	Ar+15-25%CO2	630	720	95 @ -40C	22	As-Welded
FIII	ER70S-6	Ar+15-25%CO2	520	620	100 @ - 40C	25	As-Welded
Duffer	UTP A 80 Ni	Unknown	>300	>450	>160 @ RT	>30	As-Welded
Duller	Alloy 625 LNb	Unknown	Unknown	Unknown	Unknown	Unknown	Unknown
Base	X65	N/A	482	566	313 @ - 30C	Unknown	N/A

Table 4: Mechanical Properties of Welding Materials

4.2 General Methodology

A summary of the project methodology is provided in Figure 35. Each stage of the project will depend on the one previous, so if material combinations are found to be non-favorable, they will be eliminated from analysis and further testing will not be performed.



Figure 35: General methodology

4.3 Computational Modeling

Thermo-CalcTM (version 2015b) was used to analyze the interaction between materials and screen consumable compatibility. The use most relevant to this study was the creation of pseudo-binary phase diagrams. Once two materials were selected for compatibility analysis, Scheil-Gulliver solidification simulations were executed to output phase transformation temperatures under non-equilibrium cooling conditions, which is a better relation to welding conditions than equilibrium analyses. The simulations were performed in 10% dilution increments, ranging from pure consumable to pure base material and the entirety of the phase transformation temperatures were graphed to create a pseudo-binary phase diagram. Multiple databases were used according to what element was most prevalent in the simulation. If iron was the largest contribution in the simulation, TCFE5 and TCFE8 were used. If nickel was the largest contribution in the simulation, TTNI7 and TCNI8 was used. Different iron and nickel databases were used due to software upgrades in Thermo-Calc[™] that updated each database. Carbon was designated as a fast diffusing element and no phases were rejected from the simulations. Silicon and impurity elements (S and P) were omitted from the analyses due to their inclusion often underestimating the solidus temperature. Also, the solidus temperature for each simulation was determined as the temperature when 98% solid exists.

Thermo-Calc[™] was also used to compare the segregation characteristics of various elements in each material combination. Partition coefficients, which are a measure of how likely an alloying element will segregate to grain boundary liquid during solidification, were chosen as the metric for segregation characteristics. Alloying

elements that do segregate to the liquid during solidification are of importance to this study because solidification cracking response may worsen if the last liquid to solidify has a reduced freezing temperature due to the partitioning elements. For the combination of ER100S-G and Alloy 625, niobium and molybdenum were chosen for modeling and partition coefficients were calculated. For the ER100S-G and Alloy 686 combination, tungsten and molybdenum were chosen for the analysis. For each dilution, the composition of each element in the liquid and solid was determined when the alloy was 98% solid. Using a ratio of the composition of each element in the solid and liquid (Equation 2), the partition coefficients were calculated.

4.4 Welding & Design of Experiment Studies

A Motoman MA1400 six-axis robotic arm (Figure 36) equipped with a CMT-GMAW welding torch and DX100 pendant (Figure 37) was used in conjunction with a Fronius CMT Advanced power source (Figure 36) equipped with an RCU 5000i user interface (Figure 37) and VR 7000 digitally controlled wire feeder to carry out all welding experiments. Electrical characteristics were measured using a voltage divider (Figure 38) and current shunt (Figure 39) connected to an instruNet Model 100 analog/digital input/output data acquisition system (Figure 40). The voltage divider and current shunt were needed to step down the electrical characteristics to a safe level that would not damage the data acquisition system. The voltage divider equation (Equation 6) was used to step down the voltage by a factor close to 10, which ensured that the data acquisition system was not damaged. To step the electrical characteristics back up to their original values, the voltage reading from the voltage divider was multiplied by the ratio of resistances (Equation 6) and the voltage reading from the current shunt was divided by the resistance of the shunt (1.667*10⁻⁴). The information gathered by the data acquisition system was later used to calculate average instantaneous power. This was done by calculating the power for each data point provided by the data acquisition system to calculate instantaneous power for each point. These instantaneous power values were averaged over the length of the weld to output average instantaneous power, which was later used to calculate average instantaneous heat input. When determining heat input, two different travel speeds were used in the calculations. The longitudinal travel speed was used as well as the oscillating travel speed of the torch tip to ascertain if either metric had a direct effect on pertinent outputs.



Figure 36: (A) Motoman MA1400 robot, (B) Fronius CMT Advanced power supply



Figure 37: (A) DX100 pendant, (B) Fronius RCU 5000i pendant



Figure 38: Voltage divider used to step down voltage; $1 = 36.03 \text{ k}\Omega$, $2 = 3.15 \text{ k}\Omega$



Figure 39: Current shunt (1.667*10⁻⁴ ohms)



Figure 40: instruNet Model 100 analog/digital input/output data acquisition system

$$V_{\text{Out}} = V_{\text{In}} * \frac{R_2}{R_1 + R_2}$$

Equation 6: Voltage Divider Equation

For this study, low alloy steels, ER100S-G and ER70S-6, were welded directly over nickel-based alloys, Alloy 625 and Alloy 686, in the flat, horizontal position to

simulate LAS fill passes welded over a Ni-based root pass. Buffer layers, UTP A 80 Ni and Alloy 625 LNb, were also tested with a layer of buffer material separating the Nibased alloy and LAS. Defect formation for all combinations was studied. For the base material, 12" x 12" x 1" P11 plates were used. Mill and oxide scale were removed with an angle grinder and cleaned with ethanol. When welding, first a layer of Ni-based alloy (Alloy 625 or Alloy 686) was deposited onto a P11 plate with 50% overlap between weld passes using the Cold Metal Process (CMT) with synergic line 1393 (Figure 41). Tri-mix welding gas was used to deposit this layer (Ar/30%He/0.5%CO₂). Similar welding parameters (Table 5) were used for both nickel-based alloys to clad the steel substrate. The plate was allowed to cool to room temperature, then the surface of the Ni-based layer was prepared using an angle grinding wheel to remove the surface of the weld beads. The grinding wheel was used to obtain a fresh, smooth surface of nickel-based alloy free of oxides and contaminants. For the buffer layer experiments, identical techniques were used to create a layer of overlapping weld passes of UTP A 80 Ni and Alloy 625 LNb over a layer of Alloy 625. The welding parameters to weld both buffer materials is summarized in Table 6.

Prior to the first weld with low alloy steel on the nickel-based layer, as well as between passes, ethanol cleaning was used to maintain clean weld surfaces. For each set of weld parameters, a two pass weld was made with 50% overlap (Figure 42), allowing the plate to cool below 100°F between welds. The shielding gas used was Ar/25%CO₂ mix and both CMT and pulsed gas metal arc welding (GMAW-P) were used with synergic lines 1362 and 378 respectively.



Figure 41: Nickel-based alloy layer over P11 substrate made with CMT to simulate the root pass in groove welding

Table 5: Welding Parameters for Nickel-Based Alloy "Root" Passes for Bead-on-Plate Layers (Alloy 625 and Alloy 686)

Process	СМТ
Shielding Gas	Tri-Mix
Flow Rate (ft^3/hr)	50
Wire Size (mm)	1.2
Travel Speed (mm/s)	5.9
Wire Feed Speed (mm/s)	188.4
Weave Frequency (Hz)	3.9
Amplitude (mm)	8
Synergic Line	ER NiCrFe-7A (1393)
Push Angle (deg.)	3
Tie-in Angle (deg.)	10
Dwell Time (s)	0

Material	UTP A 80 Ni	Alloy 625 LNb
Process	CMT	CMT
Shielding Gas	Tri-Mix	Tri-Mix
Flow Rate (ft^3/hr)	50	50
Wire Size (mm)	1.2	1
Travel Speed (mm/s)	5.5	6.35
Wire Feed Speed (mm/s)	52.92	95.25
Weave Frequency (Hz)	3.9	3.9
Amplitude (mm)	2.5	2.5
Synergic Line	ER NiCrMo-3 (1137)	ER NiCrMo-3 (974)
Push Angle (deg.)	3	3
Tie-in Angle (deg.)	10	10
Dwell Time (s)	0	0

Table 6: Welding Parameters for Nickel-Based Alloy "Buffer Layer" Passes for Bead-on-Plate Layers (UTP A 80 Ni and Alloy 625 LNb)



Figure 42: Two pass welding method for LAS over nickel-based layer showing example of single and double pass welds

Preliminary weld testing was performed to develop welding parameters for CMT and GMAW-P as well as study material combination compatibility and defects formation. These welding experiments were performed in the flat, horizontal position as a first step toward the final goal of groove welding on pipes. The summary of runs is provided in Table 18 in the Appendix, and the main parameters that were varied were wire feed speed (33.9-76.2 mm/s), weave amplitude (2-6 mm), welding process (CMT and GMAW-P), welding gas (Ar/25%CO₂ and Ar/30%He/0.5%CO₂), and nickel based substrate (Alloy 625 and Alloy 686). ER70S-6 consumable was also used for one of the weld runs and another run was attempted with no weave program.

Upon identifying suitable welding parameters from the preliminary weld trials, design of experiment studies were performed to further evaluate material combinations and welding processes and their subsequent effect on defect formation. For this portion of testing, buffer layers were also attempted. The summary of the differences of each DOE is specified in Table 7. The parameters used in the DOEs were developed with preliminary weld trials and ranges for each factor were determined (Table 8). Using JMP 11 statistical software and specifying a 3 factor/3 level experiment, a randomized run order was developed (Table 9). The factors being varied were weave amplitude, wire feed speed, and travel speed. In order to compare the DOEs accurately, the remaining variables in the experiments were held constant (Table 10). Upon completing DOE 1 and 2, the amount of welds for DOE 3 and 4 were reduced from 12 to 7 since some of the runs were slightly unstable. For DOE 5 and 6, the buffer layer experiments, the run number was reduced to 6 and both CMT and GMAW-P were used in each experiment. Due to difficult quantifying the outputs of the experiments, they were not able to be statistically evaluated with JMP 11, but instead were used for qualitative evaluation.

DOE	Welding Process	Nickel- based Alloy	Low Alloy Steel			
1	CMT	625				
2	CMT	686	-			
3	GMAW-P	625	-			
4	GMAW-P	686	ER100S-G			
5	CMT/GMAW-P	625/UTP A	LICIOUS G			
		80 Ni	_			
6	CMT/GMAW-P	625/Alloy				
		625 LNb				

Table 7: Differences between Design of Experiments

Table 8: Parameter Ranges for each Factor

Factor	Range	Units
Weave Amplitude	2-3	mm
Wire Feed	29.6-	mm/soo
Speed	46.6	IIIII/Sec
Travel	4.65-	mm/soo
Speed	6.35	mm/sec

Run	Weave Amplitude (mm)	Wire Feed Speed (mm/sec)	Travel Speed (mm/sec)
1	2	29.6	4.65
2	3	46.6	4.65
3	3	38.1	5.5
4	2.5	29.6	5.5
5	2.5	38.1	5.5
6	3	29.6	4.65
7	2	46.6	4.65
8	2	29.6	6.35
9	2.5	38.1	6.35
11	2	46.6	6.35
12	3	46.6	6.35
13	3	29.6	6.35

Table 9: Parameters for each DOE (Run 10 removed due to duplication)

Table 10: Unchanging Parameters for all DOEs

Welding Consumable	ER100S-G				
Shielding Cas	Ar/25%CO ₂ @				
Silleluling Gas	50 CFH				
Weave	39Hz				
Frequency	5.7 HZ				
	Mechanical				
Cleaning	grinding, wire				
Method	brush, and				
	ethanol cleaning				
Push Angle	3 degrees				
Tie-in Angle	10 degrees				



Figure 43: Torch orientation during welding: Tie-in angle (θ) with direction of travel out of the page. Push angle (ϕ) with direction of travel towards the right

4.5 Groove Welding Experiments

Upon completion of the bead-on-plate welding experiments, groove welding commenced. To better transition from bead-on-plate welding to pipe groove welding, flat plates were machined to have grooved edges to develop parameters and evaluate weldability of different materials. 4" x 6" x 1" 1018 steel plates were machined to have a groove on the two opposing 6" edges. The groove geometry is shown in Figure 44 and an actual groove is shown in Figure 45.



Figure 44: Groove geometry drawing



Figure 45: Grooved plates

Before placing the plates in the oven for preheating, the plates were tacked together at each end of the groove. Gas tungsten arc welding with ER70S-6 filler material was used for all tacking operations. Initially, a pre-bend of 10-15 was attempted to account for distortion of the plates during welding. However, due to difficulty of clamping a plate setup that had a pre-bend, the pre-bend was abandoned and instead, the plates were tacked together in the flat position. For all groove welding experiments, furnace preheating was used. The goal was to achieve a minimum of 50°C (122°F) preheat and interpass temperatures. The plates were placed in a furnace set to 177°C (350°F) and once they reached the furnace temperature, they were removed, fixed to the welding table in the Motoman welding cell with c-clamps and welding could commence once the temperature of the plates was slightly above 50°C. Temperature measurement was performed with a Fluke 51 thermometer with contact thermocouple attachment and all preheat and interpass temperatures were recorded for each weld. There was not a need to put the plates back into the furnace between each weld to maintain the desired interpass temperature due to heating from the previous weld.

For groove welding, special torch components were used to ensure that the torch did not contact or arc to the sidewalls. Narrow contact tips specially designed for narrow groove welding were used to allow for greater flexibility during weld programing. A comparison of the contact tips used during bead-on-plate welding and groove welding is shown in Figure 46. It should also be mentioned that Loctite SF 7900 ceramic spray was used to coat the contact tips to avoid arcing to the side walls during welding. A narrow gas nozzle was also used to allow for better gas coverage inside the narrow groove. A comparison between the gas nozzle used during bead-on-plate welding and groove welding is shown in Figure 47.



Figure 46: Comparison between contact tip used for bead-on-plate welding (top) and groove welding (bottom)



Figure 47: Comparison between gas nozzles used for bead-on-plate welding (top) and groove welding (bottom)

At this point in the study, it was determined that Alloy 686 would be a more viable match with ER100S-G than Alloy 625. In order to place the root pass of Alloy 686, welding parameters were developed. The parameters resulted in a root pass that did not have excessive bead height and attained complete fusion with the sidewalls and land,

however, complete penetration through the land was not attained. The welding parameters for the root pass are shown in Table 11 and a cross section of a lone root pass is shown in Figure 48. After welding the root pass with Alloy 686, ER100S-G fill passes were placed over the root until the top of the groove was reached. Wire brushing and ethanol cleaning were performed between welds. For the welding program, it was crucial that the torch be aligned with the center of the groove where the plates met. Practice runs with no arc were attempted to check that the electrode was centered and weaving equally to each side wall. Different methods of weaving were used to eliminate lack of fusion, ranging from impeding each side wall by $\frac{1}{2}$ wire diameter, to containing a $\frac{1}{2}$ wire diameter gap between the electrode and sidewall at the apex of the weave amplitude. This distance was termed the weave apex and is shown schematically in Figure 49. Another rule that was used was that the travel speed and weave frequency were set to achieve 5 total oscillations for every inch of weld. This ensured proper fusion with the sidewalls.

Parameter	Single-Sided Groove
Process	GMAW-P
Travel Speed (mm/s)	8.47
WFS (mm/s)	112.2
Weave Frequency (Hz)	5.0
Weave Amp (mm)	1.3
Dwell Time (s)	0.1
Arc Length Correction (%)	-10

Table 11: Alloy 686 Root Pass Parameters for Single-Sided Grooves



Figure 48: Alloy 686 root pass macrograph



Figure 49: Schematic of weave apex (distance between electrode tip and sidewall at furthest point of weave amplitude) for groove welding experiments

The welds made in the groove geometry required a dwell time to be added into the weave program. The dwell occurred at the toes of the weld and assisted wetting on each sidewall. For the bead-on-plate experiments, no weave dwell was used, but all welds in the groove geometry used a 0.1 second dwell. The difference between the two different weave types is shown in Figure 50.



Figure 50: Torch path for weave programs without dwell (left) and with dwell at the weld toes (right). Weave amplitude also specified as half of the total width and a single weave cycle is noted

4.6 Metallurgical Characterization

Cross sections were extracted from each weld using a bandsaw and cut down to a

smaller size using a TechCut 5TM precision sectioning machine to be prepared for

mounting. It should be noted that the surface of interest had a small sliver cut off using

the TechCut 5^{TM} to be able to begin polishing with a smoother surface rather than the surface left from the bandsaw cut. The cross sections were mounted in Bakelite using a hot compression mounting system (Leco PR-32) and were ground using 400 and 600 grit SiC abrasive paper disks with water cooling. Between grits, ethanol rinsing was used to clean the samples and drying was conducted with a heat gun. Polishing proceeded using 9 μ m, 3 μ m, and 1 μ m diamond compound with an oil based colloidal suspension. Between polishing steps, the samples were placed in an ultrasonic bath filled with ethanol to ensure that cross contamination between polishing pads did not occur. After being immersed in the ultrasonic bath for 2-3 minutes, the samples were rinsed with ethanol and drying was carried out with a heat gun. Finally, a 0.05 µm colloidal silica compound was used for a fine polish. Care was taken not to allow any of the colloidal silica to dry on the surface of the sample. If drying of the colloidal occurred, the substance would crystallize on the surface and be very difficult to remove, often requiring a soap and water rinse or for the sample to be repolished on this step completely. The final polishing step was performed to better observe surface relief with optical microscopy due to the difficulty of etching the weld metal.

To etch the dissimilar welds, initially, a 10% Oxalic acid electrolytic etch (5 volts, 1 amp, 30 seconds) was used to reveal the microstructure in the heavily diluted weld metal, but this process destroys the nickel-based alloy layer. Since an objective of this study is to learn about the defect formation at the interface of the weld metal and substrate, a light 10% chromic acid electrolytic etch (3 volts, 0.25 amps, 1-3 seconds) was used instead in order to reveal the microstructure in the nickel-based substrate

without destroying evidence of the interfacial defects. For the groove welding experiments, 2% Nital etchant was used with an immersion and agitation technique where the sample was immersed completely in the solution and moved in a small circle for 5-8 seconds. For multiple etches used on the same sample, Nital was normally performed before others to reveal the base material and HAZ microstructure and protect it when chromic etchant was used after. A list of etchants is provided in Table 12.

Etchant	Time	Electrolytic parameters	Reveals		
10% Oxalic Acid	30 sec	5 V, 1 A	Low alloy steel heavily diluted with nickel-based alloy		
10% Chromic Acid	1-3 sec	3 V, 0.25 A	Nickel-based alloy		
2% Nital	5-8 sec	NA	Steel base material, undiluted low alloy steel		

Table 12: Etchants used over course of study

An Olympus GX51 microscope was used to obtain micrographs and was also used to measure the weld areas to measure dilution geometrically (Equation 1). For overlapping passes, the dilution of each pass was calculated with the method shown in Figure 51 with Equation 7. For optical microscopy, differential interference contrast (DIC) was also used to accentuate surface relief to analyze samples that had not been etched. The DIC filter would often reveal grain boundaries on un-etched metal if a high quality polish was performed. For energy dispersive spectroscopy (EDS), a Quanta 200 scanning electron microscope was used and images were also gathered in order to examine the internal structure of defects. EDS scans were performed across the interface of the substrate and weld metal to ascertain dilution as a function of position. The scans were performed with the voltage set between 10-15 keV and the spot size (beam current) set to 4 and adjusted accordingly to obtain the best possible image. The elements included in the EDS line scans were Fe, Ni, Cr, Mo, Mn, Nb, and W.



Figure 51: Areas of measurement for overlapping welds to measure dilution of second pass

Dilution (Second pass) =
$$\frac{2A + 2C * (\frac{1A}{1A + 1B + 2C})}{2A + 2B + 2C} * 100 (\%)$$

Equation 7: Dilution calculation for second pass of overlapping welds

Chapter 5: Results

5.1 Computational Modeling

The following section presents the results of the computational modeling with the thermodynamic simulation tool, Thermo-Calc[™]. First, the simulations were completed to assess the compatibility of the Ni-based root pass and low alloy steel fill passes with no buffer layer. The root pass materials were Alloy 625 and 686 and the fill pass materials were ER100S-G and ER70S-6. Next, other materials were chosen as potential buffer layers to isolate the root and fill passes. The buffer layer materials were a nearly pure nickel alloy, UTP A 80 Ni, and an Alloy 625 variant with no niobium addition, Alloy 625 LNb.

5.1.1 Simulations with No Buffer Layer

First, two low alloy steel materials, ER100S-G and ER70S-6 were compared. Each LAS was paired with both Alloy 625 and Alloy 686 to ascertain solidification temperature ranges. The pseudo-binary phase diagrams for ER100S-G combinations are shown in Figure 52 and Figure 53 and ER70S-6 combinations shown in Figure 54 and Figure 55. For the LAS combination with Alloy 625, there was a larger STR across all dilutions with ER70S-6 than ER100S-6. However, there was a smaller STR with the ER70S-6/Alloy 686 combination than the ER100S-G/Alloy 686 combination across all dilutions. A summary of all STR values at each dilution is provided in Table 13. The Thermo-Calc[™] simulations also revealed that the combination of ER100S-G/Alloy 686 (Figure 52) and ER70S-6/Alloy 686 (Figure 54) would produce more favorable combinations than ER100S-G/Alloy 625 (Figure 53) and ER70S-6/Alloy 625 (Figure 55). The reason low alloy steel is more compatible with Alloy 686 than Alloy 625 is that there is a closer liquidus/solidus match with Alloy 686 and overall has a small solidification temperature range (STR) across all dilutions. The combinations of ER100S-G/Alloy 625 and ER70S-6/Alloy 625 were found to be potentially susceptible to solidification cracking due to extremely large solidification temperature ranges across all dilutions, as well as the presence of a niobium carbide rich eutectic that segregates to the grain boundaries and decreases the freezing temperature of the grain boundary liquid.



Figure 52: Pseudo-binary phase diagram between ER100S-G and Alloy 686. Databases: TCFE5 and TTNI7



Figure 53: Pseudo-binary phase diagram between ER100S-G and Alloy 625. FCC2 represents Nb-rich eutectic. Databases: TCFE5 and TTNI7



Figure 54: Pseudo-binary phase diagram between ER70S-6 and Alloy 686. Databases: TCFE5 and TTNI7



Figure 55: Pseudo-binary phase diagram between ER70S-6 and Alloy 625. FCC2 represents Nb-rich eutectic. Databases: TCFE5 and TTNI7

Combination				Dilut	ion (%	6 Ni-b	ased A	Alloy)			
Combination	0	10	20	30	40	50	60	70	80	90	100
Alloy 686 in ER100S-G	41	79	104	108	97	84	105	108	109	117	119
Alloy 625 in ER100S-G	41	112	149	195	274	315	361	338	337	313	219
Alloy 686 in ER70S-6	44	62	90	95	86	73	93	97	106	107	119
Alloy 625 in ER70S-6	44	109	157	219	320	356	379	365	342	316	219

Table 13: Solidification Temperature Ranges for LAS/Ni-based Alloy Combinations

For partition coefficient calculations, only the combinations involving ER100S-G and each Ni-based root pass material were tested. The partition coefficient values for the ER100S-G/Alloy 625 combination (Figure 56) showed that niobium segregates to the liquid, or grain boundaries, at low dilutions. For k<1, a smaller k value indicates a stronger segregation effect of the particular element to be rejected by the solid and enriching the liquid. No such partitioning effect was seen for any other element in either combination. For the ER100S-G/Alloy 686 combination, there was a narrow solidification temperature range across all dilutions and no significant partitioning effect from tungsten or molybdenum.



Figure 56: Partition coefficients of Nb and Mo in the ER100S-G/Alloy 625 combination (solid line) and W and Mo in the ER100S-G/Alloy 686 combination (dotted line)

Finally, the temperature data for the pure consumable compositions is compared in Table 14. The Thermo-Calc[™] results correlated with the welding results, proving that ER100S-G and Alloy 686 are more compatible than ER100S-G and Alloy 625 when solidification cracking and shrinkage porosity formation is considered.

Material	Liquidus (°C)	Solidus (°C)	Solidification Temperature Range
Alloy 625	1369	1150	219
Alloy 686	1368	1249	119
ER100S-G	1516	1473	43
ER70S-6	1523	1479	44

Table 14: Liquidus and Solidus Temperatures predicted by Thermo-CalcTM

5.1.2 Buffer Layer Simulations

For this study, it was also hypothesized that a buffer layer could be used to isolate the nickel-based root pass from the low alloy steel fill passes. There were two potential buffer layers that were selected for analysis: UTP A 80 Ni and Alloy 625 LNb. Since these materials were being tested as buffer layers, their compatibility with both the root pass material and fill pass material needed to be evaluated. For the case of the buffer layer simulations, Alloy 625 was used as the root pass material and ER100S-G as the fill pass material.

First examining the UTP A 80 Ni buffer layer, Thermo-Calc[™] showed good compatibility with Alloy 625 at low dilutions (Figure 57). At higher dilution levels, the solidification temperature range increased. It should be mentioned that the phase FCC_L12 represents austenite. With the UTP A 80 Ni/ER100S-G combination, an early version of the Thermo-Calc[™] steels database (TCFE5) had initially predicted good compatibility between the materials with a narrow STR across all dilutions. However, when using an updated database (TCFE8), a large dip in solidus temperatures occurred at low dilutions, which was in the range of actual welding (Figure 58). Of special interest when trying to eliminate shrinkage porosity defects, there was close agreement between the liquidus/solidus temperatures of both pure ER100S-G and pure UTP A 80 Ni.



Figure 57: Pseudo-binary phase diagram between UTP A 80 Ni and Alloy 625. Database: TCNI8



Figure 58: Pseudo-binary phase diagram between ER100S-G and UTP A 80 Ni. Databases: TCFE5 and TCNI8

Next, the other buffer layer material, Alloy 625 LNb, was also tested for compatibility with Alloy 625 and ER100S-G. For compatibility with Alloy 625, a slightly large STR was predicted at low dilutions (Figure 59). For the Alloy 625 LNb/ER100S-G combination (Figure 60), there was a consistently narrow STR across all dilutions and there was also a good match between the liquidus/solidus temperatures of the two materials. A purely austenitic structure is predicted at all dilutions with no secondary phase predicted. This combination also shows potential resistance to shrinkage porosity defects due to the close match of liquidus/solidus temperatures. To better compare the STRs of all combinations, Table 15 provides STR predictions across all dilutions for the buffer layer combinations. For the pure buffer materials, the liquidus and solidus temperatures, along with the solidification temperature ranges are summarized in Table 16.



Figure 59: Pseudo-binary phase diagram between Alloy 625LNb and Alloy 625. Database: TCNI8



Figure 60: Pseudo-binary phase diagram between ER100S-G and Alloy 625 LNb. Databases: TCFE5 and TCNI8

Combination	Dilution (% Substrate Material)										
	0	10	20	30	40	50	60	70	80	90	100
Alloy 625 in UTP A 80 Ni	46	50	55	66	82	105	136	168	183	200	219
UTP A 80 Ni in ER100S-G	43	148	243	280	276	156	127	99	77	62	46
Alloy 625 in Alloy 625 LNb	79	111	141	169	196	217	225	226	224	221	219
Alloy 625 LNb in ER100S-G	43	67	87	95	92	76	69	68	65	67	79

Table 15: Solidification Temperature Ranges for Buffer Layer Combinations

Table 16: Liquidus and Solidus Temperatures of Buffer Materials predicted by Thermo-CalcTM

Material	Liquidus (°C)	Solidus (°C)	Solidification Temperature Range
UTP A 80 Ni	1421	1375	46
Alloy 625 LNb	1391	1312	79

5.2 Bead-on-Plate Welding Experiments

In this section, all bead-on-plate welding experiments that were performed in the flat, horizontal position will be presented. It quickly became apparent that there were multiple defects that could occur with this dissimilar combination. As such, these results will be presented in categories of each defect: solidification cracking, liquation cracking, and shrinkage porosity. The bead-on-plate experiments were comprised of a set of preliminary weld trials to narrow the parameters down to ranges that could be used in
welding design of experiments (DOEs). All DOEs used identical welding parameters, but the main difference between them was the materials or the welding process being used. There were 6 total DOEs attempted for this phase of the project.

5.2.1 Solidification Cracking

After completing the preliminary weld trials and metallographic preparation, it was found that six of the preliminary weld trials contained solidification cracking for the ER100S-G/Alloy 625 combination (Table 18, Appendix). The majority of the cracks propagated from the interface between the nickel-based substrate and the weld metal and extended into the weld metal (Figure 61). Solidification cracks also propagated off of weld swirls. The solidification cracking occurred over a large range of dilutions (~15-36%) and heat inputs (~250-550 J/mm), but a correlation was found between solidification cracking and weave amplitude. Solidification cracking was prevalent in welding trials that used a large weave amplitude (\geq 4 mm), but was avoided in trials using a small weave amplitude (\leq 3 mm).



Figure 61: Solidification crack in Weld Trial 1 of preliminary weld trials propagating from dissimilar interface. Note the transition zone between materials

For DOE 1 (Table 19, Appendix), which used Alloy 625 as the substrate and CMT as the welding process, four welds experienced solidification cracking, which correlated to the Runs with the largest weave amplitude (3 mm). An example of a solidification crack propagating off of a weld swirl is shown in Figure 62. Dilution is dictated by heat input, and heat input is affected by wire feed speed, oscillating travel speed, and weave amplitude. Thus, welds made with low wire feed speed, high oscillating travel speed, and large weave amplitude create low heat input and low dilution welds. It was these welding conditions that produced solidification cracking (Figure 63). Additionally, solidification cracking was avoided at small weave amplitudes (≤ 2.5 mm) and occurred at the largest weave amplitude (3 mm). The power ratio of each weld was also calculated for each weld and compared to dilution (Figure 64). When calculating power ratio, the oscillating travel speed was used in order to include weave amplitude in

the parameter. High power ratios in a large range of dilutions contained solidification cracking.



Figure 62: Solidification crack propagating from weld swirl in Run 6, DOE 1



Figure 63: Dilution vs. Heat Input (calculated with oscillating travel speed) of DOE 1 for three different weave amplitudes. "X" denotes Run that contained solidification cracking



Figure 64: Power ratio vs. Dilution for DOE 1 shows that high power ratios experience cracking

DOE 3 (Table 21) employed a higher heat input process (GMAW-P) to achieve higher dilution and alleviate solidification cracking. While no weld in this DOE contained solidification cracking, only the small weave amplitude Runs were attempted (≤ 2.5 mm) due to the results of DOE 1. Also, GMAW-P yielded consistently higher heat inputs than with CMT, however, did not consistently yield higher dilutions. Finally, it should be noted that solidification cracking only occurred in the LAS/Alloy 625 combination and was not encountered in the LAS/Alloy 686 combination. Thus, DOE 2 (Table 20), DOE 4 (Table 22), DOE 5 (Table 23), and DOE 6 (Table 24) were not considered in the results or discussion of solidification cracking.

5.2.2 Liquation Cracking

In DOE 1 (Table 19), in addition to solidification cracking, 3 Runs contained liquation cracking in the first pass that occurred during welding of the second pass. The Runs containing liquation cracking all exceeded 32% dilution, and lower dilutions did not contain liquation cracking. The liquation cracks all occurred in the same general location, roughly midway between the substrate and the surface of the weld beads. CMT was used in DOE 1 and liquation cracking occurred with the low heat input and low dilution process, however, GMAW-P was used to alleviate solidification cracking in DOE 3 (Table 21). As expected, liquation cracking was found in 5 of 7 weld Runs. An example of liquation cracking in the first pass of two pass welds is shown in Figure 65. In an attempt to quantify the defect formation, a liquation cracking factor was assigned to each Run according to defect size and quantity (1=small, few; 3=large, many). However, there was not a clear relationship with dilution or heat input when compared to liquation cracking factor (Figure 66). Cracking was seen in welds with dilutions ranging from 25-40%, however, the highest dilution run contained no liquation cracks. Since only one high dilution trial was free of cracking, it was likely sectioned in a position of the weld where no cracking occurred. It should be noted that liquation cracking only occurred in the LAS/Alloy 625 combination and was not encountered in the any other combination. DOE 2, 4, 5, and 6 were all free of liquation cracking.



Figure 65: Liquation cracking in Run 5, DOE 3



Figure 66: Dilution vs. Liquation Cracking Factor for DOE 3

5.2.3 Shrinkage Porosity

Small voids, which are referred to in this study as shrinkage porosity, were present in essentially every weld Run. The defects all developed in the same general location in each weld run; adjacent to the fusion boundary in the unmixed zone of the nickel-based substrate, in the transition zone, and in weld swirls and islands. Shrinkage porosity was not anticipated in this study, so the preliminary welding trials were not examined for this defect. It was not until the Design of Experiment studies commenced that the defect was noticed. During DOE 1 (Table 19), welding ER100S-G over Alloy 625 using the CMT process, there was a large range of severity encountered across weld runs. Some runs contained small instances of shrinkage porosity, while one instance resulted in a continuous layer of shrinkage porosity to form between the substrate and weld metal, resembling the beginning of decohesion of the weld metal and substrate (Figure 67).



Figure 67: DOE 1 Run 6 interface between ER100S-G weld metal and Alloy 625 substrate. Continuous shrinkage porosity formation as well as evidence of an unmixed zone

DOE 2 (Table 20) was run with identical parameters as DOE 1, but used an Alloy 686 substrate. The resulting welds contained fewer and smaller defects than when welded over Alloy 625. Of the first two DOEs comparing nickel-base substrate, ER100S-G was more compatible with Alloy 686 (DOE 2) than with Alloy 625 (DOE 1) due to smaller and fewer shrinkage porosity defects encountered during cross-section examination. It should be reiterated that the analysis of shrinkage porosity was qualitative due to the random nature of the defects and only one cross section of each weld run was analyzed. Upon creation of a diagram to visualize the dilution/temperature data of welds of ER100S-G over Alloy 625 (Figure 68) and Alloy 686 (Figure 69), it was hypothesized that a higher dilution may decrease susceptibility to shrinkage porosity. The higher dilution weld metal would more closely match the liquidus/solidus temperatures of the Ni-based substrates. The dilution was calculated using compositional information gathered in an EDS traverse that began in the Ni-based substrate and ended in the ER100S-G weld metal. It should be mentioned that the welds analyzed had differing dilution levels.



Figure 68: EDS/Thermo-CalcTM Interface Model to visualize interfacial transition of LAS weld over Alloy 625 (bulk dilution = 17%). Theoretical temperature gradient to promote uniform solidification represented by red line



Figure 69: EDS/Thermo-CalcTM Interface Model to visualize interfacial transition of LAS weld over Alloy 686 (bulk dilution = 41%). Theoretical temperature gradient to promote uniform solidification represented by red line

DOE 3 (Table 21) and 4 (Table 22) were run with GMAW-P and matched DOE 1 and 2 in every other parameter respectively. It should be noted that the number of weld runs was decreased in DOE 3 and 4 to 7 weld runs for each DOE. DOE 1 and 2 examined 12 weld runs each. The sample size was decreased due to the high weave amplitude runs experiencing solidification cracking in DOE 1, but was unrelated to shrinkage porosity. The higher heat input process of GMAW-P did not consistently result in higher dilutions than CMT, but the majority of defects in DOE 3 and 4 were small and there were no defects that resulted in a continuous layer of shrinkage porosity. For DOE 4, an attempt was made to relate shrinkage porosity to dilution of each weld pass by ranking the defects according to size (shrinkage porosity factor). However, the results were inconclusive and only a weak trend was realized (Figure 70) indicating that dilution levels between 25 and 35% were most susceptible to the defect.



Figure 70: Shrinkage Porosity Factor vs. Dilution for DOE 4

When using GMAW-P, Alloy 625 substrate yielded slightly larger and more numerous defects than the substrate of Alloy 686, but the difference between the two was not as drastic as when using CMT. It was concluded that Alloy 686 was more resistant to shrinkage porosity defects due to the slightly smaller size and number of defects than Alloy 625 when acting as a substrate for ER100S-G using GMAW-P. Again, these results are quantitative based on the analysis of single cross sections. Examples of shrinkage porosity defects at the interface of the ER100S-G weld metal and the Ni-based substrate are shown in Figure 71.



Figure 71: Shrinkage Porosity defects at interface of ER100S-G weld metal and Alloy 625 substrate, (A) Run 2, DOE 1, (B) Run 6, DOE 1

For the buffer layer experiments, DOE 5 and 6, shrinkage porosity was also encountered. For DOE 5 (Table 23), shrinkage porosity was seen at the interface of Alloy 625 and UTP A 80 Ni, but none was seen at the interface of UTP A 80 Ni and ER100S-G. This was the first instance of shrinkage porosity avoidance at an interface involving ER100S-G. In DOE 6 (Table 24), shrinkage porosity was found at both the interface between Alloy 625 and Alloy 625 LNb and the interface between Alloy 625 LNb and ER100S-G. The cross sections analyzed indicated that the GMAW-P process resulted in more complete mixing of the ER100S-G weld metal that was diluted by the Alloy 625 LNb substrate. Due to better mixing, there were less weld swirls and islands where shrinkage porosity defects can form. Examples of shrinkage porosity defects when using a UTP A 80 Ni buffer layer are shown in Figure 72.



Figure 72: Shrinkage porosity defects at interface of Alloy 625 and UTP A 80 Ni buffer layer

5.2.4 Summary of Defects

This section will summarize the defects seen in each stage of experimentation and why each experiment was chosen. From the preliminary weld trials, solidification cracking was found to be related to weave amplitude, but no relation with dilution was found. When the DOE phase of the experiments began, smaller weave amplitudes were selected to decrease the internal stresses that are experienced in the welds. DOE 1 (Table 19) was completed to examine the solidification cracking response of ER100S-G welded directly over Alloy 625 at a refined parameter range. Solidification cracking was found to be related to weave amplitude, with small amplitude welds containing no cracks. Solidification cracking was also found to occur when low heat input welds were performed that resulted in low dilution, which was contrary to the initial hypothesis. Since liquation cracking and shrinkage porosity defects were also found, DOE 2 (Table 20) was performed to try to simultaneously eliminate solidification cracking, liquation cracking, and shrinkage porosity. Alloy 686 as a substrate would eliminate the NbC rich eutectic that was hypothesized to be causing solidification cracking and the solidification temperature range was narrow, which would decrease the size of the unmixed zone in the substrate material. The hypothesis was partially correct and only shrinkage porosity was encountered. The shrinkage defects also were noticeably smaller and fewer compared to DOE 1.

For DOE 3 (Table 21), a higher heat input welding process, GMAW-P, was used to weld ER100S-G directly over Alloy 625. Higher dilution welds were thought to be more resistant to solidification cracking and shrinkage porosity, however, they could be more susceptible to liquation cracking. That prediction was correct, with 5 out of 7 welds containing liquation cracking. There was no clear relation with dilution level and liquation cracking in this experiment. Qualitatively, the shrinkage porosity defects were smaller than in DOE 1 when using CMT and solidification cracking was not encountered. In order to eliminate liquation cracking, the substrate was changed to Alloy 686 for DOE 4 (Table 22). The higher heat input of GMAW-P was also thought to decrease the shrinkage porosity defects. No liquation cracking occurred in this DOE, but shrinkage porosity defects still occurred at the interface of the Alloy 686 and ER100S-G. Unfortunately, attempts to quantify the defect were met with limited success.

Next, buffer layer materials were tested to isolate the Ni-based substrate from the ER100S-G weld passes. UTP A 80 Ni was used to weld a layer of material over a layer Alloy 625 using both CMT and GMAW-P in DOE 5 (Table 23). This approach was the only combination tested that resulted in no solidification cracking, liquation cracking, or shrinkage porosity at the interface of ER100S-G and the substrate material. However, shrinkage porosity defects were found adjacent to the interface of Alloy 625 and UTP A 80 Ni. Due to the absence of shrinkage porosity defects at the LAS interface, this buffer layer was chosen for groove weld experimentation. As a final attempt to find a more suitable combination, Alloy 625 LNb was attempted as a buffer layer between Alloy 625 and ER100S-G weld passes in DOE 6 (Table 24). Cracking, both solidification and liquation, was not found, but shrinkage porosity defects were found at both interfaces of the buffer layer. It was noted that the shrinkage porosity defects in the GMAW-P welds were less severe as the defects in the CMT welds, which was attributed to better mixing in the welds of GMAW-P with less instances of weld swirls and islands. Alloy 625 LNb was not recommended for groove welding experiments. A complete summary of the DOE portion of the study is shown in Table 17.

DOE	Welding Process	Ni-based Alloy	LAS Alloy	Number of Welded Samples	Travel Speed Range (mm/s)	WFS Range (mm/s)	Weave Frequency Range (Hz)	Weave Amplitude Range (mm)	Defects
1	СМТ	625	ER100S- G	12	4.65- 6.35	29.6- 46.6	3.9	2-3	1, 2, and 3
2	CMT	686		12				2-3	3
3	GMAW- P	625		7				2-2.5	2 and 3
4	GMAW- P	686		7				2-2.5	3
5	Both	625/UTP A 80 Ni		6				2-2.5	3 (at root/buffer interface only)
6	Both	625/625 LNb		6				2-2.5	3 (both interfaces)

Table 17: Summary of Bead-on-Plate Welding Experiments

1 = Solidification Cracking

2 = Liquation Cracking

3 = Shrinkage Porosity

5.3 Groove Welding Experiments

This section presents the results on the welding experiments that were carried out in narrow grooves in the flat, horizontal position. After establishing stable welding parameters for ER100S-G fill passes over the Alloy 686 root pass, 8 groove welding trials were completed. The purpose of these experiments was to eliminate all defects from the weldment (solidification cracking, shrinkage porosity, lack of fusion). The summary of each groove weld trial with the parameters and relevant defects is presented in Table 25. The parameters, defects, and relevant comments about each pass in each groove weld are shown in Table 26 and Table 27. The heat input for each pass was calculated by gathering the instantaneous power during welding, however, dilution could not be calculated geometrically due to the complex shapes of the beads and interactions with neighboring materials.

It was initially thought that when setting the weave program amplitude, there should be 1 wire diameter distance between the tip of the wire and the sidewall at the apex of the weave pattern. This method was used for Groove 1-3. Groove 1 resulted in aesthetically pleasing weld beads, however, the weave program was wrongfully set to weave by "time" rather than "frequency", which resulted in an extremely slow weave. Centerline cracks were found during welding in Pass 1 and 2. The presence of these cracks were verified when a cross section was extracted. Lack of fusion (LOF) defects were also present at both triple-points of each weld, meaning the area of the weld where the weld contacts both the previous weld and the sidewall. Finally, no shrinkage porosity defects were found at any interface, however, it was difficult to identify the defect due to pitting in the root pass during polishing. The microstructure of the first pass was primarily austenitic, while the second pass was a mixture of austenite and martensite. Passes 3-6 likely consist of ferrite and martensite. Cross sections of Groove 1 are shown in Figure 73. A complete description of each pass is provided in Table 26.

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Figure 73: Groove 1 macrograph with solidification cracking and LOF defects

For Groove 2, the weave program was corrected to weld by "frequency" rather than "time". The wire feed speed was also slightly increased from Groove 1. No solidification cracks were found in this trial and no shrinkage porosity was found at any interface. However, all weld passes contained varying sizes of LOF at the triple-points. The microstructure of the first pass was primarily austenitic, while the second pass was a mixture of austenite and martensite, determined by examination at high magnification. Passes 3-6 likely consist of ferrite and martensite. Cross sections of Groove 2 are shown in Figure 74. A complete description of each pass is provided in Table 26.



Figure 74: Groove 2 macrograph with LOF defects and a micrograph of the microstructure of Pass 2 indicating a mixture of austenite and martensite from the partially dendritic structure with needle-like formations

In order to eliminate the LOF defects of Groove 2, the wire feed speed was increased for all fill passes in Groove 3. The travel speed was also increased to attempt to have the wire enter the weld pool at the leading edge of the weld pool rather than near the middle, leading to better fusion at triple-points. Solidification cracking was found in Pass 1 and 2. Two LOF defects were encountered at the triple-points of Pass 4 due to an error during welding where the wire began depositing up the sidewall. Other than this welding mistake, no other LOF defects were found. One small shrinkage porosity defect was encountered at the interface of the root pass and Pass 1 in the unmixed/transition zone. The same microstructure trend occurred as in Groove 1 and 2. Cross sections of Groove 3 are shown in Figure 75. A complete description of each pass is provided in Table 26.



Figure 75: Groove 3 macrograph with solidification cracking, LOF, and a shrinkage porosity defect

For Groove 4 and 5, a different weave apex was attempted to alleviate the LOF defects that were occurring at the triple-points. The weave amplitude for each pass was set to impinge on each sidewall by $\frac{1}{2}$ of a wire diameter to ensure fusion with the sidewalls. Also, since the microstructure of passes 3-6 on Grooves 1-3 were nearly identical and no cracking or shrinkage porosity was encountered in those passes, Grooves 4-7 were welded with only 4 fill passes. Groove 4 closely matched the parameters of Groove 2, but the weave apex was changed and the weave frequency was halved to achieve 10 "bumps" per inch of travel. Groove 4 experienced undercutting during most of the fill passes due to the weave apex impinging the sidewall, and there were a two small, and one large LOF. A small centerline solidification crack (<100 µm length) was found in Pass 1 and no cracking was found in Pass 2. At the interface of the root pass and Pass 1, a large shrinkage porosity defect was found in a filler deficient swirl. The same microstructure trend occurred as in Grooves 1-3. Cross sections of Groove 4 are shown in Figure 76. A complete description of each pass is provided in Table 26.



Figure 76: Groove 4 macrograph with a small solidification crack, LOF defects, and a shrinkage porosity defect, with clear lack of mixing near the interface

For Groove 5, the travel speed was increased to see the effect on solidification centerline cracking. The wire feed speed was also increased. The weave frequency on the remainder of welding trials was 2 Hz. The weave apex was identical to that used in Groove 4 and impinged on each sidewall ½ of the wire diameter. Two small centerline solidification cracks were encountered in Pass 2. Undercutting was encountered in many of the fill passes and two large LOF defects occurred at triple-points. However, no shrinkage porosity was found at any interface. The same microstructure trend occurred as

in Grooves 1-4. Cross sections of Groove 5 are shown in Figure 77. A complete description of each pass is provided in Table 27.



Figure 77: Groove 5 macrograph with small solidification cracks and LOF defects

Groove 6 used a different weave apex than what was used in Grooves 4 and 5. To reduce undercutting, the apex was set to reach ½ wire diameter away from each sidewall. The push angle of the welding torch was also increased from 3° to 5° to further push the heat towards the leading edge of the welding pool and eliminate LOF defects. All other parameters matched Groove 4. An extremely small solidification crack was found in Pass 2. A shrinkage porosity defect was also found at the interface of the root pass and Pass 1

in the unmixed zone of the Alloy 686. Finally, many LOF defects were found at the triple-points in the fill passes. The same microstructure trend occurred as in Grooves 1-5. Cross sections of Groove 6 are shown in Figure 78. A complete description of each pass is provided in Table 27.



Figure 78: Groove 6 macrograph with a micro solidification crack, a shrinkage porosity defect, and LOF defects

Groove 7 also used the weave apex used in Groove 6 to reduce the undercutting defects during welding. In addition, the increased push angle was used, however, the WFS was reduced in an attempt to decrease the bead height. No solidification cracking

was found in any pass. A single shrinkage porosity defect was found in an island in Pass 1 near the interface of the root pass and Pass 1. The island primarily consisted of Alloy 686. LOF was encountered at the triple-points of fill passes. The same microstructure trend occurred as in Grooves 1-6. Cross sections of Groove 7 are shown in Figure 79. A complete description of each pass is provided in Table 27.



Figure 79: Groove 7 macrograph with a shrinkage porosity defect in an island and LOF defects. Lack of mixing at interface of Root pass and Pass 1 also shown

The final groove welding trial, Groove 8, used a weave apex that had the wire electrode touching each sidewall, but not actually impinging into each sidewall. All other

welding parameters matched Groove 6. No solidification cracking was found in any pass. One shrinkage porosity defect developed at the interface of the root pass and Pass 1 in the unmixed zone of the root pass. Two small lack of fusion defects were found in the fill passes at the triple-points. This trial was the best parameter set that was attempted in groove welding due to having no solidification cracking and limited LOF. The same microstructure trend occurred as in Grooves 1-7. Cross sections of Groove 8 are shown in Figure 80. A complete description of each pass is provided in Table 27.



Figure 80: Groove 8 macrograph with a shrinkage porosity defect and LOF defects

Chapter 6: Discussion

6.1 Computational Modeling

The overarching goal of this study was to ascertain if low alloy steel can be welded as fill passes over an Alloy 625 root pass in pipelines to be used for subsea presalt applications. The sections that follow offer a discussion of the results with Thermo-CalcTM modeling with and without a buffer layer.

6.1.1 Simulations with No Buffer Layer

When down selecting low alloy steel consumables, ER100S-G and ER70S-6 were first analyzed for nickel-based alloy compatibility with Thermo-Calc[™]. ER70S-6 did not meet the 550 MPa minimum yield strength requirement, but was included in this stage of the analysis since a spool was available for welding. The results of each low alloy steel suggested that ER100S-G had better viability when mixed with Ni-based Alloy 625 due to a smaller solidification temperature range across all dilutions. As such, ER100S-G was selected as the primary low alloy steel consumable for fill pass testing. It was decided that ER70S-6 would undergo limited weld testing to check the results provided by Thermo-Calc[™].

For the Ni-based substrate analysis, it was realized that the combination of ER100S-G and Alloy 625 would suffer from an extremely large solidification

temperature range (STR) and also the formation of Nb-C eutectic that further lowers the solidus temperature. Because the combination could be prone to solidification cracking, Alloy 686 was also considered as a potential replacement for the Alloy 625 root pass. Alloy 686 was selected for this study due to the absence of niobium in the material. The hypothesis for the omission of Nb from the system was that the STR would shrink and there would be no significant partitioning effect from any of the alloying elements. The results of the initial Thermo-Calc[™] models indicated that welding ER100S-G directly over Alloy 625 (Figure 53) would be a difficult task because solidification cracking seemed probable. The large STR and partitioning effects of Nb to the grain boundaries would render the welds susceptible. However, the Thermo-Calc[™] models of ER100S-G and Alloy 686 (Figure 52) predicted a more compatible combination, with the smaller STR and absence of Nb, which eliminates the presence of the low melting point Nbcontaining eutectic constituent. It was decided that weld testing would be performed for both nickel-based alloys due to the initial goal of welding ER100S-G over Alloy 625. ER70S-6 would undergo a single weld test to validate the Thermo-Calc[™] results, but further testing was not expected. It should be mentioned that solidification cracking was the main concern at this stage of the study and liquation cracking and shrinkage porosity defects had not been considered.

6.1.2 Buffer Layer Simulations

Two buffer layer materials, UTP A 80 Ni and Alloy 625LNb were tested to evaluate their compatibility with the root and fill pass materials. UTP A 80 Ni consists of 96.5% Ni and 3% Ti with other small alloying additions and has a narrow STR. The other reason why it was chosen for analysis was an attempt to reduce shrinkage porosity defect formation. The material would act as a transition between Alloy 625 (Figure 57) and ER100S-G (Figure 58) because the liquidus/solidus temperatures of UTP A 80 Ni fell between the respective temperatures of the root and fill materials. Ideally, the unmixed zone formation in the substrate at each interface would be minimized, leading to less shrinkage porosity defects. Another advantage of this buffer layer is the absence of niobium, which Thermo-CalcTM predicted to have a detrimental effect on STR and subsequent solidification cracking. A disadvantage of this material was that Thermo-CalcTM predicted a large STR at low dilutions with Alloy 625, which occurs in the dilution range of typical bead-on-plate welds. This may potentially have a negative effect on solidification cracking response at low dilutions, but due to the omission of Nb from the system, UTP A 80 Ni was chosen for bead-on-plate welding trials.

For Alloy 625 LNb, good compatibility was predicted with Alloy 625 (Figure 59) and even better compatibility with ER100S-G (Figure 60). The narrow STR across all dilutions and closely matching liquidus/solidus temperatures made the Alloy 625LNb/ER100S-G combination an ideal match to avoid solidification cracking as well as shrinkage porosity. Alloy 625 LNb was also selected to continue to bead-on-plate welding trials.

6.2 Bead-on-Plate Welding Experiments

The subsequent sub-sections discuss the results of the flat, horizontal bead-onplate welding experiments. These experiments include preliminary weld trials as well as 6 DOEs. These experiments were performed with the mindset that if a combination was not defect free in the bead-on-plate configuration, it would likely contain defects in narrow groove welding and should be eliminated from the analysis.

6.2.1 Solidification Cracking

The initial goal of the welding trials was to evaluate solidification cracking susceptibility and determine if there was an operational window of dilution where cracking could be avoided. ER70S-6 was quickly eliminated from the analysis due to extensive solidification cracking when tested in the preliminary weld trials (Trial 4, Table 18). ER100S-G proved more resistant to the defect than ER70S-6, so all subsequent experiments only utilized ER100S-G as the potential fill pass material.

During the initial weld trials, the main finding was that small weave amplitude weld beads were resistant to solidification cracking for a large range of dilutions. However, similar dilution levels produced solidification cracking at larger weave amplitudes. The finding was different than the expectation of solidification cracking being heavily dependent on dilution. While it may relate to dilution, solidification cracking susceptibility was better related to weave amplitude due to small weave amplitudes having a lower internal restraint during solidification, which did not create enough restraint to cause solidification cracking. Wider weave amplitudes caused additional intrinsic restraint and resulted in cracking. It can also be understood when thinking about bead width. Small bead width generates lower total solid state solidification shrinkage than wider bead width, resulting in lower strain concentration in the interdendritic liquid. Another result of the preliminary weld trials was that under the same welding parameters and conditions, ER100S-G welded over Alloy 625 experienced solidification cracking, while ER100S-G welded over Alloy 686 contained no such defects due to a lack of Nb-rich eutectic constituent.

An explanation of the location of the solidification cracks was found by consulting the pseudo-binary phase diagrams, which were created with Thermo-CalcTM simulations. Upon examination of the diagram created for the ER100S-G/Alloy 625 combination (Figure 53), the solidification temperature range of the bulk weld metal composition can be easily determined. The majority of the welds of ER100S-G over Alloy 625 had dilutions in the range of 15-40%, which have a large STR. However, the region of the diagram that appears to be most susceptible to solidification cracking is from the range of 50-90% dilution due to a consistently larger STR. While the bulk weld metal may be somewhat susceptible to solidification cracking (~15-40% dilution), there exists a transition zone between the Alloy 625 substrate and ER100S-G weld metal which begins at 100% substrate composition and ends at the dilution of the bulk weld metal. It is in this transition region where the majority of solidification cracks nucleate due to the high STR experienced in the dilution range of the zone (100% to WM dilution).

Having achieved the goal of determining if welds of ER100S-G over Alloy 625 could avoid solidification cracking, design of experiments were performed using successful parameters developed during preliminary welding trials. DOE 1 (Table 19), which involved welding ER100S-G over Alloy 625, further concluded that solidification cracking was related to weave amplitude. Each individual weld bead was analyzed and a graph of dilution vs. heat input (Figure 63) was created that indicated that the combination of small weave amplitudes, high wire feed speed, and high oscillating travel speed created welds that were resistant to solidification cracking. The combination of the aforementioned factors subsequently resulted in higher dilution and higher heat input (calculated with oscillating travel speed), so it was determined that solidification cracking may be related to the internal restraint levels (weave amplitude) as well as the dilution achieved. This result is contrary to the initial belief that low dilution welds would be more resistant to solidification cracking. No solidification cracking was found in DOE 2, 3, 4, 5 or 6. No solidification cracking was found when welding ER100S-G over Alloy 686, UTP A 80 Ni, or Alloy 625 LNb.

6.2.2 Liquation Cracking

At the beginning of this study, liquation cracking was not a large concern. Instead, solidification cracking was the main defect that was thought to be encountered. However, liquation cracking was found in welds of ER100S-G over Alloy 625 in DOE 1 (Table 19) and 3 (Table 21). The Thermo-CalcTM partition coefficient results (Figure 56) indicated that Nb would segregate to the solidification sub-grain boundaries and form a eutectic constituent, which has a lower melting point than the surrounding matrix. The heat input from the subsequent weld raises the temperature of the heat affected zone of the previous

weld to cause melting of the NbC-rich grain boundaries. In the case of welding ER100S-G over Alloy 686 (DOE 2 and 4), UTP A 80 Ni (DOE 5), and Alloy 625LNb (DOE 6), no liquation cracking was found due to the absence of Nb and any elements that segregate to the sub-grain boundaries. The results of DOE 1 (Table 19) indicated that liquation cracking (Figure 65) occurs at high dilutions. CMT was used in this DOE, however, in order to attempt to alleviate solidification cracking, GMAW-P was used in DOE 3 (Table 21). Since GMAW-P is a higher heat input process, liquation cracking was hypothesized to worsen. While the GMAW-P process consistently resulted in higher heat input than CMT, it did not yield consistently higher dilutions when using identical parameters. However, the hypothesis was correct, with almost all of the welds in DOE 3 containing liquation cracking. The susceptible dilution range in DOE 3 was 25-40% dilution, however, DOE 1 contained multiple runs under 30% dilution with no liquation cracking, so the defect may be more closely related to heat input than dilution since DOE 3 experienced more instances of the defect than DOE 1. Both DOE 1 and 3 involved welding ER100S-G over Alloy 625.

6.2.3 Shrinkage Porosity

The mechanism proposed by Kou and Yang [72-78] accurately describes the nature of the interface between higher melting point weld metal and lower melting point base metal. However, their studies did not identify defect formation. Their mechanism proposed that, in the case of welding a consumable with a greater liquidus temperature (T_{LW}) over a base material with a lesser liquidus temperature (T_{LB}) , complete mixing in

the weld is impossible. Additionally, it was determined that the liquid weld metal boundary in contact with the substrate will cause a region of the base material to be at a temperature that is greater than T_{LB} . Thus, a "stagnant or laminar-flow layer of liquid base metal must exist regardless of weld pool convection." This region of liquid base material that does not mix with the weld pool is known as a "filler-deficient beach" (unmixed zone), and there is evidence of filler-deficient beaches in the welds of low alloy steel over nickel-based alloy (Figure 81). Also, it was noted that the thickness of the filler-deficient beach increases with increasing liquidus temperature difference (T_{LW}- T_{LB}). The difference in this experiment, calculated by Thermo-CalcTM (Table 14), is 147°C (1516°C –1369°C) with the combination of ER100S-G and Alloy 625 and 148°C (1516°C – 1368°C) with the combination of ER100S-G and Alloy 686. Since the liquidus temperature differences are similar, the thickness of the beaches in both cases should be similar. However, the solidus temperatures are drastically different for Alloy 625 and 686 at 1149°C and 1250°C respectively. This large difference in solidus temperatures suggests that, given a constant heat input, Alloy 625 will experience a larger filler deficient beach. A larger amount of Alloy 625 substrate material will reach the solidus temperature than with Alloy 686, which will cause more volume to fully melt.



Figure 81: Micrograph confirming existence of unmixed zone solidification with a fast cooling rate (smaller dendrites)

Next, considering the buffer layer materials, the liquidus temperature difference between UTP A 80 Ni and Alloy 625 was 52° C (1421° C – 1369° C) and the difference between ER100S-G and UTP A 80 Ni was 95° C (1516° C – 1421° C). These smaller liquidus temperature differences indicates a smoother transition from the Ni-based root pass to the LAS fill passes, with smaller filler-deficient beaches at each interface. For the Alloy 625 LNb buffer layer, the liquidus temperature difference with Alloy 625 is 22° C (1391° C – 1369° C) and with ER100S-G is 125° C (1516° C – 1391° C). When considering the potential compatibility of the buffer layers with the root and fill passes, UTP A 80 Ni would likely be more resistant to the shrinkage porosity defect due to a smoother transition between each material, minimizing the size of each filler-deficient beach. The Alloy 625 LNb has a drastic liquidus temperature difference with ER100S-G, but a small difference with Alloy 625, so shrinkage porosity defects would likely occur at the ER100S-G/Alloy 625 LNb interface.

Based on the mechanism proposed by Kou and Yang, a mechanism for shrinkage porosity formation in welds with $T_{LW} > T_{LB}$ is proposed. The existence of a continuous filler-deficient beach results in weld metal intrusions into the beach and solidification of the intrusions can occur while the beach is still in its liquid state (Figure 26 and Figure 27). These weld metal intrusions are random and in the case of this study, have resulted in filler-deficient peninsulas that resemble weld metal swirls, and also create fillerdeficient islands. These swirls, islands, and beaches were encountered in all DOEs (Figure 82). It is in these areas where the shrinkage porosity defects form due to the fact that the liquid beach, swirls, or islands become trapped between solidifying weld metal and solid base metal. When the liquid trapped between two solids attempts to finally solidify once T_{LB} is reached, thermal shrinkage is occurring and there is not enough solidifying liquid material to fill the space that the liquid material previously filled. Complete coalescence of the growing dendrites does not occur in small areas and a shrinkage porosity defect is formed (Figure 83).

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Figure 82: DOE 2 Run 8, Weld swirls, islands, and unmixed zone formation at the interface of ER100S-G weld metal and Alloy 686 substrate. Shrinkage porosity defect shown in weld swirl



Figure 83: SEM image of internal structure of a defect occurring in transition zone or filler deficient swirl between Alloy 686 substrate and ER100S-G weld metal (DOE 2, Run 12)

It was realized that attempting to quantify the frequency of occurrence of these defects and relate that data to dilution and/or heat input would be difficult due to examining merely one cross section of the weld. Nonetheless, a model was created to describe and visualize the phenomenon. EDS traverse compositional data and Thermo-Calc[™] data were combined to create EDS/ Thermo-Calc[™] Interface Models. The dilution, liquidus, and solidus values were graphed to show the transition at the interface. The traverse was performed on a weld of ER100S-G welded over Alloy 625 (Figure 68), and in order to compare nickel-based substrates, a similar traverse was made on a weld of ER100S-G over Alloy 686 (Figure 69). This allowed a visualization of the difference between the two substrates. It should be reiterated that the average dilutions of the two welds were different, with the weld over Alloy 625 having 17% dilution and the weld over Alloy 686 having 41% dilution. These values were calculated based on EDS data from the traverses with the portion of the traverse that existed in the weld metal outside the transition region. For both substrates, higher dilutions result in larger solidification temperature ranges according to Thermo-Calc[™].

Firstly, Alloy 625 has a larger liquidus/solidus temperature difference with ER100S-G than Alloy 686. Also, when comparing both traverses, the transition zone of the higher dilution weld over Alloy 686 (Figure 69) is much smaller than the transition zone of the lower dilution weld over Alloy 625 (Figure 68). The higher dilution caused the liquidus/solidus temperatures of the weld to more closely match the liquidus/solidus temperatures of the Ni-based substrate. This indicates that a higher dilution weld may be

more resistant to shrinkage porosity defects in the unmixed and transition zones because these zones will be smaller than for lower dilution welds. Next, Alloy 625 has a larger solidification temperature range than Alloy 686, so the filler deficient beach (unmixed zone) of Alloy 625 will need a longer time, given the same temperature gradient, to solidify. This extra time for solidification will allow additional shrinkage to occur during cooling due to thermal contraction and result in more shrinkage porosity defects. Finally, the Alloy 625 substrate experiences a dip in solidus temperature in the transition zone of the weld, while the Alloy 686 substrate shows only a small dip. This finding supports the fact that ER100S-G welded over Alloy 625 is more susceptible to shrinkage porosity defects than when welded over Alloy 686. It should also be mentioned that when the EDS traverses were being performed, it was noticed that there were small islands of material near the interface that closely matched the composition of the pure welding consumable, ER100S-G. These islands of nearly pure ER100S-G were found when welding over both Alloy 625 and Alloy 686. This indicates a lack of mixing at the interface that may also worsen shrinkage porosity defect formation.

Upon further examination of the EDS/Thermo-Calc[™] Interface Model, it was theorized that a steep temperature gradient could promote solidification to begin in the unmixed zone, proceed through the transition zone, and then continue in the bulk weld metal. Further work should be done to ascertain if such a gradient can be achieved using CMT or GMAW-P for this mode of solidification to occur. The low thermal conductivity of the substrate materials (Alloy 625 and Alloy 686) will also reduce this gradient.

It was realized after the creation of the EDS/Thermo-CalcTM Interface Model that a higher weld dilution would cause the liquidus and solidus temperatures of the weld to more closely match the liquidus and solidus of the nickel-based substrate, both 625 and 686. It was because of this observation that GMAW-P was attempted to increase the heat input and subsequent dilution of the welds. The CMT process was used in the first two DOEs due to an effort to avoid solidification and liquation cracking by attaining low dilution welds. GMAW-P was used in DOE 3 and 4 to examine the effect of higher heat input on shrinkage porosity defect formation. To correlate to earlier results, identical welding parameters were used for DOE 3 and 4 as were used in 1 and 2. This welding process change was expected to produce much higher dilution welds due to increased heat input. However, consistently higher dilutions were not achieved even though the heat input was consistently higher for GMAW-P than for CMT. As a result, there was not a good comparison of defect formation according to welding processes. For the buffer layer DOEs (5 and 6), a potentially viable combination was found with UTP A 80 Ni. The absence of shrinkage porosity defects at the UTP A 80 Ni/ER100S-G interface was predicted by Thermo-CalcTM with closely matching liquidus/solidus temperatures of the two materials. Despite shrinkage porosity defects at the interface with Alloy 625, this material was recommended for groove welding experiments because it has the best compatibility with ER100S-G. The welding experiment with Alloy 625 LNb (DOE 6) contained shrinkage porosity defects at both interfaces in nearly every trial, however, a constructive observation was made. The welds made with GMAW-P were relatively free of weld metal intrusions in the filler-deficient beach since there was better overall mixing

in the weld pool. This led to few and smaller defects due to less potential nucleation sites. With the CMT process, there was a noticeable lack of mixing near the fusion boundary, so more defects were able to form in the weld metal intrusions and islands.

It should be reiterated that while quantifying the defects was unsuccessful, a qualitative examination showed that Alloy 625 was more susceptible to shrinkage porosity defects than Alloy 686 when acting as a substrate for welds of ER100S-G with both CMT and GMAW-P. In addition, the qualitative analysis for the buffer layer materials was that UTP A 80 Ni is more resistant to this defect than Alloy 625 LNb at the interface with ER100S-G, so it should be selected for groove welding experiments. The difficulty in quantifying the defects stemmed from the fact that only one cross section of each weld run was prepared for metallography and subsequent examination. Due to the random nature of weld pool flow with formation of the weld metal intrusions into the unmixed zone, one cross section likely did not allow enough data on the defects to be collected and a qualitative approach was used instead. Future studies may benefit from the extraction of multiple samples to gain a better metric for the defects, such as size and quantity for each weld pass.

6.2.4 Summary of Defects

When welding low alloy steel over nickel-based alloy, this study has found that multiple defects can be encountered. Two types of weld cracking, solidification and liquation, were found as well as a new type of defect, shrinkage porosity, which has only been reported in castings before this study [59-62]. While the main hurdle to the ER100S-G FM/Alloy 625 combination was initially thought to be merely solidification cracking, the other two types of defects, liquation cracking and shrinkage porosity, were encountered during experiments and need to be addressed. During the course of this study, the main reason for solidification cracking was found to be high restraint (large weave amplitudes) and low dilution. Solidification cracking was only found in the combination of ER100S-G welded over Alloy 625 and was never encountered when welding ER100S-G over Alloy 686, UTP A 80 Ni, or Alloy 625 LNb in the bead-on-plate geometry.

Liquation cracking was also found in the combination of ER100S-G and Alloy 625, but not with LAS and any other Ni-based alloy. Liquation cracking was correlated to high heat input and high dilution. High dilution causes more Nb content to exist in the WM and the higher heat input would impose a greater temperature to exist in the HAZ of the underlying material. As far as cracking is concerned, when welding LAS over Alloy 625, an attempt to alleviate one problem may worsen the other. If high dilutions are needed to avoid solidification cracking, then liquation cracking could occur. This problem only gets worse when shrinkage porosity is considered in this situation. While shrinkage porosity was encountered in most combinations of LAS and Ni-based alloy, it was concluded that if the weld metal closely matches the liquidus/solidus of the substrate, shrinkage porosity may be alleviated due to the formation of a smaller unmixed zone. This conclusion points towards higher dilution welds being attractive to avoid the shrinkage porosity defects. However, higher dilution welds may also worsen liquation cracking susceptibility. It becomes clear that while attempting to weld ER100S-G over Alloy 625 may have the possibility of being metallurgically compatible, the chances of avoiding all three aforementioned defects are slim. It should be reiterated that the combination of LAS and Alloy 625 experienced all three defects, while the ER100S-G and Alloy 686/Alloy 625 LNb combinations only experienced shrinkage porosity. When welding ER100S-G over UTP A 80 Ni, no defects are experienced, however, shrinkage porosity occurs at the interface of Alloy 625 and UTP A 80 Ni. This leads to the conclusion that LAS is not a viable solution to weld directly over Alloy 625, however, may be possible to weld over Alloy 686 and UTP A 80 Ni. Alloy 686 should be selected for groove welding trials to act as the root pass and UTP A 80 Ni should be tested as a buffer layer in groove welding trials. Since the bead-on-plate welding tests showed that welding of UTP A 80 Ni over Alloy 625 contained shrinkage porosity at the interface, and that Alloy 625 is likely more prone to the defect than Alloy 686, groove welding trials should utilize Alloy 686 as the root pass for the UTP A 80 Ni buffer layer experiments.

The bead-on-plate experiments have provided baseline metallurgical information that was necessary to better understand what problems may be encountered when groove welding is attempted in future experiments. It is postulated that the higher restraint conditions of groove welding could worsen solidification cracking susceptibility. Also, for the shrinkage porosity issue, the heat extraction conditions of the groove geometry may alter defect formation.

6.3 Groove Welding Experiments

During the groove welding experiments with ER100S-G fill passes welded directly over Alloy 686 root pass, parameter development was attempted to eliminate solidification cracking, shrinkage porosity, and lack of fusion defects. It should be noted that bead-on-plate trials of ER100S-G welds over Alloy 686 substrate never resulted in solidification cracking, however, solidification cracking was found with this combination in narrow groove welds. Liquation cracking was not experienced with this combination in both bead-on-plate trials and groove welding. CMT was chosen as the primary welding process for groove welding experiments and was used for Groove 1-8 due to preliminary groove welding experiments with GMAW-P having an unstable arc due to the shielding gas composition. A lower CO₂ content is needed to stabilize the arc during groove welding with GMAW-P. The parameters and outputs for each trial are provided in Table 25 and for each pass in Table 26 and Table 27.

6.3.1 Solidification Cracking in Groove welding

After preliminary groove welding trials to establish welding parameters in the groove geometry, 8 groove welds were created and analyzed. Groove 1, with the weave program mistakenly set to weave by "time" rather than "frequency", resulted in a sluggish weave. Solidification cracking was found in Pass 1 and 2 and the cracking as apparent during welding as centerline cracks. The microstructure in Pass 1 consisted of primarily austenite, although it was unable to be etched due to the high nickel and iron contents. Pass 2 consisted of a mixture of austenite and martensite (Figure 74). A

dendritic structure was present with needle-like formations. No cracks were found propagating from the interface of the root pass and Pass 1 or the interface of Pass 1 and 2. This suggests that while the centerline cracking may be related to the metallurgical combination, they are also related to the welding parameters. This trial resulted in the weld pool to solidify in a tear drop shape, leaving a defined centerline, which opened a crack as solidification progressed. Also, despite the inability to measure dilution quantitatively, a qualitative examination of Groove 1 (Figure 73) shows a high dilution of Pass 1 by the Alloy 686 root pass when compared to the cross sections of other trials. This was likely a byproduct of the sluggish weave causing more melting due to the slow oscillating travel. To remedy, the centerline cracking, Groove 2 was attempted with an appropriate weave program, that was set to weave by frequency rather than time, and solidification cracking was not found in any pass. The weld pool shape for Groove 2 (Figure 74) was an ovular shape rather than a tear drop shape, which is known to be more resistant to centerline solidification cracking. The reason why centerline solidification cracking is believed to be a process related defect is that the microstructure of Pass 1 and 2 of Groove 2 were similar to the microstructures in Groove 1, but no cracking occurred.

For Groove 3 (Figure 75), the travel speed and wire feed speed were increased to test the limits of travel speed and cracking response. The presence of solidification centerline cracks indicates that welds with travel speeds of 5.1 mm/s (12 inch/min) are susceptible to the defect while welds with travel speeds of 4.2 mm/s (10 inch/min) are more resistant. Groove 5 (Figure 77) also used a travel speed of 5.1 mm/s, but a smaller weave apex, with remaining parameters similar to those used in Groove 3 and centerline

cracking was present in Pass 2, but not pass 1. Groove 4 (Figure 76) used the lower travel speed (4.2 mm/s) and an extremely small centerline crack was found in Pass 1. The travel speed may need to be decreased below 4.2 mm/s to ensure cracking does not occur.

Groove 6 (Figure 78) matched the parameters of Groove 4, but the weave apex was set to reach ¹/₂ wire diameter away from the interface and the push angle was increased. A micro crack was found in Pass 2, but it was discontinuous. It occurred near the centerline of the weld, so its existence can be attributed to excessive travel speed. As in Groove 4, a slightly lower travel speed should be used to avoid these small centerline cracks. No cracking was found in Groove 7 (Figure 79) or 8 (Figure 80).

All of the solidification cracking that was seen in groove welding experiments occurred along the centerline of Pass 1 and/or Pass 2 of the ER100S-G fill passes welded using CMT. Their presence at the centerline suggest that the cracking can be eliminated with proper control of welding parameters, namely, travel speed. If special care is taken to ensure that the weld pool solidifies in an elliptical shape, cracking can be avoided when welding ER100S-G fill passes over Alloy 686 root pass in the narrow groove configuration. Additionally, if the bead height was reduced for the fill passes, solidification stresses along the centerline would also be reduced and cracking could be eliminated in that manner.

6.3.2 Shrinkage Porosity in Groove Welding

Shrinkage porosity was encountered in Groove 3 (Figure 75), 4 (Figure 76), 6 (Figure 78), 7 (Figure 79), and 8 (Figure 80), however, the occurrence of this defect

appears to be random. In these trials, only one shrinkage porosity defect was found per trial near the interface of the root pass and Pass 1. They appeared in the unmixed zone of the root pass material, swirls of root pass material that extends into the Pass 1 weld metal, and islands of root pass material in Pass 1. There is a lack of mixing that occurs near the interface of the Alloy 686 root pass and Pass 1 when using CMT. This lack of mixing may be attributed to the low heat input of CMT, with GMAW-P likely achieving better mixing. The higher heat input will make the WM reach higher temperatures and cause the areas near the substrate to better mix with the substrate material due to this higher temperature. The CMT process was chosen as the primary welding process for groove welding experiments in an attempt to decrease the dilution of Pass 1 and 2 by the Alloy 686 root pass to lessen the chances for solidification cracking and liquation cracking, but a change to GMAW-P may increase overall weld mixing. Better mixing would eliminate potential shrinkage porosity nucleation sites such as fill deficient swirls and islands, but may increase the size of the unmixed zone in the root pass during welding of Pass 1.

6.3.3 Lack of Fusion Defects in Groove Welding

When considering lack of fusion defects in the groove welding experiments, Groove 1 will not be considered due to an improper weave program. The primary variable that was changed to eliminate LOF was the weave apex of the weld passes. Initially for Groove 1-3, the weave amplitude was programed at a value such that the weave apex was 1 wire diameter away from each sidewall. Groove 2 (Figure 74) exhibited LOF on nearly every pass at the triple-point of each weld pass. Groove 3 (Figure 75), with an increased travel speed, wire feed speed, and weave frequency had LOF defects, but the defects only occurred within a pass where the arc began wandering up the sidewall. On all passes that deposited well, no LOF was found. Due to solidification cracking in this trial, different parameters were attempted to simultaneously eliminate solidification cracking and LOF.

After Grooves 1-3, it was suggested by industry experts to attempt a much lower weave frequency and a much farther weave apex. The weave frequency was halved to a value of 2 Hz and it was noted that this value at these travel speeds allowed 5 "bumps" per inch on each sidewall during the weld. Prior attempts resulted in nearly 10 "bumps" per inch. The weave apex was also increased to impinge each sidewall by ½ wire diameter with the intention of increasing the sidewall interaction. Groove 4 (Figure 76) matched most of the parameters of Groove 2, with the exception of weave frequency and weave apex. The LOF defects were drastically reduced, with only a few indications of the defect. However, during welding of the fill passes, it was noted that severe undercut was occurring at the sidewalls. Severe undercut also occurred in Groove 5 (Figure 77), which had a higher travel speed and wire feed speed than Groove 4. There were only two instances of LOF, but since undercut was encountered, it was realized that the large weave apex was not going to be acceptable.

Grooves 6 (Figure 78) and 7 (Figure 79) were performed with a weave apex that was ¹/₂ wire diameter away from the sidewall to decrease the amount of undercutting that occurred in Grooves 4 and 5. The push angle of the welding torch was also increased from 3° to 5° to push more heat ahead of the weld pool rather than the heat being directed into the top of the weld pool. These tactics did not eliminate the LOF defects that occurred at the triple-points of the fill passes. For the cross sections examined, it appeared that the weave apex of ½ wire diameter away from the sidewall experienced more LOF defects than the trials with weave apex of ½ wire diameter into each sidewall (Groove 4 and 5). However, the wider weave apex also caused undercutting. At this point, it was determined that a median weave apex should be attempted. Also, the increased torch angle appeared to move the wire electrode to the front edge of the weld pool, which would increase the heat into the previous pass and sidewalls rather than merely heat the weld pool.

Groove 8 (Figure 80) was completed with a weave apex that had the electrode tip touching each sidewall, the median apex between ½ wire diameter away from and into the sidewall. The increased 5° push angle was also used in this trial. Some undercutting was seen during welding, but it was much reduced from the trials that used the wider weave apex (Groove 4 and 5). Much better fusion was achieved at the triple-points, but two small defects still occurred. This trial was the best parameter set due to no cracking and very limited LOF. It is recommended that future iterations of welding ER100S-G over Alloy 686 in the groove configuration should use the weave apex where the wire tip touches each sidewall.

Due to Cold Metal Transfer being a low heat input process and not normally being used for narrow groove welding, LOF were difficult to eliminate in the limited number of groove welding trials that were completed. The trend from the experiments indicates that the weave apex plays a large role in eliminating LOF. An overly large apex (1/2 wire diameter into sidewall) will cause undercutting and, while the wider apex will ensure melting of both sidewalls, can potentially lead to LOF forming in the undercut defects. On the opposite end of the spectrum, a weave apex that is too small (1/2 to 1 wire diameter from wall) will not result in undercutting, but LOF defects can be encountered. The heat from the welding arc does not reach the triple-points of the groove and small defects can result. When using the median weave apex (wire touching each sidewall), a few lack of fusion defects may still occur, but they will be much smaller than when using other weave apexes. Also, the increased torch angle of 5° should also be used due to the observation of the wire tip entering the weld pool at the leading edge, rather than towards the center with a 3° angle. This increased angle should focus more heat into the previous pass and sidewalls so that proper fusion can be achieved.

Chapter 7: Summary & Conclusions

7.1 Computational Modeling

- Based on Thermo-Calc[™] simulations, ER100S-G is more compatible with Alloy 625 than ER70S-G. Both low alloy steel materials have similar compatibility with Alloy 686. Primary austenitic solidification is predicted across all dilutions. ER100S-G was selected as the primary low alloy steel consumable for welding experiments.
- Alloy 625, when diluted into low alloy steel, experiences a large STR across all dilutions. Alloy 686 has a narrower STR across all dilutions when diluted into low alloy steel. Alloy 686 is more compatible, but both alloys were selected to undergo weld testing with ER100S-G.
- Alloy 625 also suffers from a NbC eutectic that forms at the grain boundaries, as predicted with partitioning coefficients in the dilutions range of actual welds (10-50% dilution).
- 4. For the buffer layer materials, a better transition of liquidus/solidus temperatures was experienced with UTP A 80 Ni than Alloy 625 LNb when considering the root and fill pass materials. Both were selected for weld testing as a buffer between the Ni-based alloy root pass and low alloy steel fill passes due to the good transition that UTP A 80 Ni provides as well as excellent compatibility of Alloy 625 LNb and ER100S-G.

7.2 Bead-on-Plate Welding Experiments

- ER70S-6 was verified to be extremely susceptible to solidification cracking when welded over Alloy 625. When using the same welding parameters, ER100S-G was relatively resistant.
- 2. Solidification cracking can be avoided in the bead-on-plate geometry when welding ER100S-G over Alloy 625 with CMT and GMAW-P processes. Solidification cracks that propagate from the interface and weld swirls were experienced at large weave amplitudes (≥ 3 mm). At large weave amplitudes, low heat input (calculated with oscillating travel speed) and low dilution (15-30%) welds experienced solidification cracking. Solidification cracking was not experienced in welds of ER100S-G over Alloy 686, UTP A 80 Ni, or Alloy 625 LNb.
- 3. Welds of ER100S-G over Alloy 625 may be susceptible to liquation cracking at high dilutions (≥ 30%) and when welded with high heat input welding processes. CMT welds were more resistant to the defect than GMAW-P welds. Cracks were found in the first pass that occurred during welding of the overlapping second pass that propagate from the interface of pass one and two. Liquation cracking was not experienced in welds of ER100S-G over Alloy 686, UTP A 80 Ni, or Alloy 625 LNb.
- 4. Welds of ER100S-G over Alloy 625, Alloy 686, and Alloy 625 LNb are susceptible to shrinkage porosity, a defect previously undocumented in welding.

No such defects were found at the interface of ER100S-G and UTP A 80 Ni. These defects occur in the unmixed zone of the Ni-based substrate, the transition zone, adjacent to weld metal intrusions into the unmixed zone, and islands of unmixed Ni-based material that exist near the interface. A theory of the mechanism of shrinkage porosity formation was developed when welding a high T_m material over a low T_m material. Due to the random nature of the defect and only examining one cross section from each weld, only qualitative observations were made pertaining to shrinkage porosity.

5. The materials and configurations recommended for groove welding trials are as follows: ER100S-G fill passes welded directly over Alloy 686 root pass. ER100S-G fill passes welded over a buffer layer of UTP A 80 Ni that is welded over Alloy 686 root pass.

7.3 Groove Welding Experiments

- With optimized parameters, it is possible to weld ER100S-G fill passes directly over Alloy 686 root pass in the narrow groove configuration with limited defect formation. Solidification and liquation cracking can be avoided and LOF can be drastically reduced. Small shrinkage porosity defects may exist at the interface of the root pass and fill Pass 1.
- All solidification cracks that were found occurred along the centerline of the welds in Pass 1 and/or 2. Pass 1 had an austenitic microstructure and Pass 2 had a combination of austenite and martensite.

- 3. Solidification cracking was found to be related to travel speed. Large cracks in Passes 1 and 2 occurred when the travel speed was set to 5.1 mm/s. A travel speed of 4.2 mm/s yielded only tiny cracks in Passes 1 and 2 with some trials showing no cracking. This indicates the upper limit for travel speed for solidification crack formation.
- 4. Shrinkage porosity defects were found at the interface of the root pass and Pass 1 in some of the trials. These defects occurred in the unmixed zone of the Alloy 686 root pass, swirls of Alloy 686 that extend into Pass 1, and islands of root pass material that were swept up into Pass 1. No correlation could be made since dilution was not measured for the groove welds.
- 5. Lack of fusion defects were present at the triple-points (where the weld contacts the prior pass and sidewall) of fill pass welds. A range of weave amplitudes were attempted to obtain different weave apexes for different trials. The weave apex that resulted in the least lack of fusion defects was when the electrode tip contacted each sidewall at the farthest point of the weave amplitude. A steeper torch push angle (5°) also assisted in pushing more heat into the previous weld and eliminating LOF.

Chapter 8: Future Work

Due to the success during groove welding trials of ER100S-G fill passes welded directly over Alloy 686 root pass, this combination should be attempted in a groove weld on an X65 pipe. The same preheat and interpass temperatures should be used in pipe welding as groove welding with flat plates. Due to the furnace size restriction, preheating and interpass temperatures should be achieved with a propane torch and rosebud attachment. No further testing of ER100S-G welded directly over Alloy 625 should be attempted due to the possibility of multiple defects (solidification cracking, liquation cracking, and shrinkage porosity). Also, if GMAW-P is attempted for ER100S-G fill passes over Alloy 686, there should be a consideration to change shielding gases to an Ar/CO₂ mix with a lower CO₂ content (10-15%). This may help with arc stability with this welding process.

Due to time constraints, additional groove welding parameters were not attempted. It may be beneficial to lessen the bead height of the fill passes to lessen the solidification stresses occurring at the weld centerline. A shorter bead may also be advantageous if it becomes necessary to use a bead tempering technique.

Next, due to the success of welding ER100S-G over a buffer layer of UTP A 80 Ni in bead-on-plate trials (DOE 5), this buffer layer should be attempted in groove welding on flat plates to assess compatibility. This was the only combination that did not have shrinkage porosity. In the bead-on-plate experiment, welding of the buffer layer was performed over a layer of Alloy 625, however, the recommendation for groove welding would be to weld over an Alloy 686 root pass. The liquidus/solidus temperatures will more closely match with Alloy 686 than with Alloy 625. If successful, pipe welding with this buffer layer solution should be attempted. If needed, for all groove welding, Energy Dispersive Spectroscopy (EDS) should be used to calculate the dilution of each weld pass since it is not feasible to calculate geometrically.

Also, mechanical testing on the groove welds should be attempted for whatever configuration is free of defects. Mechanical testing should only be carried out on groove welds using X65 base material (pipe). The flat plates of 1018 have a lower strength and may alter the test results. Tensile testing should be performed to ascertain that the 100MPa strength overmatch is obtained. Due to the presence of a gradient of mechanical properties in the groove due to varying dilution levels of ER100S-G fill passes, tensile testing may need to be performed for each weld pass. Transverse bend testing should also be performed.

Finally, since the main focus of this study was to find metallurgical compatibility between the low alloy steel fill passes and Ni-based root pass, interactions with the pipe material were not explicitly analyzed. Hardness tests can be performed on the groove welds on the X65 material to ensure that hardness over 250 HV is not encountered (NACE requirement [102]). If it is found that the hardness is over the limiting amount, temper bead techniques should be studied in an attempt to temper prior welding passes as well as the surrounding HAZ. Also, due to the potential for high hardness microstructures, the Delayed Hydrogen Cracking Test (DHCT) can be used to quantify susceptibility to HIC.

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Appendix: Welding Parameters & Outputs

This section contains the welding parameters and important outputs from the welding experiments, both bead-on-plate and groove welding.

- 1. Preliminary Weld Trials
- 2. DOE 1
- 3. DOE 2
- 4. DOE 3
- 5. DOE 4
- 6. DOE 5
- 7. DOE 6
- 8. Groove Welding Trials

	Inputs											Outputs			
Trial	Filler Wire	Substrate	Process	Shielding Gas	Weave Amplitude (mm)	Weave Frequency (Hz)	Wire Feed Speed (mm/s)	Travel Speed (mm/s)	Heat input (J/mm)*	Heat input (J/mm)**	Dilution (First Pass)	Dilution (Second Pass)	Solidification Cracking	Shrinkage Porosity	
1	ER100S- G	Alloy 625	CMT	Ar/25%CO2	4		44.4	5.5	291.9	25.6	31.81%	14.94%	First Pass	Yes	
2	ER100S- G	Alloy 625	CMT	Tri-Mix	6	-	72.0	5.5	362.8	21.3	14.93%	7.78%	First Pass	No	
3	ER100S- G	Alloy 625	GMAW- P	Tri-Mix	6	3.9	76.2	5.5	547.3	32.1	35.78%	16.71%	First Pass	Yes	
4	ER70S-6	Alloy 625	CMT	Tri-Mix	4		48.7	5.5	251.3	22.1	16.37%	11.02%	Both Passes	No	
5	ER100S- G	Alloy 625	CMT	Tri-Mix	4		44.4	5.5	255.7	22.4	15.07%	5.45%	First Pass	Yes	
6	ER100S- G	Alloy 625	GMAW- P	Ar/25%CO2	Stringer	N/A	114.3	9.7	469.9	N/A	55.54%	44.66%	Both	No	
7	ER100S- G	Alloy 625	GMAW- P	Ar/25%CO2	3		42.3	5.5	304.6	35.6	29.69%	17.13%	None	Yes	
8	ER100S- G	Alloy 625	CMT	Ar/25%CO2	2		33.9	5.5	223.9	38.9	36.24%	16.83%	None	No	
9	ER100S- G	Alloy 686	CMT	Ar/25%CO2	4	-	44.4	5.5	304.3	26.7	28.64%	22.61%	None	Yes	
10	ER70S-6	Alloy 686	CMT	Ar/25%CO2	4	3.9	48.7	5.5	323.2	28.4	32.60%	21.04%	First Pass (Small)	Yes	
11	ER100S- G	Alloy 686	GMAW- P	Ar/25%CO2	3	-	42.3	5.5	295	34.4	41.70%	20.57%	None	No	
12	ER100S- G	Alloy 625	CMT	Ar/25%CO2	4	-	44.4	5.5	289.2	25.4	27.59%	20.28%	First Pass	Yes	
13	ER100S- G	Alloy 625	CMT	Tri-Mix	2	-	33.9	5.5	207.4	36	2.61%	10.20%	None	Yes	

Table 18: Welding Parameters of Preliminary Weld Trials

	DOE 1: Factors								DOE 1: Outputs						
Run	Weave Amplitude (mm)	Wire Feed Speed (mm/sec)	Travel Speed (mm/sec)	OTS (mm/sec)	Power Ratio** (W/mm^2)	Dilution 1st bead (%)	Dilution 2nd bead (%)	Heat Input* (J/mm)	Heat Input** (J/mm)	Solidification Cracking	Liquation Cracking	Shrinkage Porosity			
1	2	29.6	4.65	31.5	1049.2	30.5	20.7	240.1	35.4	None	None	First, Small			
2	3	46.6	4.65	47	1425.1	32.5	18.8	343.9	34.0	None	First, Small	Both, Medium			
3	3	38.1	5.5	47.1	1561.1	31.5	18.3	259.8	30.3	Second, Small	None	Both, Small			
4	2.5	29.6	5.5	39.4	1290.3	32.0	20.3	199.3	27.8	Second, Large	None	First, Small			
5	2.5	38.1	5.5	39.4	1309.7	32.7	20.3	260.3	36.4	None	First, Large	Second, Small			
6	3	29.6	4.65	47	1581.4	28.5	16.2	242.4	24.0	Both, Large	None	Both, Large			
7	2	46.6	4.65	31.5	959.5	37.2	22.2	345.7	51.0	None	Micro	Second, Micro			
8	2	29.6	6.35	31.8	1081.2	28.7	20.4	179.5	35.8	None	None	Both, Medium			
9	2.5	38.1	6.35	39.5	1332.3	25.6	13.5	229.0	36.8	None	None	Both, Medium			
11	2	46.6	6.35	31.8	980.4	37.8	19.1	256.2	51.1	None	None	Second, Micro			
12	3	46.6	6.35	47.2	1508.2	35.5	26.6	265.4	35.7	None	None	Second, Micro			
13	3	29.6	6.35	47.2	1573.0	21.9	19.5	175.8	23.6	First, Large	None	Both, Small			

Table 19: DOE 1 Results for Defect Formation (Alloy 625, CMT)

		DOE 2	: Factors		DOE 2: Outputs						
Run	Weave Amplitude (mm)	Wire Feed Speed (mm/sec)	Travel Speed (mm/sec)	OTS (mm/sec)	Dilution 1st bead (%)	Dilution 2nd bead (%)	Heat Input* (J/mm)	Heat Input** (J/mm)	Shrinkage Porosity		
1	2	29.6	4.65	31.5	30.0	23.3	249.7	36.8	Both, Small		
2	3	46.6	4.65	47	29.5	18.5	358.5	35.4	Both, Medium		
3	3	38.1	5.5	47.1	32.7	19.2	264.9	30.9	Second, Small		
4	2.5	29.6	5.5	39.4	30.1	14.6	209.0	29.2	Second, Small		
5	2.5	38.1	5.5	39.4	29.5	24.0	268.8	37.5	Second, Micro		
6	3	29.6	4.65	47	22.8	15.3	244.9	24.2	Second, Micro		
7	2	46.6	4.65	31.5	32.3	24.9	346.8	51.1	Both, Small		
8	2	29.6	6.35	31.8	30.1	16.0	185.0	36.9	Both, Small		
9	2.5	38.1	6.35	39.5	30.7	25.4	228.3	36.7	Both, Small		
11	2	46.6	6.35	31.8	35.6	18.7	253.4	50.5	First, Micro		
12	3	46.6	6.35	47.2	33.9	20.5	268.9	36.1	First, Large		
13	3	29.6	6.35	47.2	24.1	19.2	180.9	24.3	Second, Large		

Table 20: DOE 2 Results for Defect Formation (Alloy 686, CMT)

		DOE 3: 1	Factors			DOE 3: Outputs						
Run	Weave Amplitude (mm)	Wire Feed Speed (mm/sec)	Travel Speed (mm/sec)	OTS (mm/sec)	Dilution 1st bead (%)	Dilution 2nd bead (%)	Heat Input* (J/mm)	Heat Input** (J/mm)	Shrinkage Porosity (pass and severity)	Liquation Cracking		
1	2	29.6	4.65	31.5	36.2	14.3	254.2	37.5	First, Large, 3	Medium, 2		
4	2.5	29.6	5.5	39.4	19.0	10.2	214.8	30.0	Second, Small, 1	None, 0		
5	2.5	38.1	5.5	39.4	32.6	23.2	278.7	38.9	Second, Medium, 2	Large, 3		
7	2	46.6	4.65	31.5	40.1	26.7	395.9	58.4	Second, Small, 1	Medium, 2		
8	2	29.6	6.35	31.8	25.2	11.1	192.1	38.3	Second, Large, 3	Small, 1		
9	2.5	38.1	6.35	39.5	32.0	17.5	232.4	37.4	Second, Small, 1	Small, 1		
11	2	46.6	6.35	31.8	43.5	21.3	281.5	56.1	Second, Medium, 2	None, 0		

Table 21: DOE 3 Results for Defect Formation (Alloy 625, GMAW-P)

		DOE 4: H	Factors	DOE 4: Outputs						
Run	Weave Amplitude (mm)	Wire Feed Speed (mm/sec)	Travel Speed (mm/sec)	OTS (mm/sec)	Dilution 1st bead (%)	Dilution 2nd bead (%)	Heat Input* (J/mm^2)	Heat Input** (J/mm^2)	Shrinkage Porosity	
1	2	29.6	4.65	31.5	30.9	15.3	274.7	40.5	First, Medium, 2 Second, Small, 0.5	
4	2.5	29.6	5.5	39.4	25.5	20.6	242.2	33.8	First, Large, 4 Second, Small, 0.5	
5	2.5	38.1	5.5	39.4	42.1	24.8	304.7	42.6	First, None, 0 Second, Small, 1	
7	2	46.6	4.65	31.5	36.1	25.9	442.9	65.3	First, Small, 1.5 Second, Medium, 2	
8	2	29.6	6.35	31.8	32.9	19.0	202.8	40.4	First, Small, 1 Second, Medium, 2.5	
9	2.5	38.1	6.35	39.5	30.7	17.8	238.8	38.4	First, Large, 3 Second, None, 0	
11	2	46.6	6.35	31.8	43.7	23.5	326.1	65.0	First, Small, 1.5 Second, Medium, 2.5	

Table 22: DOE 4 Results for Defect Formation (Alloy 686, GMAW-P)

		DC	DE 5: Factors				DOE 5: Outputs						
Run	Welding Process	Weave Amplitude (mm)	Wire Feed Speed (mm/sec)	Travel Speed (mm/sec)	OTS (mm/sec)	Dilution 1st bead	Dilution 2nd bead	Heat Input*	Heat Input**	Shrinkage Porosity			
1A		2	29.6	4.65	31.5	11.5%	2.6%	229.5	33.8	***			
5A	- CMT	CMT 2.5		5.5	39.4	12.3%	3.6%	257.6	36.0	***			
11A		2	46.6	6.35	31.8	15.3%	6.2%	255.9	51.0	Yes (buffer/root interface)			
1B		2	29.6	4.65	31.5	12.6%	N/A	253.8	37.4	Yes (buffer/root interface)			
5B	GMAW-P	GMAW-P 2.5		5.5	39.4	15.4%	N/A	272.3	38.0	***			
11B	_	2	46.6	6.35	31.8	22.5%	N/A	300.6	59.9	***			

Table 23: DOE 5 Results for Defect Formation (Alloy 625 with UTP A 80 Ni buffer, CMT and GMAW-P)

*Calculated with longitudinal travel speed **Calculated with oscillating travel speed ***Only visible after etch at the buffer/root interface, none seen at buffer/fill interface
		DC	E 6: Factors	DOE 6: Outputs						
Run	Welding Process	Weave Amplitude (mm)	Wire Feed Speed (mm/sec)	Travel Speed (mm/sec)	OTS (mm/sec)	Dilution 1st bead	Dilution 2nd bead	Heat Input*	Heat Input**	Shrinkage Porosity
1A		2	29.6	4.65	31.5	22.3%	10.1%	246.3	36.3	First, Small, 1 Second, Small, 1
5A	CMT	2.5	38.1	5.5	39.4	25.3%	14.6%	267.3	37.3	First, Micro, 0.5 Second, Micro, 0.5
11A		2	46.6	6.35	31.8	29.8%	16.6%	281.4	56.1	First, Micro, 0.5 Second, Small, 1
1B		2	29.6	4.65	31.5	31.3%	N/A	N/A	N/A	First, Small, 1
5B	GMAW-P	2.5	38.1	5.5	39.4	26.8%	N/A	N/A	N/A	None
11B		2	46.6	6.35	31.8	27.2%	N/A	283.7	56.6	First, Micro, 0.5

Table 24: DOE 6 Results for Defect Formation (Alloy 625 with Alloy 625 LNb buffer, CMT and GMAW-P)

*Calculated with longitudinal travel speed **Calculated with oscillating travel speed

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Trial	Welding Process	Ts (mm/s)	WFS (mm/s)	Weave Amp. (mm)	Weave Freq. (Hz)	Dwell Time (s)	Push Angle (deg.)	Weave Apex	Weave Type	# of Passes	Lack of Fusion	Solidification Cracking	Shrinkage Porosity
Groove 1		4.2	120.7	2-2.5	3-4		3	1 wire diameter from sidewall	By Time	6	Present at most all triple-points	Extensive in Pass 1 and 2 (centerline)	None
Groove 2	-	4.2	127	1.8-2.8	4	-	3	l wire diameter from sidewall	By Frequency	6	Present at most triple-points	None	None
Groove 3	_	5.1	135.5	1.8-2.7	4.3-4.5	_	3	l wire diameter from sidewall	By Frequency	6	1 at triple-point	2 large cracks in Pass 1 and 2 (centerline)	One small at root/Pass 1 interface
Groove 4	CMT	4.2	122.8	2.85- 4.0	2	0.1	3	¹ / ₂ wire diameter into sidewall	By Frequency	4	2 large at triple- points	l extremely small crack in Pass 1 (centerline)	One large at root/Pass 1 interface
Groove 5	-	5.1	131.2	2.7-4.0	2	-	3	¹ / ₂ wire diameter into sidewall	By Frequency	4	2 large at triple- points	2 medium cracks in Pass 2	None
Groove 6	-	4.2	122.8	2.5-3.4	2	-	5	¹ / ₂ wire diameter from sidewall	By Frequency	4	5 medium at triple-points	1 extremely small crack in Pass 2	One large at root/Pass 1 interface
Groove 7		4.2	112.2	2.5-3.4	2		5	¹ / ₂ wire diameter from sidewall	By Frequency	4	3 medium, 1 large at triple- points	None	1 small in island near root/Pass 1 interface
Groove 8	-	4.2	122.8	2.65- 3.85	2	-	5	Touching each sidewall	By Frequency	5	2 small at triple- points	None	1 medium at root/Pass 1 interface

Table 25: Su	mmary of all	Groove	Welding	Trials
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Trial	Pass	Weave Amp. (mm)	Heat Input (LTS) (J/mm)	Heat Input (OTS) (J/mm)	Preheat Temp (F)	Comments	Defects
	1	2	1046	263	123/123	Good profile, visible centerline crack	1, 3
	2	2	1041	262	117/120	Excessive convexity, visible centerline crack	1, 3
Groove	3	2	1054	225	120/120	Excellent appearance	3
1	4	2.2	1074	211	120/123	Excellent appearance	3
	5	2.5	1080	189	119/119	Deposited on one side more than other	3
	6	2.5	1069	187	126/126	Bad deposition on one side	3
	1	1.8	1194	280	123/126	Great appearance and flat bead	3
	2	1.8	1171	274	123/123	Great appearance and flat bead	3
Groove	3	2.1	1170	239	123/124	Excessive penetration on sidewalls	3
2	4	2.4	1176	213	120/124	Great appearance	3
	5	2.6	1178	198	121/121	Great appearance	3
	6	2.8	1178	185	125/124	Great appearance	3
	1	1.8	1065	275	122/127	Great appearance, dug into sidewall	1, 2
	2	1.8	1031	266	120/123	Great appearance, centerline crack, dug into wall	1
Groove	3	2.1	1044	236	122/122	Great appearance	None
3	4	2.4	1023	205	124/124	Weave too wide, deposited up sidewall	3
	5	2.4	1040	209	123/123	Great appearance	3
	6	2.7	1040	188	122/123	Great appearance	None
	1	2.85	967	230	128/128	Flat, slightly convex. Higher buildup on right side. Did not impinge sidewall due to contact tip	1, 2, 3
Groove	2	3	948	215	120/120	Slightly convex, great appearance	3
4	3	3.5	950	188	118/118	Flat, slightly convex, great appearance	3
_	4	4	993	173	121/121	Slightly convex, great appearance	None

Table 26: Summary of Welding Passes for each Groove Trial (Groove 1-4)

1 = Solidification cracking 2 = Shrinkage porosity 3 = Lack of fusion

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Trial	Pass	Weave Amp. (mm)	Heat Input (LTS) (J/mm)	Heat Input (OTS) (J/mm)	Preheat Temp (F)	Comments	Defects
	1	2.7	806	237	125/124	Flat, slightly convex, right side had higher buildup. Did not impinge sidewall due to contact tip	3
Groove	2	3	802	215	117/117	Slightly convex, great appearance	1
3	3	3.5	827	193	119/117	Flat, great appearance	3
	4	4	805	166	119/119	Slightly convex, great appearance	3
	1	2.5	1000	268	127/126	Flat, great appearance	2, 3
Groove	2	2.8	1007	243	130/127	Flat, great appearance	1, 3
6	3	3.1	1017	224	130/131	Great appearance, no undercut	3
	4	3.4	1018	206	134/132	Great appearance, no undercut	3
	1	2.5	940	250	125/126	Did not deposit on one sidewall	2, 3
Groove	2	2.8	983	236	133/131	Flat, great appearance	3
7	3	3.1	993	218	133/132	Great, flat appearance, slight undercut	3
	4	3.4	1000	201	136/134	Great, flat appearance	3
	1	2.65	1111	281	115/115	Deposited more on right side. flat, slightly convex	2, 3
	2	2.95	1106	253	123/120	Flat, great appearance	None
Groove 8	3	3.35	1117	228	134/132	Flat, great appearance. Undercut on right side	3
0	4	3.65	1111	209	145/145	Flat, great appearance. Undercut on right side	None
	5	3.85	1111	199	156/156	Flat, great appearance, slight undercut	None

 Table 27: Summary of Welding Passes for each Groove Trial (Groove 5-8)

1 = Solidification cracking 2 = Shrinkage porosity 3 = Lack of fusion