Hybrid Laser/arc Welding of Difficult-to-Weld Thick Steel Plates in Different Joint Configurations: Issues and Resolutions

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HYBRID LASER/ARC WELDING OF DIFFICULT-TO-WELD THICK STEELS PLATES IN DIFFERENT JOINT CONFIGURATIONS: ISSUES AND RESOLUTIONS

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HYBRID LASER/ARC WELDING OF DIFFICULT-TO-WELD THICK STEELS PLATES
IN DIFFERENT JOINT CONFIGURATIONS: ISSUES AND RESOLUTIONS

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by

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DEDICATION

This dissertation is dedicated to my lovely Mother (Roya Mirroknian), and my dearest father (Mohammad Taghi Yazdian) whose love and guidance are with me in whatever I pursue.
Hybrid Laser/arc Welding of Difficult-to-Weld Thick Steel Plates in Different Joint Configurations: Issues and Resolutions

Advisor: Professor Radovan Kovacevic

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Difficult-to-weld steels are ferrous alloys that are characterized by a low thermal conductivity, and large thermal expansion coefficient. These intrinsic features contribute to a high level of distortion and cracking susceptibility during joining of these types of steels. In an effort to address the issues associated with difficult-to-weld steels, highly concentrated beam spots like electron and laser beam welding were developed. Usage of tightly focused heat sources have been accompanied by several challenges. An extremely precise fit-up requirement was considered as the most significant issue corresponding to application of either laser or electron beam. Recently, it was found that the combination of arc and laser in close proximity could lead to development of a new technology called hybrid laser/arc welding (HLAW). Simultaneously, HLAW takes advantage of arc-induced wider molten pool and laser-induced deeper penetration capabilities. Both of which would be mutually exclusive. However, application of this technology in the case of real usage is limited by the high complexity of the process and existence of numerous variables. Furthermore, humping, formation of porosity, and weak corrosion resistance of the welding region have been recognized as the principal issues concerning HLAW joints. In this regard, the major objective of the current Ph.D. research project was to comprehend the difficulties surrounding HLAW of difficult-to-weld steels as well as implementation of associated practical guidelines to make the HLAW process more practical for widespread industrial applications.
High-strength quenched and tempered steels (HSQTSs) are one of the big group of alloyed steels that are vastly used in shipbuilding industry. These types of steels are typically prone to cold cracking due to formation of the hard-martensitic phase during solidification. The common way to weld this type of alloyed steel is to use preheating process integrated with the multi-pass gas metal arc welding. However, that method generated a large level of thermally-induced tensile residual stress and distortion coupled with a large softened area in the vicinity of the fusion zone. Addressing the above issues, a feasibility study of hybrid laser/arc welding of 8-mm-thick HSQTS in butt and T-joint configurations was conducted experimentally and numerically. This chapter focused on developing the processing parameters to produce a sound full-penetrated weld in a single pass. Characterization techniques, including microstructural analysis, hardness, and tensile testing were employed to get a better understanding of the quality and mechanical integrity of the weld. Furthermore, an experimentally-calibrated thermomechanical simulation using SYSWELD commercial software, was introduced to analyze the weld-caused residual stress fields and distortion.

Austenitic stainless steel (ASS) is another subcategory of difficult-to-weld steels due to low thermal conductivity and large thermal expansion coefficient. In practice, stainless steel is majorly found and applied in circular hollow sections, thereby making ASS welding necessary. Thus, orbital welding of ASS pipes are an inseparable part of any manufacturing process. Currently, multi-pass arc-based welding process is a widely developed process for girth-welded pipe joints. However, a large heat affected zone (HAZ) is the main concern regarding the welds that are produced by arc-based processes. It was found that excessively wide HAZ makes additional heat treatment process necessary to retain the microstructure and required surface properties. Addressing the above problem, the feasibility of orbital welding of AISI304L stainless steel tubes by HLAW in two relative positions of the laser with respect to the arc to attain a free pore weld was studied. The effect of welding speed on the formation of porosity
were also investigated. To obtain a rough estimation about expansion of HAZ throughout the base metal the heat distribution in the welding region was evaluated through thermal simulation using commercial ANSYS code. The micro-hardness and the tensile strength of joints under optimal conditions were analyzed to reveal the relation between microstructural and mechanical attributes.

The weld-induced heterogenous microstructure of austenitic stainless welds is generally considered to be the weakest area when exposed to a corrosive environment. This conclusion is drawn because of micro-segregation of the main alloying elements such as Cr and Ni during the solidification. One of the most important attributes of HLAW process is the addition of filler metal that gives the weld a remarkable advantage over autogenous laser welding. Based on this advantage, it is expected that the proper filler wire will maintain the corrosion resistance of the fusion zone. Accordingly, the effect of wire type on microstructural alteration and corrosion resistance of AISI304 stainless steel joints produced by HLAW was studied. To gain a deeper understanding, an integrated characterization of fusion zone using the scanning electron microscope (SEM), X-ray diffraction (XRD), and cyclic potentiodynamic polarization analysis (CPPA) was conducted to quantitatively evaluate the corrosion resistance and explain the formation and growth of corrosion pit mechanisms.
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CHAPTER 1

REVIEW OF WELDING TECHNIQUES OF DIFFICULT-TO-WELD STEELS

1.1. Introduction

There are various steel grades in different industrial applications that can be regarded as difficult to weld. The inherent properties such as low thermal conductivity and large expansion coefficients highly complicate the fusion joining of these types of steels and make them sensitive to the formation of welding-induced distortion as well as cold and hot cracking [1,2]. Table 1.1 presents the physical and mechanical properties of grades of steels that have been widely used in industry. As can be seen, the addition of alloying elements for strengthening and promotion of mechanical properties changed the thermal properties of those steels. For example, the austenitic stainless steel group possesses the greatest thermal expansion coefficient and the lowest thermal conductivity.

Table 1.1 Mechanical and thermal properties of the most applicable steels in industrial applications [1-5].

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>Physical and mechanical properties</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Thermal conductivity (W/mK)</td>
</tr>
<tr>
<td>Mild low carbon steels (AISI 1018)</td>
<td>52</td>
</tr>
<tr>
<td>High-strength low alloyed steels (HSLA-100)</td>
<td>27</td>
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<tr>
<td>High-strength quenched and tempered steels</td>
<td>47</td>
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<td>Austenitic stainless steels (AISI304L)</td>
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</table>

During the cooling cycle of the welding process, solidification shrinkage and solid-state contraction of the weld and adjacent base metal are prevented by the regions that are farther...
from the welding area. As a result, residual tensile stresses exist in the weld metal. Increasing the thermal expansion coefficient and reducing the thermal conductivity significantly scales up the tensile residual stress field throughout the welding region, thereby intensifying the distortion [6]. Moreover, formation of hard phases such as martensitic microstructure particularly for high-strength steels increases the weld propensity for cold cracking [7]. The excessively large heat input of traditional arc-based welding processes expose large portions of the workpiece to thermal cycles, and the residual tensile stress fields expand through the base metal. Accordingly, an effective way to reduce the thermally-caused side effects of the fusion joining process is the usage of heat sources equipped with very a high-power density. Consequently, the total heat input per unit length of the weld is much smaller than that in the arc weld, resulting in a very narrow heat-affected zone and little distortion. This chapter discusses current fusion welding techniques and recent developments for the joining of difficult-to-weld thick steels. Furthermore, the most current research work on the investigation of laser-based welding techniques are reviewed. Some solutions to the problems in joining thick steels are presented.

1.2. Multi-pass arc-based welding techniques

In the manufacturing industry, arc welding is considered as a diversified group of welding methods and has been widely implemented in different industrial sectors. It covers all techniques where metals are melted by the heat of an electric arc. These processes are differentiated by some of the benefits that they receive from traditional mechanical joining such as flexibility of design, cost savings and weight reduction [8]. However, for thick materials with low penetrability from the arc, one weld pass may not be enough to attain full penetration and ensure a strong joint. Over a long time, multiple-pass arc welding is a common term that has been bound to thick structural parts. Generally, there are three stages in multiple-pass arc welding [9]. The root pass where a full penetration weld occurs is the first and most
important step. The intermediate weld stage constitutes filler passes that can range one to several hundred passes, depending on the joint size. Eventually, the final stage forms a wide convex weld face on the bead (Fig. 1.1). It should be noted that the weld pass quality and microstructural evolution during multiple-pass welding are key factors that control the mechanical integrity of the final joint. Therefore, between the welding passes, each welding pass should be inspected for defects to ensure that a sound layer is obtained. Another important aspect of the multiple-pass arc welding is the edge preparation that makes the extra machining process necessary. In all edge preparation methods, regardless of groove shape and bevel angle, the primary purpose is to provide the welding torch with sufficient space to reach the proximity of the joint bottom.

![Figure 1.1 A schematic of multiple-pass arc welding passes [9].](image)

At present, a great amount of effort has been expanded to improve the efficiency of multiple-pass arc welding. Mounting the arc torch on robotic arms integrated by computer numerical control (CNC) machines have made this process more flexible and fast enough for mass production [10]. To increase the deposition rate, Chen et al [11] established a double-sided double arc welding (DSADW) with two robots to weld 35-mm thick HSLA steel. It was found that the full penetration weld with the lowest distortion would be possible by preparation of a double V-groove shape as well as proper welding sequences. Fig. 1.2 showed a typical
weld bead cross-section of the welded joint by DSADW. As can be seen, more than 20 welding passes had to be done to obtain a full-penetrated weld.

![Image](image.png)

Figure 1.2 (a) The real setup of double-sided double GMAW to weld 35-mm thick HSLA steel and (b) a typical macrograph of weld cross-section produced by multiple pass arc welding [11].

Among arc-based welding methods, a submerged arc welding (SAW) has come into great attention over other welding processes in thick-walled pipe production. The SAW features, such as a high deposition rate and deep penetration are the reasons for this attention. In the SAW process, an intense arc is generated between a consumable electrode and a work piece. Arc is held in place along the weld puddle by a flux powder [8]. Recently, the tandem submerged arc is equipped with multiple electrodes to improve the efficiency of the welding process. This process has been gained a specific status to join thick tubular sections. Moeinifar et al. [12] found that the usage of a four-wire tandem submerged arc welding process succeeded in generating a consistent weld bead for 13-mm thick X80 micro alloyed steel. However, reduction in the fracture toughness in the heat affected zone was observed as the main obstacle of this process, imposing a post heat treatment process to remove the weld-induced brittleness. Actually, due to the application of a couple of electrodes, increased overall welding current and voltage leads to an extremely large heat input and reduces the cooling rate remarkably.
Formation of large residual austenite grains coupled with martensite-austenite (M-A) constituents (Fig. 1.3) were characterized as the localized brittle zones [13]. It was proved that in the presence of an external load, such regions have an accelerating effect on nucleation and propagation of micro cracks. These cracks result in a weld with poor mechanical properties. It was observed that if the width of the HAZ, containing large retained austenite grains and M-A constituents was more than 1-mm, it would be detrimental to the weld. Evidently, fracture toughness can be improved by narrowing this area [14].

![Figure 1.3](image)

Figure 1.3 (a) The typical microstructure of HAZ, (b) and (c) SEM pictures of fracture surface of Charpy impact test corresponding to the weldment of 14-mm X100 pipeline steel produced by tandem submerged arc welding [13,14].

In context of the tandem submerged arc welding process, it was found that additional cold wire feeding into the weld pool increased the deposition. The cold wire feeding also moderated the heat input introduced to the weld. In fact, the cold wire feeding consumed the energy of the trailing electrode as the wire melted into the weld puddle. As a result, a higher cooling rate along with lower heat input per mass of deposited material is expected [12]. As
shown by Fig. 1.4, the width of HAZ was reduced significantly. However, the size of the fusion zone was still excessively large. The coarse dendrite that was formed within the middle of the FZ had an adverse effect on the mechanical integrity of the joint. It was found that the absorbed energy of the joint during the Charpy test was reduced notably with respect to the parent metal.

Figure 1.4 (a) The arrangement of torches and cold wire for cold wire assisted with tandem SAW, (b) optical macrograph of weld cross-section and (c) the width of HAZ zone [12]

In all arc-based welding methods, the main issue is the formation of an extremely wide HAZ that negatively influences the tensile properties of the weld (Fig. 1.5.) [15]. Fig. 1.6 shows the typical fracture location of a tensile sample that was extracted from the multiple-pass weld of 7-mm thick AISI316 stainless steel [16]. To retain the HAZ microstructure, the post weld heat treatment is considered as a mandatory post-processing stage to compensate for the
thermally-induced side effects of the welding process. Post heat-treatment is more critical for large structures and multiple-pass welding where there is no control over the microstructures of intermediate passes. Due to overlapping between two successive welding passes, the coarsening and formation of undesirable phases are inevitable. The undesirable phases make the microstructural modification crucial.

**Figure 1.5** An optical macrograph of weld cross-section produced by tandem SAW process [15].

**Figure 1.6** The fracture location of tensile coupons of multi-pass weldment of 7-mm AISI 316 stainless steel [16].

Based on the above discussion regarding arc-based welding methods, following are some of the main shortcomings that limit the application of those processes to join thick-gauged sections of difficult-to-weld steels:
1. Due to an inherently low-diffusive arc, joining thick materials are barely possible with a single-pass arc welding process even with an additional torch under a tandem condition.

2. The difficulty to control weld passes in terms of quality and consistency requires an extra cost to enable robust on-line monitoring setups that examine the correctness of the weld.

3. The necessity of designing preset grooves that provide required space for the arc torch and access for the through-thickness weld passes makes an extra machining process inevitable. This extra process negatively impacts industrial productivity.

4. The excessively large high heat input given by arc forms the extremely wide HAZ surrounding the FZ, resulting in poor mechanical properties. The highly expanded thickness-through residual stress induces undesirable distortion. This distortion is another negative aftermath that requires an additional post heat treatment process and correcting methods. In turn, a significant amount of time and money are sacrificed.

5. Especially for difficult-to-weld steels, heat input should be considered as a critical matter. Concerning the intrinsic features of these materials, any extra heat input can dramatically exacerbate issues such as thermally-induced distortion and cold cracking.

1.3. Friction stir welding technology

One of the joining methods that has been commercially attractive to weld thick materials is the friction stir welding (FSW) process. As a basic definition, this solid state joining technique includes the rotating motion of a non-consumable tool that generates a softened plasticized zone. The transverse motion of the tool along the weld trajectory path produces a mechanically-durable joint [17]. Exclusively, this process was developed for grades of steels that are vulnerable to solidification cracking and could become sensitized at elevated temperature holds. Additionally, this process promises to be an effective method of suppressing
sensitization in welds because of its low heat input, and eliminates the issues that transpire from melting and solidification. It is well-known that the FSW process generates high quality joints with the finer homogenous microstructure and superior mechanical properties relative to the traditional arc welding processes. The HAZ adjacent to the weld is confined to a narrow band as well [18].

Quite an amount of research has been conducted to weld thick stainless steels. Han et al. [19] investigated the microstructural evolution and mechanical properties of friction stir welds of 6-mm ferritic-based stainless steel. It was observed that using a composite tool with a high-volume percentage of CBN in W-Re matrix incorporated with proper rotational and welding speeds could produce high-quality welds. It was revealed that the welds had a hardness much higher than that of the base metal (Fig. 1.7) Extreme levels of localized plastic deformation as well as dynamic recrystallization were taken into account as the main mechanisms of microstructural refinement, resulting in stronger weld. Another possible area where FSW was claimed to be beneficial is in the joining of duplex stainless steels where it is critical to maintain the approximately equal austenite and ferrite phase fractions. Gioajao et al. [20] conducted orbital FSW to join an 8-mm thick duplex stainless steel tube. It was found that in-situ thermally-induced dynamic recrystallization during the process played a crucial role to maintain the homogeneity of the microstructure of the welding region and keep the integrity of the joint. It was shown that tensile coupons failed at lower hardness sites, located in the base metal (Fig. 1.8).
Figure 1.7 (a) An example of cross-section of FSW joint at the optimized welding condition and (b) micro-hardness profile of the FSW joint in the centerline [19].

Figure 1.8 (a) The schematic of arrangement of FSW tool and pipe and (b) fractured tensile specimen of duplex stainless steel joint [20]
Unlike the mentioned benefits of the FSW process, there are critical challenges that complicate usage of that process to join thick steels. The main problems are as follows:

1. The high melting point of steel, that causes tool wear is identified as a serious challenge. Fig. 1.9 shows the typical temperature dependence of the yield strength of Al alloy compared with that of steel. Evidently, a harsh torment that an FSW tool would have to go through in the case of steel would be much larger than that for Al. This condition is particularly true when temperature reaches in excess of 700-800°C. Over that temperature range, steel must be sufficiently plasticized to allow the material flow to generate a sound weld [21]. Although a group of novel ceramic-based composites was successfully examined, the fabrication cost of such material is not economical particularly for mass production.

2. The order of magnitude for applied welding speed in this process is comparable to that of traditional arc-based welding process. Increasing the welding speed is associated with larger heat generation and a great risk of abnormal temperature elevation of the base metal. This result sacrifices the strength and integrity of the joint. Besides, higher speed results in greater wear rate of the FSW tool. Thus, to ensure that the FSW tool lasts longer, the welding speed should be restricted to the range of 1-2 mm/s. Worth noting, this range is much lower than that of robotized arc-based welding process (5-10 mm/s). Consequently, application of this process for thick and long geometrical structures of steel is highly limited. The low efficiency and productivity originates form lagging characteristics that are inherent to the FSW.

3. It was shown that due to quite a high number of processing parameters that are involved in FSW process, the high quality repeatable joints could be barely obtained in a wide range of parameter combinations. Thus, identification of the window for processing variables to achieve an acceptable level of robustness requires a significant amount of time and manufacturing costs. Furthermore, in the robotic FSW, large structure assemblies tend to have multiple setups
and significant variation of parts. Variations such as unpredictable gaps, mismatch, and tool deviation from the weld joint intensify the complexity of the FSW process.

![Figure 1.9 Typical temperature dependence of hot strength of Al alloys and steels [21].](image)

### 1.4.1 Electron beam welding (EBW)

As discussed above, an excessively large heat input received from the arc coupled with its low penetrability are the fundamental reasons for subsequent issues. These issues include a very wide HAZ, lack of penetration, highly distorted joints, and necessity of further post heat treatments as well as correction methods to minimize the side effects. To address the above issues, application of high energy density processes like electron beam can offer significant technological benefits for full penetration of thick gauge sections of steels. EBW is famous for its excellent penetration ability. An extremely high-power density in the order of $10^7$ W cm$^{-2}$ at the small focal point with a diameter range the in the order of 0.3-0.8 mm enables the electron beam to vaporize the material. As results, a deep-penetrating cylindrical cavity called key-hole is produced [8]. The electron beam-induced key-hole can achieve the deep penetration in a single pass. Low heat input characteristics of EBW would produce a narrow FZ with a narrow HAZ, including minimal shrinkage and distortion. Regardless of the thickness of the work piece, no edge preparation is needed. This efficiency is as another advantage of EBW over the traditional arc welding process [22]. Currently, EBW has been developed to weld thick difficult to weld steels. Yoon et al. [23] employed the single pass electron beam welding process to join
a 7-mm thick high-strength steel. Fig. 1.10 shows a typical microstructure of the optimized E-beam welded joint. As can be observed, the electron beam-induced keyhole generated a narrow fusion zone with a minimal HAZ. It was found that tensile specimens that were prepared from the EBW joints would not show any noticeable property deterioration owing to localized hardening in the HAZ and FZ so that all of tensile coupons were fractured at the base metal as shown in Fig. 1.11.

![Figure 1.10 Cross-section of the weld bead of a joint produced by EBW](image)

**Figure 1.10** Cross-section of the weld bead of a joint produced by EBW [23].

![Figure 1.11 (a) Micro-hardness profile across the EBW joint and (b) fracture location of tensile coupons](image)

**Figure 1.11** (a) Micro-hardness profile across the EBW joint and (b) fracture location of tensile coupons [23].
One of the most noticeable concerns about the quality of the joint that is produced by EBW is non-uniformity of mechanical properties through the weld depth (Fig. 1.12). Alali et al. [24] reported that an upheaval in microstructural features from the weld face side toward the root part was the main reason. As observed in Fig. 1.13 a relatively slow cooling rate of the top area result in axial and columnar grain structure, whereas a faster cooling rate induces finer columnar equiaxed grains at the root region of the weld. With regard to welding of austenitic stainless steels, Krasnorutsky et al. [25] pointed out that the EBW resulted in excessive fertilization due to the rapid cooling rate, which, in turn negatively affected the mechanical properties and corrosion resistance of the welded material. Ku et al. [26] also found that the EB weld of stainless steel exhibited a lower impact strength than that of the base metal. This lower impact strength was attributed to the loss of nitrogen during welding process. In addition to the high cooling rate of EBW, the following drawbacks impedes this process to achieve a dispensable status for joining thick, difficult-to-weld steels:

1. The entire process was performed in a vacuum. Therefore, the dimensions of the vacuum chamber impose a limitation on the size of the part to be handled. This limitation was particularly true for long geometrical components and large tubes.

2. With regard to magnetic grade steels, electron beam is readily deflected by a localized magnetic field that is formed around the surface of the work piece. Alignment correction of the beam can burden the entire process with additional modules that cannot be justified economically.

3. Precise fit-up requirement is another restriction that is associated with configurations where the zero-gap weld is difficult to achieve. Thus, an extra expense is required to design a specific clamping condition to achieve the zero-gap weld. The necessity for rigid clamping is because the electron beam’s gap-bridging ability is poor relative to other heat sources such as the arc.
Figure 1.12 UTS vs penetration for the joint produced by EBW [24]

Figure 1.13 Macro and through-thickness microstructure of EBW joint weld [24]

1.4.2. Laser beam welding process

Over the past three decades, the laser welding and laser-assisted joining processes have gained good acceptance to meet joining requirements in several industrial areas including automobile, medical, electronic, aviation, energy harvesting, shipbuilding, oil pipelines, etc. Based on the statistical survey performed in the year 2006, among all the laser materials processing application segments sold worldwide (such as cutting, drilling, surface treatment, micro processing), laser welding has had an exceptional status. Laser welding constituted 12%
of the entire expenses that have been invested in the field of laser applications [27]. To achieve a better picture, Fig. 1.14 compares the laser welding with conventional welding processes in terms of the power density and consequent effects on weld features such as the depth of penetration, and heat affected zone size. Additionally, among all affected parameters by heat source, welding speed is the most substantial parameter describing the process productivity [28].

**Figure 1.14 Effect of power density on welding process and function [28].**

Like any other thermal fusion welding process, the laser welding process should be first and foremost considered in terms of laser generation sources. To date, high power lasers have witnessed lots of interesting improvements in terms of laser wavelength and beam quality. CO$_2$ lasers have had a commendable status in industry for welding applications for a long time [29]. However, due to some of its intrinsic limitations like controlling the laser source’s stabilization, laser beam quality, focus dimension, and highly absorptive laser-induced plasma plume, lots of works have been performed to modify its drawbacks and develop the CO$_2$ laser joining process applications [30]. The next generation of the lasers that have been recognized as reliable sources are solid-state sources like the lamp pumped ND: YAG laser. Lasers are generally characterized by three main parameters, their wavelength, beam quality, and power
output. They are also looked at under other practical considerations such as efficiency, compactness, ease of maneuverability and delivery, and cost management. These practical parameters are followed more closely by the industry of laser manufacturing and can be taken account as the deciding factors for the purchase of a laser. The important features of common industrial lasers for welding are summarized in Table 1.2. The pros and cons of these types of laser are also illustrated in Table 1.3. [31].

### Table 1.2 Common industrial laser for welding [31].

<table>
<thead>
<tr>
<th>Parameters</th>
<th>CO₂ laser</th>
<th>Nd:YAG laser</th>
<th>Diode laser</th>
<th>Fiber laser</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wavelength (µm)</td>
<td>10.06</td>
<td>1.06</td>
<td>0.8-0.98</td>
<td>1.07</td>
</tr>
<tr>
<td>Efficiency</td>
<td>5-20%</td>
<td>10-20%</td>
<td>30-60%</td>
<td>10-30%</td>
</tr>
<tr>
<td>Output power CW</td>
<td>Up to 20kW</td>
<td>Up to 16kW</td>
<td>Up to 14kW</td>
<td>Up to 10kW</td>
</tr>
<tr>
<td>Beam quality factor</td>
<td>3-5</td>
<td>0.4-20</td>
<td>10-100</td>
<td>0.3-4</td>
</tr>
<tr>
<td>Fiber delivery</td>
<td>Not possible</td>
<td>Possible</td>
<td>Possible</td>
<td>possible</td>
</tr>
<tr>
<td>Maintenance period (hrs)</td>
<td>2000</td>
<td>200</td>
<td>100000</td>
<td>100000</td>
</tr>
<tr>
<td>Price ($/W)</td>
<td>35-120</td>
<td>100-120</td>
<td>60</td>
<td>100</td>
</tr>
</tbody>
</table>

### Table 1.3 Advantages and disadvantages of different laser types [31].

<table>
<thead>
<tr>
<th>Laser type</th>
<th>Advantages</th>
<th>disadvantages</th>
</tr>
</thead>
<tbody>
<tr>
<td>CO₂ Laser</td>
<td>Simple design, reliable and compact</td>
<td>Low absorption in metals, sensitive to laser induced plasma specifically for thick welding unable to deliver by fiber</td>
</tr>
<tr>
<td>ND-YAG laser</td>
<td>Higher absorption than CO₂ laser, flexible beam delivery, less sensitive to laser induced plasma</td>
<td>Limited life of lamps Poor beam quality at higher laser powers</td>
</tr>
<tr>
<td>Fiber laser</td>
<td>Compact, excellent beam quality, high efficiency, air-cooled and maintenance-free operation, long lifetime, scalable to higher power, flexible beam delivery</td>
<td>High capital and maintenance cost</td>
</tr>
<tr>
<td>Diode laser</td>
<td>Most efficient, most compact, high reliability, low running cost, easy integration with CNC, flexible beam delivery</td>
<td>Poor beam quality</td>
</tr>
</tbody>
</table>
Typical benefits of using laser as a heat source are as follows:

(1) Accessibility of high power densities. For instance, in the keyhole mode of laser welding, power densities higher than $10^6 \text{ W/cm}^2$ can be obtained. However, conventional arc welding processes are able to provide power densities in the limited range of $10^4 \text{ W/cm}^2$.

(2) Deep and narrow fusion zone with depth-to-width aspect ratios of 10:1 enable welding of thick materials with a single pass. This feature eliminates multi-pass welding issues and facilitates the process with one-sided accessibility.

(3) Ease of shielding of the molten pool. The shielding can be facilitated because the molten pool size can be reduced significantly.

(4) Elimination of pre-groove making that is considered as an inseparable part of conventional arc welding. It implies that laser welding is accompanied by simple joint design that results in high productivity.

(5) Ease of combining laser equipment with multi-functional robots offers a high welding speed as well as an amenability to automation

(6) Due to the highly concentrated and precise heat input, a small heat affected zone (HAZ) incorporated with lower thermally-induced distortion is generated.

(7) Due to application of a wide variety of focusing optics, laser welding not only offers remote control welding feasibility especially for inaccessible places, but also non-contact zero-force processing can be pragmatic [28].

Owing to the above advantages, application of new generation of high power lasers to weld thick steels has been the scope of many research projects. Sokolov et al. [32] studied the feasibility of the autogenous laser welding of 20-mm thick S355 steel using a high power disk laser. It was found that increasing the welding speed resulted in decreasing the average HAZ width and increasing the average weld hardness. In spite of generating full penetration, sagging defects at the face side were produced for all welds (Fig. 1.15).
Although, weld penetration will be enhanced by raising the laser power density, higher power laser under current technology is very costly. In addition, it was reported that by increasing the laser power, the laser-induced plasma volume above the keyhole grew considerably. The result was a more unstable plasma plume [33]. A stronger and unstable plasma plume tremendously weakens the energy absorption. It has been proved that in the case of high power laser welding, most of the laser energy was reflected, absorbed, scattered, and defocused by plasma plume. Consequently, efficiency of the laser absorption by metallic molten wall around the keyhole was reduced significantly [34].

![Figure 1.15 Cross-section macrographs of the welds that were produced by autogenous laser welding under different welding speeds ((a) 1.2 m/min and (b) 2.4 m/min)) [32](image)

There have been several studies to increase depth of penetration and stability of the molten pool by controlling the plasma during laser welding. Wang et al. [35] analyzed the influence of the side assisting gas on the laser-induced plasma behavior and energy transmission both theoretically and experimentally. It was observed that increasing the gas flow rate reduced the required laser power to achieve full penetration (Fig. 1.16). To determine the reason behind this result, spectroscopy analysis was conducted to monitor spectral lines coming out from the plasma plume during the welding process. Fig. 1.17 presents plasma spectrums.
under different gas flow rates. As the flow rate increased, intensity of the spectrum decreased. This result implied that shrinking of volume of laser-induced plasma occurred above the key-hole.

![Butt joint weld of galvanized steel at two different gas flow rate and laser power](image)

**Figure 1.16** Butt joint weld of galvanized steel at two different gas flow rate and laser power (a) 1800W and 0.1 m³/h and (b) 1200W and 1 m³/h [35].

![Plasma spectrums under different gas flow rates](image)

**Figure 1.17** Plasma spectrums under different gas flow rates [35].

To attain a better comprehension of the shielding gas effect on the plasma plume behavior, plasma in two different states of gas flow rate was monitored by CCD camera during the welding process (Fig. 1.18). During the condition when the flow gas rate was low, two parts of plasma, the brighter and smaller one called the residual plasma and the darker and larger one
called the diffusive plasma could be clearly recognized. The residual plasma that exists at the outlet of the keyhole constituted of a high-density metal vapor that is highly ionized. Within that part of the plume, the temperature and pressure are very high. It was reported that this type of plasma is not able to be blown away completely as so-called residual plasma. On the other hand, the diffusive plasma was assumed to be formed by expansion of residual plasma. For this type of plasma, the temperature and internal pressure is much lower than for residual plasma. In the case of the higher shielding gas flow rate, the diffusive plasma was almost absent. When the laser beam passes throughout the two parts of plasma, the absorbed amount of energy is more than the case where just the residual plasma is present. Therefore, in the case of a smaller shielding gas flow rate, the laser energy that was absorbed by the molten pool surrounding the keyhole wall was reduced significantly. In addition, it was found that diffusive plasma increases the degree of laser beam defocusing. The larger volume of diffusive plasma results in a more serious defocusing effect. In turn, this defocusing effect reduces the laser power density at the focal point [27]. The schematic illustration of the deflected beam after passing through the plasma with or without diffusive plasma is shown in Fig. 1.19.

![Figure 1.18 Photos of plasma captured by CCD camera obtained at (a) low flow rate and (b) high flow rate of shielding gas [35].](image-url)
Geometrical calculations showed that the deflection radius ($\Delta R$) was increased in presence of the diffusive plasma. According to the proposed model by Wang [35], diffusive plasma not only absorbed the laser energy significantly but also seriously deviated the focal point of the laser beam. Thereby, the efficiency of laser energy transmission was reduced. Thus, it can be explained that strong shielding gas could efficiently blow away the diffusive plasma and maximize efficiency of the laser energy transmission.

As discussed, formation of a strong and unstable plasma plume above the keyhole especially at higher laser power and lower welding speed reduces the efficiency of laser absorption of the target material that is being melted. Recently, it was shown that laser welding under sub-atmospheric pressure increases the depth of penetration considerably. A typical

Figure 1.19 Deflection of laser transmission direction by either diffusive or residual plasma [35].
prototype of a local vacuum chamber facility that was developed by Lue et al. [36] is presented in Fig. 1.20.

![Diagram of laser welding setup](image)

**Figure 1.20. The prototype setup used for laser welding in the partial vacuum [36].**

Application of partial vacuum could reduce the plasma plume volume above the keyhole. Furthermore, the boiling point of metals is highly pressure dependent. Based on the Calasius-Clapeyron equation, the melting point of industrial alloys is almost independent of pressure difference due to a small change of the volume during melting. However, the boiling temperature experiences a significant reduction in the order of 1000K when the ambient pressure range is between 1000 and 0.1 hPa (Fig. 1.21) [37]. This reduction implies that less energy is required to vaporize the base metal to form the keyhole. In addition, a lower temperature gradient between the almost unaffected melting point and reduced boiling point makes the keyhole more stabilized. Thereby, the amount of molten material around the keyhole is thinner and the amount of pressure being applied on the keyhole is reduced. Moreover, due
to the dropped pressure above the keyhole, the density of metal vapor decreases, and there is less condensation of the metal particles above the keyhole. The lower accumulation of metal particles could be effective in such a way that it would change the refractive index of plasma plume and reduce the absorption and scattering of the laser beam. Relevant studies in this field showed that under partial vacuum conditions, the laser energy is mainly absorbed by Fresnel reflection. In the atmospheric case, the laser beam is absorbed by a combination of Fresnel and inverse bremsstrahlung absorption of plasma. It was reported that in the case of the sub-atmospheric laser welding process, almost no spattering on the top side of the weld seam occurred and the process was very stable [38]. Also, a lower ambient pressure makes degassing the weld pool much easier and leads to very low rates of porosity.

![Figure 1.21](image)

**Figure 1.21 Change of the boiling and melting point of iron with respect to ambient pressure [37].**

Chen et al [39] studied the effect of sub-atmospheric pressure on the depth of penetration for joints of 40-mm thick high-Mn low alloy steels produced by laser welding. It was found that a threshold pressure in the order of 10-20 kPa enhanced markedly the weld penetration depth. The plasma plumes as well as cross-sections of bead-on-plate welding trials under different ambient pressures are presented in Fig. 22.
Figure 1.22 Cross sections as well as plasma plume images obtained by CCD at the laser power of 8kW [39]

The partial-vacuum assisted laser welding could give a deep penetrated weld with high quality. However, that process may lose its competitive benefit over the EBW. Access to an ultra-low pressure of vacuum requires evacuation of the chamber where the workpiece is located. This evacuation may be time-consuming and seriously influence the welding efficiency.

Another important concern regarding the autogenous laser welding of thick steels is the formation of cold cracks. A couple of reasons account for this phenomenon. Autogenous laser welding is intrinsically associated with a high cooling rate. Thus, the formation of brittle phases during the cooling cycle of this process is an inevitable result. Another aspect of autogenous laser welding is formation of a fusion zone with a high depth to width ratio that increases the risk of hot cracking [40]. Besides, the solidification structure of the weld metal influences in particular the occurrence of solidification cracking. Formation of columnar grains perpendicular to the welding direction that meet at the center line was taken account as the typical microstructural feature of autogenous laser welding under high speed. This boundary generally has a limited ability to accommodate liquid films and strain, thereby causing the hot cracks [41]. Particularly, hot cracking is well known among autogenous laser welding of stainless steels because of various solidification modes that could be experienced by the molten
pool during rapid cooling. Bollinghaus et al. [42] found that the hot crack located within the center of plate thickness for the overlap AISI 201 stainless steel welds produced with a CO$_2$-laser (Fig. 1.23). The effect of laser focal position on the formation of hot cracking of 12-mm C-Mn steel weldments was studied by Gittos et al. [43]. A similar feature was observed for all welding conditions (Fig. 1.24) where the laser beam focal position did not appear to have a marked effect on the through-thickness extent of cracking.

Figure 1.23 Overlap joint cross section of AISI201 produced by high power autogenous laser welding [2].

Figure 01.24 Effect of laser beam focal position on hot crack propensity of the welds of 12-mm C-Mn steels. a) -4-mm b) 0-mm c) +5-mm and d) +8-mm [43].
Typical challenges of autogenous laser welding to join thick gauged sections of difficult-to-weld steels are as follow:

1. Similar to EBW, the high cooling rate of the laser welding process forms unwanted brittle phases throughout the FZ of high strength steel joints and increases the risk of cold cracking.

2. The extremely precise fit-up requirement that arises from a very restricted gap tolerance is considered as the major technical issue. The current technical means are unable to comply with the demands for accurate part positioning. This issue is evident particularly when the weld geometries are long and the joint gap changes because of heat expansion or slight distortion during welding.

3. The sensitivity of the laser joints to formation of large pores under a low welding speed regimen due to unstable and collapsed keyhole.

4. During autogenous laser welding, no filler metal additions are made, which significantly reduces the range of alloys that can be laser welded. For stainless steels, filler metal additions are required to retain the microstructure and compensate for the sacrificed corrosion resistance of the FZ.

1.4.3. Cold wire assisted laser welding process

As discussed above, the autogenous laser welding is not able to deposit any extra material at the welding site, and the unfilled area at the face side appears as a defect. In addition, stringent fit-up tolerances are required during welding process. Such inabilities result in a critical problem while joining thicker metal plates with larger gaps between them. Thus, the joint will exhibit serious defects that threaten the integrity of the structural part. To reduce the limitations that raise from autogenous laser welding, a laser welding assisted with cold wire has emerged (LWACW). This usage of filler material can offer a bridge between costs and weld quality. Modification of the chemical composition of the fusion zone receives the most
important benefit through the addition of filler metal that improves the weld mechanical properties [44]. Multiples studies have used cold wire incorporated with laser welding to improve bridging capabilities in the simple fit-up joint [45-47]. In regard to the position of the wire tip with respect to the laser beam, Yong et al. [45] reported that when the laser beam was focused on the center of the wire, the filler wire could be melted properly, though, the wire was wandering at a certain degree (within 1 mm). Kong and his colleagues. [46] successfully generated a sound weld of 7-mm thick high-strength steels plats in butt-joint configuration, using cold wire assisted laser welding. Atabaki et al. [47] compared autogenous laser welding and cold wire assisted laser welding to assess the welding capability of thick structural steel in a horizontal-joint configuration. It was found that the additional wire could remove the unifilled-area at the top and back side of the weld. However, formation of the reinforcement at the top side of the bead was not observed for the cross-sections of the welds obtained by LWACW (Fig. 1.25). It was revealed that the keyhole-induced capillary effect of the molten pool was the main reason that caused a concave-shaped bead at the face side of the weld bead [48]. Such concavity could be considered as a stress-concentrated area that would reduce fatigue strength under cyclic loading.

Figure 1.25. The LWACW weld bead cross sections of (a) 7-mm thick high-strength steel in butt-joint configuration and (b) 8-mm thick structural steel in horizontal joint configuration [46,47].
1.4.4. Hybrid laser/arc welding

To overcome the limitations raised by both autogenous laser welding and laser welding assisted with cold wire, the addition of a secondary heat source such as adding an arc to the laser welding enabled joining operations with a faster welding speed, higher productivity, lower production cost, and reduced product time-to-market. Hybrid laser/arc welding (HLAW) is a possible alternative, because it combines the two processes and preserves their merits while eliminating their disadvantages. HLAW produces a wider weld pool than autogenous laser welding and a deeper penetration than arc welding with the same parameters [49]. The capability of the HLAW process to combine with the filler metal addition gives it a remarkable advantage over pure laser beam welding. LAW also enables a broader range of difficult-to-weld steels to be welded in a single pass under a high welding speed. Due to the presence of an arc, lower cooling rate of the weld pool could help to avoid formation of the brittle phase as well as the reduced presence of pores [50]. However, because of the combination of two heat sources, the process is relatively complicated and many processing parameter must be carefully controlled. The optimal processing parameters of HLAW cannot be readily adjusted from the combination of optimal parameters of the individual welding processes. The interaction between the laser beam and the electric arc causes the optimal value of the parameters to change. The main processing parameters that are involved in HLAW are discussed in detail in the following sections.

1.4.4.1. Heat input

Laser power and welding speed are the most important parameters during HLAW that determine the amount of heat that transfers to the wire or base metal. The best way to consider the interaction of these parameters is to define the heat input based on the following equation [51]:

\[ HI_t = \frac{\eta L P + \eta V I}{V} \]  

(1.1)
Where \( P_L \) is the laser power, \( V \) is the arc voltage, \( I \) is the arc current, \( \nu \) is the welding speed, and \( \eta_A \) and \( \eta_P \) are the efficiencies of the GMAW and laser, respectively. It was shown that the heat input adjustment is a critical step to mitigate various types of defects [52-56]. More and more researchers have invested their efforts to evaluate the effect of the heat input on the soundness of the joint produced by HLAW. Atabaki et al. [52] analyzed the effect of welding speed and groove geometry on the amount of heat input as well as weld quality of the HLAW weld of Armor steels. It was found that increasing the welding speed and making a groove with lower bevel angle could reduce the total amount of heat input and successfully produce a weld with acceptable ballistic behavior. The lower heat input range associated with a higher cooling rate could lead to the desired metallurgical evolution throughout the FZ and HAZ. The role of heat input on the pore mitigation has been the subject of many investigations. Another study performed by Atabaki et al. [53] showed that a larger heat input could create an enlarged weld pool. This enlargement could lead to a longer solidification time that would allow more bubbles to escape from the molten pool (Fig. 1.26). It was explained that a larger heat input enhances the fluidity of the molten material, and allows the pores to move upward as the molten material fills the keyhole.

![Figure 1.26 Longitudinal cross-sections of welds in different heat input levels. (a) 250 kJ/mm and (b) 300 kJ/mm [53].](image)

The effect of the welding speed on the weld integrity can be considered as an independent variable within the concept of heat input. Increasing the welding speed can be favorable not only for industry to improve productivity but also to reduce the key-hole induced
pores and alleviate one of the most important issues in the HLAW of thick materials. Liu et al. [54] reported that a high welding speed could make the keyhole more stable and reduce the pores during the HLAW of thick 17-4 PH martensitic stainless steel. Typical macrographs of the weld longitudinal cross-sections under different welding speed regimens are shown in Fig. 1.27.

Figure 1.27. 12–mm 17-4PH weldments produced by HLAW process under various welding speeds [54].

Another important defect that highly corresponds to the HLAW process is humping. Humping refers to the formation of a periodic non-uniform root reinforcement during the welding process. The gravity force was recognized as the main mechanism of humping formation. To prevail over this force much effort has been made and some solutions have been suggested. Molten pool-induced surface tension pressure in the upward direction was recognized as the main force that could counter the gravitational force and avoid humping [55]. It was found that the best remedy to suppress root humping and improve the surface tension pressure of the molten pool is to increase the molten pool viscosity by reducing the heat input. Pan et al. [56] conducted a series of HLAW trials on 11-mm thick high-strength steel in different laser power ranges to find the best welding condition. It was reported that in the case
of higher laser power and faster welding speed, the greater surface tension-induced force could overcome the gravitational force. The greater surface tension–induced force facilitates the flow of the molten pool behind the laser keyhole, avoid a narrow point solidification, and thus prevent the humping (Fig. 1.28).

![Figure 1.28. Surface appearances and cross-sections of HLAW welds of 11-mm thick high-tensile-strength steel in different laser powers under fast welding speed (25 mm/s)](image)

**1.4.4.2. laser-arc separation spacing**

Distance between the tip of the wire and the laser beam (D_{LA}) is one of the most effective variables that is unique to the HLAW process. The distance variable should be controlled in such a way that a synergetic effect is produced. D_{LA} has a profound effect on the arc characteristic, droplet transfer mode and weld bead geometry [57]. Interaction between the arc-induced plasma and the laser-induced plasma is the key mechanism that renders the synergetic effect. It was reported that in a close proximity between arc electrode and laser beam (3–4 mm), laser-induced plasma reduces arc resistance and improves the arc stabilization. A small part of the laser energy is absorbed by the arc plasma, further ionizing the arc plasma and reducing its electrical resistance. Besides, considerable vaporization of the material occurs at the location where the laser hits on the surface of the work-piece, and metal vapor is then transported into the arc-induced plasma. Relative to shielding gas atoms, metal atoms have a
much lower ionization potential. Thus, the effective ionization potential of the arc-induced plasma is reduced and a more conductive stable plasma channel for the arc column is obtained [58]. Liu et al. [59] captured the arc-induced plasma shape in different pure arc welding and HLAW cases by means of a high-speed color camera (Fig. 1.29). As can be seen, the arc-induced plasma in the case of HLAW is more compressed and its volume is markedly smaller compared to that of pure-arc welding. The smaller the arc plasma, the lower the electrical resistance arc column and the more stable the droplet transfer.

![Figure 1.29. Plasma shapes of various welding methods, (a) gas metal arc welding and (b) HLAW [59].](image)

Reutzel et al. [60] investigated the influence of $D_{LA}$ on the weld penetration of 7-mm A36 structural steel joints obtained by HLAW. According to Fig. 1.30, once $D_{LA}$ is within the range of 2-4 mm then full penetration is obtainable. In that distance range, the laser and the arc energy directly acted on the upper part of the weld pool, and heat input accumulated more efficiently in this region, which could be testified by the smaller arc-induced plasma as shown in Fig. Another study performed by Compana et al. [61] revealed that when the distance between the heat sources is equal to 0, the repeatability of the welds is greatly compromised. It was reported that the dispersion of results could be attributed to the high turbulence of the molten pool that resulted from a destructive interference of the laser and arc-induced plasmas.
Disturbance of the keyhole due to impingement of filler metal droplets onto the keyhole wall was also identified. That disturbance was shown to intensify the instability of the whole process.

![Cross-sections of welds showing DLA changes the penetration depth](image)

**Figure 1.30** Cross-sections of welds show that how DLA changes the penetration depth [60].

### 1.4.4.3. Arc mode

Owing to the dynamic nature of the arc, the current and voltage changes continuously during the arc welding. The amount of wire feed rate determines the required voltage and current to attain a stable arc. However, to produce consistent weld bead regardless of the wire feed rate, the mode of metal transfer should be controlled in such a way that frequency of droplet transfer toward the molten pool should be as high as possible with the least amount of fluctuations. Generally, there are three different modes of metal transfer recognized during arc welding: projected/spray transfer, gravitational/globular transfer and short circuit transfer. Arc power, electrode stick-out, electrode diameter, and shielding gas composition are the main variables that control the mode of metal transfer [57]. There have been several researches to evaluate the effect of arc mode on the final quality of the weld [62-65]. Cai et al. [62] analyzed the influence of laser on droplet behavior in three of the mentioned modes and found that in the case of large heat input coupled with high welding speed the droplet transfer under the spray mode is much preferred for HLAW to ensure deep penetration. In addition, lower turbulence of the weld pool was observed during the spray mode metal transfer due to a
continuous projection of small filler droplets into the weld pool. Increased magnitudes of the electromagnetic force and plasma drag force were considered as the main reasons for faster droplet transfer frequency in the spray mode. Welding current wave was introduced as the criteria to explain the efficiency of droplet transfer. The minimum current fluctuations were observed for the spray mode droplet transfer (Fig. 1.31)

![Figure 1.31 Welding current waves in (a) short-circuiting, (b) globular and (c) spray mode of arc in the HLAW [62].](image)

Recently, an advanced arc welding mode has been developed that in terms of wire feeding is even more controllable than traditional droplet transfer modes. Cold metal transfer (CMT) or mechanically-assisted droplet deposition are the terms that have been used for this technology. The droplet transfer mode in this method is with a controlled short circuit that periodically pulls back and forth on the wire instead of using a continuous wire feed. In the CMT process, once the electrode wire tip touches the molten pool, the servomotor of the welding torch is reversed by digital process control. This function leads to retraction of wire, thereby assisting the liquid fracture and promoting droplet transfer. During droplet transfer, the current reduces to near-zero and spatter generation can be prevented. At the final stage, as soon as the droplet transfer is completed the arc is reignited, and the wire is fed forward [63]. It was found that the efficiency of the wire deposition rate can be improved significantly by using the CMT arc mode. A number of research groups have made great contributions to weld quality
by integration of this technology with the HLAW process [64–65]. Frostevarg et al. [64] investigated the effect of arc mode under three cases of standard, pulsed and CMT on the weld quality of 7-mm thick steel joints produced by HLAW. According to Fig. 1.32 it is observed that the CMT-mode shows higher bead stability and reduced power is supplied.

![Figure 1.32](image)

**Figure 1.32** (a) Average undercut height and (b) average arc power output for different arc modes under different welding speeds [64].

In another study conducted by Li et al [65], three kinds of hybrid welding processes, including laser-pulsed spray arc, laser CMT and laser-standard short-circuiting arc hybrid welding were analyzed in terms of spattering and width of the welding region. It was observed that the CMT transfer mode presented the most stable plasma and the least amount of spatters, while the standard short-circuiting mode had the most unsteady and jumpy plasma followed by greatest number of spatters (Fig. 1.33(a)). The narrowest width of FZ and HAZ were obtained by the CMT mode. These results were a concrete confirmation of the previously-cited research in which the CMT mode consumed the lowest amount of arc power (Fig. 1.33(b)).
1.4.4.4 Laser beam focal position

Location of laser beam focal point is the key parameter to determine the penetration depth particularly for joining materials thicker than 10-mm. The focal position specifies the spot size, that in turn determines the power density on the surface of work-piece as well as the keyhole wall [57]. It was reported that the maximum weld penetration is generally attained in the case of negative focal position where the focal point of the laser beam lies inside the material with respect to the surface (Fig. 1.34) [66]. It was explained that for the case of negative defocus, an inward-beam divergence onto the keyhole wall significantly increases the number of beam reflections. This inward-beam divergence raises compared with the case of zero or positive defocus with an outward-beam divergence. However, the amount spatters at the back side of the weld were increased remarkably for the case of the negative defocus. Presence of the focal position inside the molten pool makes propagation of laser energy deeper. Therefore, the melt flows rearward continuously with an ejection that appears as spatters.

Figure 1.33 (a) Bead appearance and plasma visualization (b) width of the FZ and HAZ under three different arc modes [65].
Pan et al. [67] evaluated the effect of laser beam focal position on the weld penetration of 12-mm thick high strength steels produced by HLAW. It was found that when the focal point was set below the surface, the height of the weld face as well as the degree of unfilled area at the bottom side increased. On the other hand, when the focal point was set above the direction of the surface, the unfilled area was formed along the face side of the weld. A relatively good weld bead was produced at the defocused distance of -1 mm (Fig. 1.35).

Figure 1.34 Surface appearances and cross-sections of the HLAW joints at different defocused distances with respect to the surface [66].

Figure 1.35 Cross-sections of HLAW beads at different defocused distances [67].
1.4.4.5 Shielding gas composition

Like any fusion welding process, usage of shielding gas to protect the molten the pool from ambient atmosphere is required for the HLAW process. As mentioned before, another primary function of the protective gas is to blow away the laser-induced plasma and improve laser energy absorption by keyhole. The latter positive effect of shielding gas is more noticeable under a lower welding speed regimen in which a larger and stronger laser-induced plasma plume is formed above the keyhole [35]. Inert gases such as helium and argon are the most commonly used constituents of shielding gases that are applied in HLAW. Generally, CO$_2$ is also added to shielding gas in order to make the arc column more stable [68]. It was proved that the shielding gas with lower ionization energy, larger thermal conductivity, heavier molecular weight, and lower inverse Bremsstrahlung absorption coefficients results in less diffusive plasma plume. Therefore, a weaker laser-absorbing plasma plume is produced [69].

As can be seen in Table 1.4, among the shielding gases, argon is better option to suppress the plasma plume. Regarding thick materials, it was proved that helium scales up weld penetration. However, the lighter molecular weight of helium relative to argon necessitates higher flow rate of helium to give better protection to the molten pool. Due to cost consideration, helium has not been widely accepted across many industries. For porosity-vulnerable steels like martensitic stainless steel and super duplex stainless steels, nitrogen was recommended to be added to the shielding gas. Nitrogen with its high solubility into molten pool of such alloys can avoid formation of pores [71].

<table>
<thead>
<tr>
<th>Gas Type</th>
<th>Molecular weight (g/mol)</th>
<th>Thermal conductivity (W/m.K)</th>
<th>Ionization energy (eV)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Helium</td>
<td>4</td>
<td>0.15363</td>
<td>24.6</td>
</tr>
<tr>
<td>Ar</td>
<td>40</td>
<td>0.01732</td>
<td>15.8</td>
</tr>
<tr>
<td>CO$_2$</td>
<td>44</td>
<td>0.01615</td>
<td>13.8</td>
</tr>
<tr>
<td>N$_2$</td>
<td>28</td>
<td>0.02520</td>
<td>15.6</td>
</tr>
</tbody>
</table>
Liu et al. [54] studied the effect of various shielding gases on pore formation of 12-mm 17-4 PH stainless steels. Based on the Fig. 1.36, it is observed that nitrogen is the most effective gas in preventing the formation of porosity under various welding speeds. Suppression of porosity was associated with the smaller size of nitrogen-induced plasma that scales down reflection and absorption as the laser passes through.

![Image of longitudinal cross-sections of thick 17-4 PH stainless steel joints](image)

**Figure 1.36** Longitudinal cross-sections of thick 17-4 PH stainless steel joints produced by the HLAW under various shielding gases and welding speeds [54].

Shielding gas can be derived in various configurations. One of the most recently developed shielding gas arrangements is a root shielding gas module that was shown highly effective to obtain a more uniform weld bead at the root side and avoid humping as well. It was explained that the shielding gas support at the bottom surface could change the surface tension of the bottom molten pool in a way that produce a slightly wider and more even weld bead [55].

**1.4.4.6. Relative positioning of heat-sources**

The order in which the laser and arc are arranged is an effective variable on the energy distribution during the HLAW process. Two possible cases can be distinguished. For the arc
leading process, the torch is located at a drag angle (in front of the laser) and during the laser leading process the arc torch is located at a push angle (in trailing side of the laser). It was revealed that the main benefit of the arc-leading configuration is to increase the cooling rate developed by laser, resulting in refinement of microstructure and superior mechanical properties [72]. In contrast, the pressure of arc in the laser-leading configuration could accelerate the out flow of bubbles and reduce the entrapped gas-induced spherical pores that tend to form close to the surface of the weld bead [73]. The typical cross-sections of the weld beads that are produced in various heat source arrangements are shown in Fig. 1.37. As can be observed, in the case of laser leading, undercut is shallower, and the width of the bead is wider. The arc pressure allows the wider molten pool face for gaseous bubbles to be removed. However, to obtain the finer microstructure, the arc leading arrangement can be more effective due to the higher cooling rate. It was reported that the arc leading arrangement during high speed HLAW of high strength steels resulted in improved tensile strength and fracture toughness of weld joints over those of the laser leading process [74].

![Typical cross-sections of the HLAW welded joints](image)

**Figure 1.37 Typical cross-sections of the HLAW welded joints in (a) laser leading and (b) arc leading configuration [74]**

### 1.5. Research Objectives

The overall objective of this research is to develop hybrid laser/arc welding to join thick gauged sections of difficult-to-weld steels in different joint configurations, using high power
disk laser coupled with arc CLOOS machine. The effects of welding variables on the weld quality are studied in detail. Experimentally-validated thermal and thermomechanical analyses are developed to investigate the temperature histories and distribution of residual stresses by the finite element methods. The proposed PhD subject is perused through the following aims:

1. Develop an HLAW process of thick high strength quenched and tempered steels (HSQTSs) in different joint configurations incorporated with experimentally-validated thermomechanical simulation to predict the residual stress and distortion.

2. Study the effect of welding speed and heat-source arrangement on porosity mitigation of AISI 304 stainless steel joints produced by the HLAW process and develop thermal analysis of the welding region.

3. Microstructural evolution of A304 stainless steel joints generated by the HLAW process coupled with tailoring corrosion resistance of the welds produced by various consumables

4. Implement the response surface method (RSM) in order to generate the processing parameter windows for main variables of the HLAW process of A304 stainless steel joints in a horizontal joint configuration.
REFERENCES


CHAPTER 2

EXPERIMENTAL AND THERMOMECHANICAL SIMULATION OF HYBRID LASER/ARC WELDING OF THICK HIGH-STRENGTH STEELS IN DIFFERENT JOINT CONFIGURATIONS

2.1. Introduction

Inside the global competitive world, ship and naval constructors have been trying to adopt lighter steel plates of medium thickness (5-10 mm). These steel plates would decrease topside weight and improve fuel efficiency while still retaining structural performance [1]. Among the various types of steels, high-strength quenched and tempered steels (HSQTS) have come to the attention of the industry. These steels not only reduce the total weight but also keep the structural integrity of the final product. This achievement is due to their high ultimate tensile strength coupled with outstanding fracture toughness and their reduced weight-to-strength ratio [2]. Welding of geometrically different parts is considered an integral part of the current trend in design and manufacturing of ship vessels.

The most commonly-used types of welding processes in the shipbuilding industry are multi-pass arc-based welding process i.e., multi-pass gas metal arc welding [3] and submerged arc welding [4]. These welding processes all share poor efficiency, because of low penetrability of the arc. An intrinsically-large heat input widens the heat affected zone (HAZ), thereby expanding the softening area and degrading the mechanical properties of the final joint [5]. In
addition, due to the excessively large heat input, various regions of the weld are subjected to highly-different heating and cooling cycles. Therefore, the welding process generates unfavorably-large tensile residual stresses and buckling distortion within and surrounding the weld. These issues are of primary concern during the arc-based welding process. In particular, the buckling distortion, referred to as out-of-plane warping, reduces dimensional tolerance and gives rise to manufacturing costs due to the need for an extra distortion-correcting process [6].

Extensive research programs have been developed to scale down weld-induced distortion during the manufacturing process. Many research attempts have invested in the development of welding processes that have high productivity and efficiency. Among the various heat sources, a highly-focused laser beam has given great expectations because it is able to improve penetration depth and narrow down the distortion and softening region of the HSQTS final joint with minimal foot print in the joining area [7]. One of the most important benefits that is offered by the laser-based welding process is the application of much simpler joint configurations such as the square-butt joint configuration instead of the flare-groove joint configurations. The latter has been widely utilized in the arc-based welding processes. Notably all flare-groove joints are followed by a extra machining process. As a result, the square-butt joint geometry can significantly reduce the time and cost of welding. Furthermore, the volume of filler material deposited is greatly reduced during laser welding when compared to the more conventional wider grooves [8]. However, the focal spot size of laser beam is so small that it sets a very strict requirement on machining and assembling of the workpiece [9]. Another issue of autogenous laser welding is the rapid cooling and high solidification rate. This rate accelerates formation of the brittle phase that intensifies weld susceptibility to hot cracking [10]. An alternative technology, the hybrid laser/arc welding (HLAW) process incorporates benefits of the laser and arc welding. Capabilities such as focused heat input and deep penetration that are associated with the laser are combined with improved gap-bridging ability,
chemical modification, and metallurgical stability that originate from the arc character [11].

Thus, HLAW has the ability to improve productivity in excess of what can be attained by either laser or arc welding alone [12]. In spite of all these advantages, use of the HLAW process in actual application is constrained by the highly complex interactions of many processing parameters that are carefully controlled. In addition, the potential appearance of spiking, humping, and cracks as the main defects exacerbate the difficulty of this process to obtain a sound weld [13]. The stand-off distance ($D_{LA}$) between the arc and laser is one of the most important parameters to determine the final properties of the weld, especially the bead appearance and joint micro-hardness [14]. It was revealed that if the $D_{LA}$ is not large enough, the vaporization of the metallic elements would not effectively reduce the electrical resistance of the arc. Therefore, bending of the arc toward the keyhole as well as the effective synergy between heat sources would not occur [15].

Generally accepted, the experimental trials need an excessively long time with additional funding to attain the optimized welding parameters. In addition, an accurate prediction of distortion is extremely critical for the design and manufacturing of stiffened welded structures. Recently, with emergence of powerful computation facilities, researchers are able to apply a finite element method (FEM) to attain a more realistic and better quantitative realization of the residual stress distributions and distortions. In turn, these results facilitate optimization of the processing parameters [16]. A great number of researchers have devised various FEMs to simulate the welding process, such as the sequentially-coupled thermo-mechanical analysis. It was found that to achieve an effective thermo-mechanical simulation two primary requirements should be satisfied. First, the numerical thermal results that are affected by temperature-dependent physical properties should be taken into account. Second, during the cooling cycle, volume change due to the solid-state transformation and temperature-dependent mechanical properties that are different for each microstructural phase have a drastic
effect on distribution of residual stress. [17]. Radaj et al. [18] reported that during the cooling cycle the elastic module and yield stress increased sharply below half of the melting point. These variation of mechanical properties had a profound impact on the distribution of the residual stresses. In order to fulfill the above requirements, SYSWELD commercial code has been employed to simulate different welding geometries. One of the interesting capabilities of SYSWELD is to consider the details of the phase transformation during the thermomechanical analysis and improve the accuracy of the prediction. Lima et al. [19] used SYSWELD to anticipate successfully the residual stress generated in the autogenous laser welding process of different thin Al joints. The effect of various heat inputs was investigated by E.D. Derakhshan et al. [20]. The results of the experiments and simulations (using SYSWELD) between the laser-based welding process and SAW for a material in thickness of 4 mm were compared. It was found that the application of the laser significantly reduced the level of residual tensile stresses inside the weld region with respect to SAW. However, quite a few studies have centered on thermo-metallurgical-mechanical simulation of the HLAW process for relatively thick plates.

In the present chapter, the effects of the processing parameters of HLAW on the weld quality of a 8-mm thick plate of HSQTS steel are studied. Taken into consideration are various joint configurations, including butt- and T-joint types. Characterization techniques, including macro-structural observations, online monitoring of the molten pool by a charged coupled device (CCD) camera, and mechanical evaluation were conducted to better comprehend the effect of the joining variables on the weld soundness. Further, an experimentally-validated finite element simulation of the HLAW process using SYSWELD commercial code was introduced to anticipate the thermal and residual stress field of the weld region for the developed welding conditions. To measure the residual stress, an X-ray diffractometer (XRD) was employed to measure the residual stress and its distribution through the welding region. In
addition, the detailed microstructural analysis was conducted to determine the phases that were formed during the weld.

2.2. Experimental procedure

HSQT steel plates with the dimensions of 127×127×8 mm were used in this investigation. The chemical compositions of the base metal (BM) and the wire (ER70S-6) used for the HLAW welding process, are presented in Table 2.1. Fig. 2.1 presents the high-power setup of the laser equipment for the joining process. A diode pumped Yb:YAG disk laser (TRUMPF, TruDisk 10003 laser), with a continuous wavelength of 1030 nm, beam quality of 8/12 mm.mrad, and maximum power of 10 kW, was applied for the laser welding process. The laser head that was equipped with a reflective focusing optic of 300 mm in a focal length, provided a laser spot diameter of 0.6 mm at the focal point. A CLOOS gas metal arc welding (GMAW) machine (Quinto GLC 403) was used as well to feed the cold wire and provide the electrical arc heat source. The laser head coupled with the arc torch was mounted on a 6-axis, high-precision KUKA KR 60-3 robot to move the laser head along the welding path.

Table 2.1 Chemical compositions of HSQTS steel and consumable wire.

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>Si</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>B</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>HSQTS steel</td>
<td>0.32</td>
<td>1.2</td>
<td>0.045</td>
<td>0.05</td>
<td>0.4</td>
<td>1.8</td>
<td>1.0</td>
<td>0.7</td>
<td>0.005 balance</td>
</tr>
<tr>
<td>ER70S-6</td>
<td>0.06-0.15</td>
<td>1.4-1.85</td>
<td>0.025</td>
<td>0.035</td>
<td>0.9</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>balance</td>
</tr>
</tbody>
</table>
The welding was performed by having the arc at the heading position, and the laser following the arc (Fig. 2.2 (a)). For a T-joint configuration, the plates were tilted into the fixture that was inclined with respect to the horizontal plane at an angle of 35˚-45˚ (Fig. 2.2 (b)). The samples were cleaned with acetone before the welding process. The laser head was tilted 5 degree off its vertical axis to protect it from the back reflection of the light. To protect the molten pool from the ambient atmosphere, shielding gas was customized to protect the weld area in two different ways. (1) The mixture of 92%Ar-8%CO₂ was conducted through the GMAW welding torch and (2) Pure argon was used as the back-shielding gas. CO₂ was added to stabilize the arc column during the welding process [21].
Figure 2.2 Schematic of the positioning of the arc and laser in relation to each other in (a) butt-joint and (b) T-joint configuration and (c) the positioning configuration of thermocouples with respect to the centerline or butt-joint arrangement.

After the weld, the specimens were etched with a solution prepared with distilled water 95 ml, and nitric acid 5 ml (5% Nital solution) that were soaked for 5 s. A high precision KEYANCE optical microscopy (VHX-1000) was employed to observe and gauge the geometrical features of the weld bead cross-section. Furthermore, higher magnification objective lenses were used for microstructural evaluation of the welding region.

The micro-hardness test was performed with a Clark micro-hardness tester, with a 4.9 N load and 15 s of dwell time. The micro-hardness was measured along lines placed approximately 2 mm from the top and 2 mm from the bottom. The micro-hardness was counted from the weld centerline to the micro-hardness of the base metal. Tensile properties of the joints were examined by an Instron mechanical testing machine at a constant strain rate of 1 mm/min. The configuration of the tensile test coupons is shown in Fig. 2.3. The online
monitoring of the welding process was performed by a CCD camera with a frame rate capability of 4000 frames/s.

![Schematic diagram of tensile test coupons (dimension in mm).](image)

Figure 2.3 Schematic diagram of tensile test coupons (dimension in mm).

The macro-structural analysis result was correlated to the heat source shape and validated the thermal analysis results. K-type thermocouple data were entered into a National-Instruments data acquisition system. The inherent accuracy of thermocouple for the temperature range of 0-800°C was 3°C. The data was used to verify the predicted temperature histories. The positioning arrangement of the thermocouples with respect to the centerline is shown in Fig. 2.2.c. The temperature measurement concerning each thermocouple was performed three times and the average temperature history was reported. An X-ray diffractometer (XRD) was utilized at the top surface of the welded plates to measure the residual stress stored in the material. The principle of this procedure is based on the interaction of the X-ray wave and inter-planar spacing in crystalline materials. The wavelength of the X-ray beam (\(\lambda_{C\alpha}=2.291\) Å) is of a similar order of lattice spacing. The relative shift in the lattice spacing represents the expansion and shrinkage, that happens during heating and cooling cycles. The lattice spacing shift is used to specify the elastic residual stress that is stored in the welded joint. Regarding the wavelength of the X-ray beam, the change in the lattice spacing can be determined by using Bragg’s law [22]:

\[
2d_{hkl}\sin \theta_{hkl} = \lambda
\]  

(2.1)
Where $\lambda$ is the wavelength of the incident X-ray beam, $d_{hkl}$ is the lattice plane spacing and $\theta_{hkl}$ is the Bragg’s angle for a given crystallographic plane, denoted by Miller indices $\{hkl\}$. To determine the elastic strain, the following expression can be used:

$$\varepsilon_{hkl} = \frac{d_{hkl} - d_{0,hkl}}{d_{0,hkl}} = \frac{\sin \theta_{0,hkl}}{\sin \theta_{hkl}} - 1 \quad (2.2)$$

Where, $d_0$ is the stress-free lattice spacing and $\theta_0$ is the corresponding diffraction angle. To calculate the final residual stress in principal states, the linear-elastic Hook’s law was acquired:

$$\sigma_i = \frac{E_{hkl}}{(1 + \nu_{hkl})(1 - 2\nu_{hkl})} \left[ (1 - \nu_{hkl})\varepsilon_i^{hkl} + \nu_{hkl}(\varepsilon_j^{hkl} + \varepsilon_k^{hkl}) \right] \quad (2.3)$$

Where $E_{hkl}$ and $\nu_{hkl}$ are the elastic modules and Poisson’s ratio of a specific crystallographic plane, respectively. The measurement was done in uneven intervals. Therefore, a 5-mm distance from the bead toe the adjacent distance between points was adjusted to be 1mm. For a distance larger than 5 mm, the interval was increased to 5 mm. To study the thermally-induced distortion, a Digital Vernier Caliper was utilized to gauge the Z displacement.

2.3. Thermomechanical simulation

2.3.1. Finite Element Model

Temperature field distribution throughout the welding region is considered an important prerequisite to evaluate side effects of the welding process such as residual stress and distortion occurring at the joint. To achieve this prerequisite, a three-dimensional thermo-metallurgical-mechanical finite element model was expanded, using the SYSWELD commercial code. The sequential steps that were followed in FEM analysis are shown in Fig. 2.4.
To increase the accuracy of thermal and mechanical analysis, the model was divided away from the centerline into three regions. The finer mesh size was picked up in the adjacent area of the weld bead, and coarser mesh was selected away from the weld centerline (Fig. 2.5.).
The element types to mesh the 3D model were an 8-noded (Hexahedron) and a 6-noded (Prism) that were associated with the codes 3008 and 3006 in SYSWELD, respectively [23]. The main assumptions used to perform the FE analysis were as follows:

1. The effects of fluid dynamics and free surface were neglected. The model was considered as a pre-constructed geometry of the joint, including top and back reinforcement in a single solid-state.
2. The thermal conductivity at temperatures higher than the melting point was altered artificially to rectify the effect of heat convection inside the molten pool [24].
3. The effect of latent heat of fusion was considered by modifying the specific heat.
4. The role of shielding gas on heat dissipation was neglected.
5. The heat-sink capability of the clamping fixture was ignored.

![Figure 2.5 Finite element mesh of butt-joint specimen (arrows represent the clamping condition).](image)

**2.3.2. Heat source modeling**

The transient energy conservation was utilized as the primary equation for thermal analysis during HLAW process by considering the temperature-dependent physical properties:

\[
\rho(T)C_p(T)\frac{\partial T}{\partial t} = K(T) \left\{ \left( \frac{\partial^2 T}{\partial x^2} \right) + \left( \frac{\partial^2 T}{\partial y^2} \right) + \left( \frac{\partial^2 T}{\partial z^2} \right) \right\} + q_{laser}(x, y, z, t) + q_{arc}(x, y, z, t) \tag{2.4}
\]
Where \( x, y, \) and \( z \) are the Cartesian coordinates, \( \rho \) is the density, \( K \) is the thermal conductivity, and \( C_p \) is the specific heat. All mentioned thermal properties are considered as a function of temperature \( T \). The temperature-dependent material properties of high strength steel and the ER70S-6 wire are shown in Fig. 2.6.

![Graphs showing thermal properties](image)

**Figure 2.6** Variation of thermal properties of high strength steel as well as the consumed wire with respect to the temperature [23].

The heat source modeling should be acquired accurately to obtain a reasonable temperature distribution in the target application. The laser-contributed power distribution was determined as the conical volumetric heat source derived by a following equation [24]:

\[
\dot{q}_{\text{laser}}(x, y, z, t) = \frac{2\eta_{\text{laser}}P_{\text{laser}}}{\pi(1-e^{-3})(r_o^2+r_e r_i+r_i^2)Z} \exp\left(-3 \frac{x^2+(y-vt)^2}{r_c^2}\right)
\]  

(2.5)
\[ r_c = r_i + (r_e - r_i) \frac{z-z_i}{z_e-z_i} \] (2.6)

Where \( \eta_{laser} \) is the laser absorption efficiency, \( P_{laser} \) is the nominal power of the laser beam, \( r_c \) is the flux distribution parameter for the cone as a function of depth \( z \), \( r_e \) and \( r_i \) are the larger and smaller radii of the cone respectively, \( Y \) is the thickness of material, and \( v \) is the welding speed. To express the arc-induced power distribution the Goldak double ellipsoidal configuration of the heat source was employed [25]. The heat flux distribution throughout the proposed shape is numerically calculated as follows:

\[
\dot{q}_{\text{arc}}^f(x, y, z, t) = \frac{6\sqrt{3}f \eta_{arc} P_{arc}}{a_f b c \pi \sqrt{\pi}} e^{-3x^2/b^2} e^{-3(z-Z)^2/c^2} e^{-3(y-vt+D_{LA})^2/a_f^2} \text{ for } y > vt
\] (2.7)

\[
\dot{q}_{\text{arc}}^r(x, y, z, t) = \frac{6\sqrt{3}f \eta_{arc} P_{arc}}{a_r b c \pi \sqrt{\pi}} e^{-3x^2/b^2} e^{-3(z-Z)^2/c^2} e^{-3(y-vt+D_{LA})^2/a_r^2} \text{ for } y < vt
\] (2.8)

Where \( P_{arc} \) is the nominal arc power, \( \eta_{arc} \) is the arc absorption efficiency, and \( D_{LA} \) is the offset distance between laser and arc heat sources. The \( a_f \), \( a_r \), \( b \), and \( c \) are associated with the front length, rear length, depth, and width of the molten pool (Fig. 2.7).

Figure 2.7 The applied volumetric heat sources in thermal analysis: (a) double-ellipsoidal and (b) conical [25].
2.3.3. Thermal boundary conditions

The initial temperature of the FE model was adjusted to the ambient temperature (20°C). The heat lost through free surfaces of the model were considered convection and radiation. The following expressions are used to define the heat loss because of the natural convection \( q_{\text{conv}} \) and radiation \( q_{\text{rad}} \), respectively:

\[
q_{\text{conv}} = h_{\text{conv}}(T - T_0)
\]

\[
q_{\text{rad}} = 5.68 \times 10^{-8} \varepsilon [(T - T_{\text{abs}})^4 - (T_0 - T_{\text{abs}})^4]
\]

Where \( T \) is the local temperature of the surface, \( T_0 \) is the room temperature, \( T_{\text{abs}} \) is the absolute zero temperature, and \( \varepsilon \) is the emissivity. The effect of radiation is almost negligible at the environmental temperature.

2.3.4. Model Mesh sensitivity

The computing precision of the FE analysis depends on the element size and number of nodes. Evidently, the finer the element size, the more accurate the solution, yet, the longer the computational time. In order to maintain the calculation accuracy, the grid independence of the FE model was studied by comparing the peak temperature of the specific node that was monitored experimentally and numerically simulated. Accordingly, a series of trials were conducted in the FE model by using finer and finer elements under the same material properties, element type, and boundary conditions. Fig. 2.8 presents the verification results of the grid independence at two different nodes.
Figure 2.8 Experimentally-recorded and numerically-predicted temperature histories of two nodes that were located at (a) 6 mm and (b) 9 mm from the weld line and (c) variation of maximum temperature with respect to node size.

As can be seen, by increasing the element size above 60000, the maximum temperature becomes almost independent of the element size. Thus, the number of nodes in the developed FE model should be larger than 60000 in order to obtain reliable numerical results.

2.3.5. Mechanical analysis

The numerically-predicted thermal history captured in the previous thermal analysis was applied as a thermal input load for subsequent mechanical analysis. The clamping conditions are shown in Fig. 2.5. The temperature dependent mechanical properties such as Young’s module, yield strength and thermal expansion coefficient that were acquired in
simulation are shown in Fig.2.9. The total strain increment of any node in the model is expressed by the following equation [6]:

\[ d\varepsilon^{Total} = d\varepsilon^{elastic} + d\varepsilon^{plastic} + d\varepsilon^{thermal} + d\varepsilon^{phase} \]  

(2.11)

where \(d\varepsilon^{Total}\), \(d\varepsilon^{elastic}\), \(d\varepsilon^{plastic}\), and \(d\varepsilon^{thermal}\) are the total, elastic, plastic, thermal strain increments, respectively, and \(d\varepsilon^{phase}\) is the phase transformation-induced strain increment that resulted from the volume misfit between the different phases. To numerically predict the distortion that occurred after the welding process, SYSWELD was equipped with geometrical equations that would consider a large distortion theory. Accordingly, the following equations are used to define the non-linear relationship between strains and displacements [6]:

\[ \varepsilon_x = \frac{\partial u}{\partial x} + \frac{1}{2} \left\{ \left( \frac{\partial u}{\partial x} \right)^2 + \left( \frac{\partial v}{\partial y} \right)^2 + \left( \frac{\partial w}{\partial z} \right)^2 \right\} \]  

(2.12)

\[ \varepsilon_y = \frac{\partial v}{\partial y} + \frac{1}{2} \left\{ \left( \frac{\partial u}{\partial x} \right)^2 + \left( \frac{\partial v}{\partial y} \right)^2 + \left( \frac{\partial w}{\partial z} \right)^2 \right\} \]  

(2.13)

\[ \varepsilon_z = \frac{\partial w}{\partial z} + \frac{1}{2} \left\{ \left( \frac{\partial u}{\partial x} \right)^2 + \left( \frac{\partial v}{\partial y} \right)^2 + \left( \frac{\partial w}{\partial z} \right)^2 \right\} \]  

(2.14)

\[ \gamma_{xy} = \frac{\partial u}{\partial y} + \frac{\partial u}{\partial y} + \frac{1}{2} \left\{ \left( \frac{\partial u}{\partial x} \right) \left( \frac{\partial u}{\partial y} \right) + \left( \frac{\partial v}{\partial y} \right) \left( \frac{\partial v}{\partial x} \right) + \left( \frac{\partial w}{\partial z} \right) \left( \frac{\partial w}{\partial y} \right) \right\} \]  

(2.15)

\[ \gamma_{yz} = \frac{\partial v}{\partial z} + \frac{\partial w}{\partial z} + \frac{1}{2} \left\{ \left( \frac{\partial u}{\partial x} \right) \left( \frac{\partial u}{\partial y} \right) + \left( \frac{\partial v}{\partial y} \right) \left( \frac{\partial v}{\partial z} \right) + \left( \frac{\partial w}{\partial z} \right) \left( \frac{\partial w}{\partial z} \right) \right\} \]  

(2.16)

\[ \gamma_{zx} = \frac{\partial u}{\partial z} + \frac{\partial u}{\partial z} + \frac{1}{2} \left\{ \left( \frac{\partial u}{\partial x} \right) \left( \frac{\partial u}{\partial y} \right) + \left( \frac{\partial v}{\partial y} \right) \left( \frac{\partial v}{\partial z} \right) + \left( \frac{\partial w}{\partial z} \right) \left( \frac{\partial w}{\partial z} \right) \right\} \]  

(2.17)

Where \(\varepsilon_x, \varepsilon_y\) and \(\varepsilon_z\) are the Green-Lagrange strain in the \(x, y\) and \(z\) directions; \(\gamma_{xy}, \gamma_{yz}\) and \(\gamma_{zx}\) are the shear strain on the \(x - y, y - z\) and \(z - x\) planes; \(u, v\) and \(w\) are the displacement in the \(x, y\) and \(z\) directions, respectively.
Figure 2.9 Variation of mechanical properties of high strength steel as well as the consumed wire with respect to temperature [23].

2.4. Results and discussion

The common goals of the two configurations in this study were to achieve visually acceptable full-penetrated weld, together with as little defect as possible on advanced HSQTS, performed in a single HLAW pass.

2.4.1. Butt-joint configuration

2.4.1.1 Selection of welding parameters

To gain a sound full-penetrated weld at the butt-joint configuration, single-pass HLAW was performed to join the HSQTS. An overview of the developed welding parameters for this process is summarized in Table 2.2. Wire feed rate, current, and voltage were matched automatically by an algorithm in the GMAW machine. The flow of parameters was set based
on the experimental results achieved, that exhibited full penetration as well as an even bead at the top and bottom side of the weld. In addition to the weld bead uniformity, a minimal amount of spatters and undercut coupled with cross-section quality were specified as the subsequent criteria for weld acceptance.

**Table 2.2** The welding parameters used during the HLAW process in butt-joint configuration

<table>
<thead>
<tr>
<th>Test number</th>
<th>Laser Power (kW)</th>
<th>Welding speed (mm/s)</th>
<th>Wire feed rate (mm/min)</th>
<th>Current (A)</th>
<th>Voltage (V)</th>
<th>Top shielding gas flow rate (l/min)</th>
<th>Root shielding gas flow rate (l/min)</th>
<th>D_{LA} (mm)</th>
<th>Gap (mm)</th>
<th>Heat input (kJ/m)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>5</td>
<td>20</td>
<td>9</td>
<td>228</td>
<td>35</td>
<td>32</td>
<td>-</td>
<td>4.5</td>
<td>&lt;0.05</td>
<td>544</td>
</tr>
<tr>
<td>2</td>
<td>5.5</td>
<td>20</td>
<td>9</td>
<td>228</td>
<td>35</td>
<td>32</td>
<td>-</td>
<td>4.5</td>
<td>&lt;0.05</td>
<td>567</td>
</tr>
<tr>
<td>3</td>
<td>6</td>
<td>20</td>
<td>9</td>
<td>228</td>
<td>35</td>
<td>32</td>
<td>-</td>
<td>4.5</td>
<td>&lt;0.05</td>
<td>589</td>
</tr>
<tr>
<td>4</td>
<td>8</td>
<td>35</td>
<td>10</td>
<td>228</td>
<td>35</td>
<td>32</td>
<td>18</td>
<td>4.5</td>
<td>&lt;0.05</td>
<td>388</td>
</tr>
<tr>
<td>5</td>
<td>8</td>
<td>35</td>
<td>10</td>
<td>228</td>
<td>35</td>
<td>32</td>
<td>18</td>
<td>3</td>
<td>&lt;0.05</td>
<td>388</td>
</tr>
<tr>
<td>6</td>
<td>8</td>
<td>35</td>
<td>10</td>
<td>228</td>
<td>30</td>
<td>32</td>
<td>18</td>
<td>3</td>
<td>&lt;0.05</td>
<td>362</td>
</tr>
</tbody>
</table>

Three main parameters including the laser power, welding speed, and D_{LA} were changed to obtain a sound weld. Fig. 2.10 presents the longitudinal views of the welds at the weld face and weld root under different conditions. In Fig. 3, “F” is the front image, and “B” is the back-side image. As can be seen, under the low welding speed (20 mm/s) and low power, a full penetration weld is not achieved (Fig. 2.10 (a and b)). Increasing the laser power improved the depth of penetration; whereas, full-penetrated weld joints were produced with regular and periodic formation of drops that was known as the humping effect (Fig. 2.10 c). The humping phenomenon was one of the most commonly occurring defects in laser-based welding processes. A competition between different forces was recognized as the main mechanism that controlled the molten pool stability at the root side, which discussed in the next section. Under welding speed of 20 mm/s, the molten pool did not seem stable enough to avoid the local accumulation of molten material incorporated with the formation of solidified drop. In addition, the presence of an unfilled area for the higher laser power case is observed (Fig. 2.11. b). Based on the cross-sectional view, porosity forms for all low welding speed conditions
Formation of pores could be attributed to the instability and collapse of the keyhole. Another issue regarding the welding under a low welding speed condition was an enlargement of width of the bead by increasing the laser power.

Increasing the welding speed and laser power incorporated with the application of root shielding gas, at the same $D_{LA}$, improved the uniformity of the weld bead on the back side to a certain extent (Fig. 2.10. d). Recent studies confirm that the utilization of back-side shielding gas would produce a long molten pool, with a slightly wider bottom part, by enhancing the velocity of the melt flow [26]. However, the non-uniformity of the weld bead at root side is still observed. A reduction in the $D_{LA}$ was recognized as an efficient way of improving the synergy between the arc and laser beam [27]. At a small $D_{LA}$, it was observed that the uniformity of the weld bead improved significantly, specifically at the back side (Fig. 2.10. (d and e)). However, undercut remained the main defect on the weld face (Fig. 2.11. d). Previous studies have shown that arc voltage has a large effect on arc shape, and a reduction of arc voltage could be helpful to mitigate the negative effect of arc on the formation of undercuts [28, 29]. Therefore, an optimum arc voltage was set to get the least amount of undercut. Based on the cross-sectional macrographs, under a high welding speed condition, there is no sign of pore. This result implies that the keyhole becomes more stable (Fig. 2.11. (c) - (e)). The weld cross-section of the optimized HLAW was symmetrical, fully penetrated, and had a smooth transition to the base metal with the least amount of undercuts or excessive reinforcement (Fig. 2.11. (e)).
Figure 2.10 Typical weld faces and weld root beads for the HLAW process at different welding conditions for butt-joint configuration (the figure numbers are associated with the numbers in Table 2.2).
2.4.1.2. Mechanism of root humping defect formation

To attain a better understanding of the formation of root humping, it is necessary to identify the type of forces on the molten pool during the welding process. Fig. 2.12 depicts a cross-sectional tube with height h and width d, that represents the simplified weld shape. The melt experiences a gravitational pressure based on the following equation:

\[ F_p = \rho g V_{FZ} \]  

(2.18)

Where, \( \rho \) and \( V_{FZ} \) are the molten pool density and volume of molten pool, respectively. The gravitational pressure is a downward force that has a crucial role in formation of the humping defect [30]. The downward force induced by the weight of molten pool is counteracted by the surface tension \( \sigma \), that acts upward at the edges of the open tube. The force that is generated by the surface tension acts along the solid-liquid interface to make the exit surface smooth. Furthermore, the surface tension-induced force imposes a reaction upward, preventing the melt from falling out of the weld zone [31]. The surface tension-induced force is angle-dependent and derived as:

\[ F_\sigma = \pi d \sigma \sin \alpha \]  

(2.19)
Where, $\alpha$ is the angle of solid-liquid interface with respect to the horizontal axis that results in the maximum value of upward force when $\alpha=90^\circ$. Therefore, the balance between the surface tension-induced upward force and weld pool weight-induced downward force is recognized as the main contributor to the formation of the root humping. The adequately high surface tension of the molten pool at the bottom side of the weld can restrain the liquid metal and discourage root defect formation. In contrast, weight of the molten pool acts to promote the formation of root defects. Thus, competition between the two forces determines whether defects will form.

The surface tension of the molten pool depends on chemical composition, temperature, and the surrounding atmosphere (e.g. oxygen produces FeO surfactants). The presence of oxygen in the molten iron has a significant effect on the surface tension. It was reported that at high oxygen contents between 0.06 and 0.1 wt-%, the surface tension of liquid iron is reduced by 50% or more of its oxygen-free value, 1.91 N/m [32]. The molten pool with low surface tension value can easily drop down at the root side of the weld and form root hump. The main source of oxygen is the surface oxide scale on top and surface of the plates that should be reduced by fine milling process before the main welding step [31]. Regarding the present study, before the main welding process, the top surface of as-received plates were milled to reduce the effect of oxide scale. The viscosity of the molten pool is another important feature that determines the surface tension. Evidently, hotter molten pool produced by larger heat input is of a lower viscosity, resulting in lower surface tension.

However, the surface tension is not sole source of the root defects. The volume of the weld pool is another key factor that specifies the molten pool-induced weight force. To simplify the effect of weight force, the cross-section area of the weld pool shape was assumed as an acceptable approximation of the molten pool volume since the length of the weld is constant for all welding conditions. Fig 2.13 presents the measured area of the FZ and magnitude of the heat input under different welding conditions. As can be observed, under low welding speed
regimen, the FZ area was highly expanded throughout the base metal. By increasing the laser power under the constant welding speed, the FZ area reached the maximum value. The volume of the molten pool highly depends on the heat input. According to Fig 2.13 for all welding conditions, there is a direct relationship between the FZ area and the amount of heat input received by the material. Under a low welding speed regimen, the heat input was sufficiently high. Thus, the higher gravitational force produced by larger volume of molten pool overcame the tension surface and caused the root humping. On the other hand, under a high welding speed regimen, the FZ area was reduced significantly. In this case, the surface tension-induced force was made dominant over the gravitational force and no sign of root humping was evident under faster welding conditions. As mentioned before, lower heat input positively improves the molten pool viscosity as well as the surface tension. Thus, the obtained results agreed well with the hypothesis that a faster welding speed has a double-positive effect to suppress the root humping formation. In fact, increasing the welding speed not only reduces the metal volume as well as the weight force, but also improves the surface tension-induced force that counters the weight force.

Figure 2.12 Schematic description of the exerted forces on a free lower exit of a liquid filled tube
Figure 2.13 Variation of the cross sectional area of the FZ and bead width under different welding conditions.

2.4.1.3. Weldment characterization

To achieve a better understanding about the synergy effect between the arc and laser, interaction of arc- and laser-induced plasma plum was studied using CCD camera under various \( D_{LA} \) s (Fig. 2.14.). As can be seen, the arc-induced plasma at the large \( D_{LA} \) was disturbed significantly. Moreover, the larger \( D_{LA} \) didn’t effectively cause a bending of the arc on top of the laser keyhole. This lack of arc-induced plasma compression does not allow the laser to penetrate properly through the coupon. The interruption of the laser penetration caused the collapse of the keyhole, resulting in the formation of pores. On the other hand, at optimal \( D_{LA} \) the arc-induced plasma was almost uniformly compressed toward the keyhole. As a result, the laser penetration throughout the keyhole occurred without interruption, thereby making keyhole more stable.
Figure 2.14 Influence of the DLA on the behavior of the molten pool captured by CCD camera during the HLA of HSQTS in the butt-joint configuration

The micro-hardness distribution through the cross-section of the joint that was produced under optimum conditions is shown in Fig. 2.15. From the weld centerline toward the base metal, three distinct regions could be distinguished. The first region corresponds to the hardening effect that occurred within the FZ. The second region next to the FZ experienced softening behavior that was related to the HAZ. The sharp reduction of hardness from the FZ through the HAZ could be attributed to the cooling rates that were experienced by these regions. Eventually, the third region retained the hardness of the base metal that was not influenced by heating and cooling cycles during the welding.

Figure 2.15 Micro-hardness distribution at the cross-section of the butt joint for top and bottom part of the weld.
The results of tensile tests of the base metal and welded joint under optimum conditions are given in Fig. 2.16. According to the measured tensile properties (Table. 2.3), the joints exhibited only 10% lower ultimate tensile strength compared to that of the BM. Moreover, it was found that elongation of the weld was reduced slightly compared to the BM. Another noticeable point is that the weld specimen failed in the HAZ. Thus, the softening behavior and premature failure that occurred through the HAZ leads to the conclusion that the HAZ was the weakest part of the weld.

Figure 2.16 (a) Engineering stress-strain curves of the HLAW weld at optimum conditions and (b) fracture location of tensile coupon.
It was reported that the strain localization is the main mechanism of premature tensile failure of the joints that happened within the HAZ. Relative to the homogenous microstructure of the non-welded plate, the joint material presented a variation in hardness. The strain localization is highly dependent on the width of the non-homogenous zones that constituted the FZ and HAZ [33]. Based on the micro-hardness results, the average width of the welding region that includes the FZ and HAZ were measured to be 4.5 mm. Thus, the local strain produced inside the FZ and HAZ of the HLAW joint during tensile loading was high enough to cause joint failure in the welding region.

**Table 2.3** Tensile properties of the as-received plate and welded coupons of HSQTS steel

<table>
<thead>
<tr>
<th>Type of joint</th>
<th>Average tensile properties</th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>YS (MPa)</td>
<td>UTS (MPa)</td>
</tr>
<tr>
<td>As-received plate</td>
<td>1388</td>
<td>1654</td>
</tr>
<tr>
<td>Weld sample</td>
<td>1196</td>
<td>1405</td>
</tr>
</tbody>
</table>

**2.4.2. T-joint configuration**

Table 2.4 shows the welding parameters to join thick plates in the T-joint configuration. The aim for this joint configuration was to obtain a single pass HLAW without any interruptions, a small back reinforcement, and the least amounts of defects. For this type of joint, the number of processing parameters affecting the weld bead quality was more than that for the butt-joint configuration. To reduce the number of experiments, the main parameters considered were the welding speed, fixture angle, $D_{LA}$ in the $y$-direction, and GMAW arc voltage.
Table 2.4 The welding parameters used during the HLAW process in butt-joint configuration

<table>
<thead>
<tr>
<th>Test number</th>
<th>Welding speed (mm/s)</th>
<th>Laser Power (kW)</th>
<th>Fixture angle to horizontal line (˚)</th>
<th>D_L in Y direction (mm)</th>
<th>Wire feed rate (mm/min)</th>
<th>Voltage (V)</th>
<th>Current (A)</th>
<th>D_L in X direction (mm)</th>
<th>Gap (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>35</td>
<td>9</td>
<td>42</td>
<td>4.5</td>
<td>18</td>
<td>42</td>
<td>228</td>
<td>3</td>
<td>&lt;0.05</td>
</tr>
<tr>
<td>2</td>
<td>25</td>
<td>9</td>
<td>42</td>
<td>4.5</td>
<td>12</td>
<td>39</td>
<td>228</td>
<td>3</td>
<td>&lt;0.05</td>
</tr>
<tr>
<td>3</td>
<td>10</td>
<td>6.5</td>
<td>42</td>
<td>4.5</td>
<td>10</td>
<td>35</td>
<td>228</td>
<td>3</td>
<td>&lt;0.05</td>
</tr>
<tr>
<td>4</td>
<td>10</td>
<td>6.5</td>
<td>45</td>
<td>4.5</td>
<td>10</td>
<td>35</td>
<td>228</td>
<td>3</td>
<td>&lt;0.05</td>
</tr>
<tr>
<td>5</td>
<td>10</td>
<td>6.5</td>
<td>38</td>
<td>4.5</td>
<td>10</td>
<td>35</td>
<td>228</td>
<td>3</td>
<td>&lt;0.05</td>
</tr>
<tr>
<td>6</td>
<td>10</td>
<td>6.5</td>
<td>38</td>
<td>3</td>
<td>10</td>
<td>35</td>
<td>228</td>
<td>3</td>
<td>&lt;0.05</td>
</tr>
<tr>
<td>7</td>
<td>10</td>
<td>6.5</td>
<td>38</td>
<td>3</td>
<td>10</td>
<td>32</td>
<td>228</td>
<td>3</td>
<td>&lt;0.05</td>
</tr>
</tbody>
</table>

Fig. 2.17 shows the appearances of the weld bead on the weld face and weld root obtained by the HLAW process. As can be observed, at higher welding speed, the back-side reinforcement was not achieved. This result was in contrast to the sound butt-joint welds obtained at the same welding speed (Fig. 2.17 (a and b)). Based on transverse cross-sectional views of the joints (Fig. 2.18 (a and b)), various types of defects were formed under high-welding speed conditions. In contrast to the butt-joint welds, different types of porosity including linear and spherical shapes were observed. Reduction of the welding speed was specified as the first step to achieve the full penetration. However, the back-side weld bead was not uniform (Fig. 2.17 (c)).

Changing the fixture angle was the next step to improve the consistency of the weld bead at the back side while other parameters were considered the same as before. It was observed that increasing the fixture angle with respect to the horizontal line was not effective to get full penetration (Fig. 2.17 d). Based on the cross-section of the joint that was produced under the large fixture angle, the laser-induced keyhole could not reach the corner of the T-joint at the back side (Fig. 2.18 d). It was seen that decreasing the fixture angle to 38˚ could put the laser beam on the back corner of the T-joint. Thus, a series of experiments were
performed under the low fixture angle. Full penetration formed along the bead; whereas, the root inconsistency was perceived as the major constraint (Fig. 2.17 e). Observation of the cross sections under the large $D_{LA}$ revealed that the synergy effect of the arc and laser was not efficient enough to generate a sound weld (Fig. 2.18 (b - e)). In other words, for all conditions under the large $D_{LA}$, regardless of fixture position coupled with its angle magnitude, an unfilled area was observed as the main effect. The unfilled area was located between the molten pool induced by arc and the keyhole made by laser. This defect could be attributed to the fact that the synergy between the arc and laser did not occur efficiently. Thus, reducing the stand-of distance in y-direction was determined as a further step to obtain a defect-free weld bead.

Eventually, a sound fully-penetrated weld with a consistent bead at the back side was produced under a 3-mm stand-off distance (Fig. 2.17 f). In addition, based on cross-sectional macrograph (Fig. 2.18 f), the unfilled area was removed, implying an effective laser-arc synergy. Similar to butt- joint configuration, undercut was formed as another main defect at the face side of the weld (Fig. 2.18 f). The arc voltage was set to produce a weld with the least amount of undercut. This problem was adjusted by controlling the arc voltage. The optimal set of parameters for the T-joint was a welding speed of 10 mm/s, fixture angle of 38°, and a 3-mm $D_{LA}$.  

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Figure 2.17 Typical weld faces and weld root beads for the HLAW process at different welding conditions for T-joint configuration
Figure 2.18 Transverse cross-sections of T-joint welds in different welding conditions.

Fig 2.19 shows the micro-hardness distribution across the weld bead of a T-joint configuration at different directions. As can be seen, the hardness distribution differs at different locations. The presence of the hardness reduction area in the C-C and B-B directions, compared to the D-D direction, implies the presence of HAZ in the C-C direction. Similar to butt-joint configuration, softening behavior is evident through the HAZ in T-joint configuration.

Online monitoring of the T-joint configuration, to track the stability of the keyhole during the welding process, was done by the CCD camera (Fig. 2.20). The keyhole was stable at the D_{LA} of 3-mm between the arc and laser in the y-direction; whereas, if the D_{LA} was selected larger, the keyhole would be unstable and would collapse. This instability led in turn to an unstable molten pool that caused the formation of pores and unfilled area at the bond region.
Figure 2.19 Micro-hardness distribution at the cross-section of the T-joint in different directions

Figure 2.20 Influence of DLA on the behavior of the molten pool captured by CCD camera during the HLAW of HSQTS in the T-joint configuration
2.4.3. Thermal analysis

In the current study, a sequentially-coupled thermomechanical analysis method was performed to simulate the HLAW process. Thermal analysis was first conducted to generate the thermal history of the weld process, because the thermal analysis outputs were acquired as the thermal load input for the subsequent mechanical analysis. The computed 3D temperature distributions for the developed welding condition is shown in Fig 2.21 in different views. According to the melting point (1450°C) and the boiling temperature of iron (2800°C), the approximate geometry of the keyhole and molten pool can be determined.

![3D temperature contours predicted by FE analysis under the developed welding condition of 8kW laser power and 35 mm/s welding speed.](image)

**Figure 2.21 3D temperature contours predicted by FE analysis under the developed welding condition of 8kW laser power and 35 mm/s welding speed.**

In order to verify the accuracy of the temperature distribution, the weld-bead shape predicted by FE analysis was compared with the actual cross-sectional macrographs of the weld bead. Fig. 2.22.b compares the numerically predicted temperature contours with the geometrical features of frozen molten pool under optimal conditions. As can be observed, tendency of the isotherm profiles kept a good match with the cross section of the bead. Furthermore, to verify the predicted width of HAZ by thermal analysis, the corresponding
micro-hardness distribution was prepared as shown in Fig. 2.22.c. As can be seen, a hardness reduction close to the FZ led to formation of the softened area. The area that showed softening behavior was considered to approximate the HAZ. The width of HAZ measured from the hardness profile was in good agreement with the size of the predicted HAZ. It should be noticed that the width of the numerically-predicted HAZ was determined based on the temperature range in which solid-state transformations took place. For HSQT steel, the mentioned temperature range was between 700 to 1450˚C.

Figure 2.22 (a) Longitudinal view of the bead, (b) Comparison of the experimentally measured and numerically predicted weld bead cross sections for a HSQTS joint produced by HLAW process, (c) the micro-hardness profile at the cross section of the weld.

The numerically-predicted temperature histories and experimentally-recorded temperature histories by thermocouples at related positions of TC1 to TC4 (Fig. 2.22.c) are
given in Fig. 2.23. As clearly shown in Fig. 2.23, temperature distribution during heating and cooling stages was predicted with relatively good precision. Deviations that were observed at the peak temperatures for all locations could be attributed to the modeling assumptions and instrumental errors concerning the thermocouples. In addition, it was reported that the joint interface of thermocouple and steel plates could negatively affect the measurement and decrease the heat conductivity at the interface [34].

![Graph](image1)

**Figure 2.23** Experimentally-measured and numerically-predicted (a) thermal history curves at TC1 to TC4 thermocouples (b) corresponding peak temperatures
2.4.4. Mechanical analysis

2.4.4.1 Study of residual stress

The thermally-induced residual stress distribution in transverse and longitudinal directions was studied numerically and experimentally for the optimum condition of butt-joint weldment, and the results are depicted in Fig. 2.24 (a). As can be seen, FE simulated and experimental results show a relatively similar trend from qualitative point of view. The inconsistency between numerically-predicted and experimentally-measured data could be referred to as a gauging error of the X-ray diffraction machine as well as the non-homogeneity of the weld-bead surface. As can be observed, the residual stress profile is recognizable by the positive residual stress (tensile mode) at area next to the FZ. The tensile residual stress is followed by a further reduction in the residual stress to zero or even a negative (comprehensive) value as the distance away from the weld bead increases. Residual stress was produced primarily due to differential plastic flow and thermally-induced strains; although, it was revealed that the role of solid-state phase transformations on the residual stress field should be considered as well [35]. As evidently seen in Fig. 2.24 (a), the maximum values of longitudinal and transverse residual stress are not exactly at the center of the welding region. Rossini et al. [35] explained that such behavior was induced by a combination of contraction and phase transformation. It was shown that the residual stress would be pure tensile at the welding region, if just thermally-induced strains are involved and no solid-state phase transformation happens. However, additional phase change makes the residual stress profile complicated, thereby resulting in a residual stress peak away from the weld centerline. For HSQT steels, the welding process caused the formation of martensite phase form austenite. Throughout the cooling cycle, that solid-state transformation further led to a local volume expansion of 4.5% in the lattice. Consequently, the phase transformation generated comprehensive residual stress may have led to the reduction of tensile residual stress [36].
With regard to the trend of residual stress change, it should be noticed that the peak of tensile transverse residual stress is relatively smaller than the maximum tensile longitudinal residual stresses in both experimental and simulated data (Fig. 2.24 (b) and (c)). Therefore, it could be suggested that expansion and shrinkage during heating and cooling cycles along the weld seam are slightly greater than perpendicular to the weld centerline. This difference resulted in buckling and was the dominant distortion in the welded samples. Ilman et al. [37] found the similar trend of the residual stress profiles for welds of Al5083 plates that were produced by arc welding.

Figure 2.24 Numerically-predicted and experimentally-measured residual stress profiles along the mid-plane section on top surface of the weld and 3D distribution of residual stress contours in (b) longitudinal and (c) Transverse direction.
One of the interesting features regarding the FE mechanical analysis is the detailed spatial and temporal information about the evolution and distribution of residual stress. Fig. 2.25 presents the evolution of the residual stress profiles as a function of the welding time along the line AB. As observed in Fig. 2.25, the line AB crossed the middle of the weld line. The approximate time when the heat source reached the line AB is 2.2 s. When the heat source was away from the line AB (t<2.2 s), the residual stress field was almost zero along the line AB. When the heat source approached the line AB, the temperature of the material near AB went up, resulting in the expansion of the heated material. The expansion was inhibited by the surrounding material, that led to the increase in the longitudinal compressive stress (2.2s<t<3 s). In comparison to the longitudinal stress, the transverse residual stress experienced little change in the welding region with the approaching heat source. This result showed that expansion of the heated material in the transverse direction was much lower with respect to the expansion in the longitudinal direction. When the heat source passed the line AB (t>3 s), the expanded material began to contract due to cooling of the material. The contracted region was constrained by the surrounding area, resulting in a decrease the transverse and longitudinal compressive stress. Thus, the compressive stress was gradually replaced by the tensile stress in the welding region when the weld cooled down.
Figure 2.25 Residual stress vs. welding time for a laser power of 7.5 kW and a welding speed of 35 mm/s. (a) transverse stress, (b) longitudinal stress and (c) Von Mises equivalent stress, and (d) temperature distributions at a time of 2.2 s.

2.4.4.2. Weld distortion evaluation

To analyze the thermally-induced out-of-plane deformation that was generated inside the weld specimen, the distribution of displacement in the Z-direction was numerically and experimentally evaluated. The results of FEM simulation and experimental measurement are presented in Fig. 2.26. As clearly seen, there is a relatively good consistency between experimental and numerically predicted results. The maximum measured displacement occurred at the centerline, and it was around 1.5 mm. A slightly lower value was predicted by simulation results (1.2 mm). It was found that the inherent out-of-plane deformation at FZ was the outcome of two dominant thermally-induced strains in the transverse and longitudinal directions. The transverse contraction that happened at the FZ would scale up the angular distortion and was significantly dependent on the type and amount of residual stress [16]. According to residual stress profiles (Fig. 2.24), the concentrated tensile residual stress around
the FZ could be considered as the part of the response with respect to the transverse contraction force developed through the weld bead during the cooling cycle. However, the rest of the mentioned response was released through the centerline as the permanent out-of-plane deformation.

![Figure 2.26](image)

**Figure 2.26** (a) Numerically predicted distribution of out-of-plane Z-displacement with 5X magnification and (b) numerically- and experimentally-obtained Z-displacement profiles along the welding direction.

### 2.4.5. Microstructural evaluation

The cross-section macrograph of but welded joints generated under an optimized condition is presented in Fig. 2.27. As visually observed, the HLAWL process produced the symmetric and fully-penetrated weld. A smooth transition from the welding region toward the base metal without undercut or excessive reinforcement was observed. Regarding the chemical composition of HSQT steel, the alloying elements like Mn, Cr and Ni leads to an increase the hardenability of this type of steel, while making the weld sensitive to the formation of cold cracks [12]. However, there was no sign of cracking either on the bead surface or inside the weld. The obtained results could be attributed to the positive role of arc to moderately reduce the amount of cooling rate and avoid the formation of cold cracks. To analyze the phase transformation throughout the FZ and HAZ, the microstructural changes from FZ toward the base metal are shown in high magnification in Fig. 2.28. As can be seen, tiny martensite lath
was formed as the dominant phase throughout the FZ. The highest average hardness measured inside the FZ (Fig. 2.14) could be correlated to the single-phase microstructure of martensite for this area (Fig. 2.28.d). A similar result was reported for HLA W of high-strength low alloy (HSLA) steel by Cao et al. [12]. They observed that a relatively high cooling rate of the HLA W process promoted the formation of a diffusion-less martensitic microstructure. It was found that acicular ferrite is the destructive phase that speeds up the crack nucleation and results in cold cracking and hydrogen embrittlement in the weld of high-strength steel. The acicular ferrite phase was more often observed in joints generated by pure laser welding where the cooling rate was excessively high. However, it was reported that due to the moderate cooling rate of the arc, formation of acicular ferrite was postponed thermodynamically during the HLA W process [12]. Thus, absence of cracks inside the FZ might be referred to as non-formation of acicular ferrite in this area. The transition of thin martensite lath to relatively thick martensite morphology was the main feature of FZ/HAZ interface under higher magnification (Fig. 2.28.b and c). A mixture of the relatively coarse grain size of residual austenite and thick martensite phase was the characteristic feature of the HAZ. The lowest average hardness in this area could be accounted as the solid evidence for the multiple-phase microstructure of HAZ, that contained residual austenite and marteniste (Fig. 2.28.b). Eventually, inside the base metal the tempered martensite phase was detected as the main phase that remained intact during the welding process (Fig. 2.28.a).
The produced microstructure throughout the FZ and HAZ could be explained by a study of the numerically-predicted cooling curves that were generated by thermal analysis. The curves were incorporated into a continuous cooling transformation (CCT) diagram that was developed for HSQT steel [38]. According to Fig. 2.29, the critical cooling rate (CCR) that is required to produce the fully-marteniste microstructure should be higher than $7 \times 10^4 \, ^\circ \text{C/s}$. According to the temperature histories that were experienced by FZ and HAZ, the cooling rates of both areas were much higher than the required CCR for the CCT diagram. Therefore, based on the CCT diagram, the anticipated phase inside both FZ and HAZ should be composed of martensite as the predominant phase (Fig. 2.28.b and d). The obtained result was well consistent with the microstructural evolution that occurred in the welding region.
Figure 2.28 Microstructural details of different zones from FZ toward the base metal

Figure 2.29 Numerically-predicted cooling curves generated by SYSWELD simulation as well as CCT diagram of HSQT steel [38].
2.5. Conclusion

In this study, the welding feasibility of thick HSQTS in zero-gap butt- and T-joint configurations by HLAW was investigated. The parameters of the HLAW were adjusted to get a sound non-porous joint. The optimal weld was analyzed in terms of micro-hardness, microstructure, and tensile test evaluations. Also, a three-dimensional thermo-metallurgical-mechanical finite element model based on commercial SYSWELD software was introduced in order to study the temperature distribution, thermal histories, generated thermally-induced residual stresses, and distortions of the joint made by the HLAW process. The following conclusions were made:

1. For a butt-joint design, fully penetrated welds were obtained at a welding speed of 35 mm/s and the optimum $D_{LA}$ (3 mm). Reduction of arc voltage was capable of decreasing the amount of undercut.

2. For a T-joint design, a weld with porosity was found at higher welding speeds. Reduction of the welding speed down to 10 mm/s effectively removed the porosity. The $D_{LA}$ in the y-direction was considered as an important factor for controlling the number of unfilled areas, and 3 mm was the optimal value to eliminate the unfilled area.

3. The numerically predicted temperature distribution inside the FZ and HAZ agreed well with the molten pool geometry, and the softened area that was measured by micro-hardness tests, respectively. A slight discrepancy between the simulated peak temperatures and data recorded by thermocouple could be attributed to the initial assumptions of FEM and instrumental errors.

4. Microstructural evaluation throughout the welded joint disclosed that FZ and HAZ were mainly composed of martensite and non-transformed austenite phase, respectively. The results agreed with the microstructural phases predicted by the CCT diagram and the numerically-produced cooling curves.
5. Mechanical analysis showed a low prediction of errors and a relatively good correlation between the experimentally-measured and numerically-predicted results of the residual stress fields. Furthermore, the welded plate distortion was around one millimeter, confirming low thermal shrinkage induced by the HLAW process.
REFERENCES


CHAPTER 3

HYBRID LASER/ARC WELDING OF AISI 304L STAINLESS STEEL TUBES IN ORBITAL JOINT CONFIGURATION: PORE MITIGATION, THERMAL ANALYSIS AND MECHANICAL INSPECTION

3.1. Introduction

Since the middle of the 20th century, austenitic stainless steels (ASS) have been used extensively in a wide variety of industries. Usage includes oil and gas transmission lines and pressure vessels in the chemical industries. In the construction of structural parts in the fusion reactors, ASS provides superior corrosion resistance coupled with acceptable mechanical properties at elevated temperatures [1-2]. In practice, an increasing demand for stainless steel in circular hollow sections makes ASS welding inevitable. Consequently, girth-welded pipe joints are frequently required due to their long geometry with respect to diameter and the wall-thickness of pipes [3]. Conventional fusion welding, such as gas metal arc welding (GMAW), submerged arc welding (SAW), and gas tungsten arc welding (GTAW), more often leads to additional challenges. These challenges include hot cracking, a large heat affected zone (HAZ) and inability to meet the fast-growing requirements of the manufacturing throughput [4]. Furthermore, the inherently low thermal conductivity of ASS is one-third of that of mild steel at ambient temperature. Coupled with a high thermal coefficient of expansion that is one and a half times larger than that of mild steel, ASS results in a large weld-induced residual stress and distortion [5]. This issue is exacerbated in the presence of a considerable amount of heat input.
As a result, additional distortion correction and reworking are often required. In terms of weld-induced microstructural evolution, precipitation of chromium carbides along the grain boundaries within the HAZ makes the joints sensitized to the intergranular corrosion. This failure mechanism also deteriorates the mechanical properties. The results is premature failure through the service condition [6]. Therefore, many attempts have been made to reduce HAZ significantly through the application of welding procedures with a highly concentrated heat source.

Among the novel fusion joining methods, electron beam welding (EBW) and laser beam welding (LBW) have received great attention for joining thick sections of ASS [7,8]. Advantages including highly focused heat input, deeper penetration, smaller HAZ, and lower distortion give a unique status to both methods. Albeit, EBW can be hardly accessible for large-scale girth welding without a large vacuum chamber. Moreover, a tightly rigorous gap tolerance should be considered in the joint [9]. LBW has the capability to be automated easily and generate joints at higher welding speed. Unlike EBW, LBW can be performed in an ambient atmosphere that eliminates the requirement of reworking [10]. Moreover, the new generation of disk lasers incorporate remarkably high beam quality while operating at high power. This development is considered fundamental for welding gauges exceeding 8 mm [11]. Despite all advantages, several common challenges are presented in high-power autogenous laser welding of thick sections. These challenges include undercut on the top surface of the weld, formation of brittle phases due to excessively high cooling rate, and inherently low gap-bridging ability [12].

Recently, the hybrid laser arc welding technology (HLAW) has gained significant attention because of its many benefits [13-15]. HLAW renders a remarkable outcome over pure laser welding and arc welding alone. In fact, HLAW incorporates the benefits of the laser welding including high efficiency, deep penetration, and higher travel speed with the benefits
of arc welding such as its high gap-bridging ability and the possibility to modify the chemical composition of the weld by filler wire. The advantages offered by arc welding are critical, especially for ASS. The weld zone as well as HAZ are susceptible to localized corrosion. Therefore, it is necessary to change the chemical composition of the weld area in such a way as to retain corrosion resistance of the joint [16]. Literature has suggested that the grade of ASS filler should be different from the type of ASS substrate [17]. It is well-known that the best synergistic effect between arc and laser-induced plasmas can be formed once the distance between arc and laser (D_{LA}) is 2-3 mm. At this optimal distance, it was shown that perturbation of keyhole formation that resulted from turbulence in the molten pool was minimal [18].

One of the most important issues concerning the laser welding of thick sections of ASS is the formation of porosity. Porosity is considered a hidden weld defect in deep-penetrated laser welds that can negatively influence the integrity of the joint. The type of porosity in laser welds is more often key-hole induced porosity. It was shown that once the laser radiation is completed if the solidification rate of the molten pool at the upper part of the keyhole exceeds the back-filling speed of liquid metal at the same part, the porosity is found at the middle and root side of the keyhole [19]. Much effort has been made to mitigate and remove the keyhole-induced pores. [20-21]. All proposed methods were based on how to prolong the solidification rate or reduce the amount of the molten pool volume at the upper part of the keyhole [22-25]. Application of electromagnetic force has been shown to be able to scale up the back-filling speed of the weld pool during the keyhole collapse process [22-23]. Wang et al. [22] applied the axial magnetic field to improve integrity of the HLAW weld of 304 stainless steel. It was found that the magnetic field enlarged the weld width and made it easier for pores to escape. Chen et al. [25] reported that a static magnetic field could modify the microstructure and suppress pores significantly for the HLAW joint of 304 stainless steel by controlling the backfilling speed of the molten pool. However, application of a static magnet during girth
welding of pipe may challenge the robustness of the process and make the interaction of processing parameters more complicated. Another suggested method to resolve the keyhole-induced pores was associated with the positioning configuration of the two heat sources along the welding direction [24-25]. There are two arrangements including the arc-push hybrid process with the laser beam leading the arc and the arc-dragging hybrid process with the arc preceding the laser beam. Cao et al. [24] studied the effect of heat source positioning on joint quality of 8 mm thick high-strength steel plates made by hybrid welding. It was determined that the arc-pushing HLAW could produce wider top width with less porosity defects than arc-dragging HLAW. Katayama et al. [25] reported that for the laser-leading HLAW, the arc pressure depresses the molten pool and removes most of the bubbles generated from the keyhole.

Numerical approaches are mostly used to investigate the temperature field in the HLAW process. They help better understand the welding process in terms of penetration, generated residual stresses, thermal induced cracking, and deformation. The advantages of utilizing FE simulation to analyze the thermal field and residual stresses have been discussed in a number of publications. The majority of FE studies in HLAW processes were conducted on flat butt joints [26]. In pipe joining several works in either single pass or multi-pass welds were conducted by means of GMAW, TIG, and submerged arc welding processes to study the temperature field and finally thermally induced residual stresses [27]. With the recent developments in utilizing the laser-GMAW hybrid process for pipeline girth welding, the studies on thermomechanical simulation ASS steels are scarcely limited [28-29].

The purpose of this study is primarily to investigate the feasibility of the girth welding of A304L stainless steel tubes by HLAW in two relative positions of the laser with respect to the arc to attain a free-pore weld. It must be noticed that no effort at optimization of processing parameters was made. To describe the heat distribution in the weld for both of the arc-laser
arrangements, numerical simulation using commercial ANSYS code was proposed. Then, the micro-hardness and the tensile strength of samples were analyzed to reveal the relation between the microstructural and mechanical attributes.

3.2. Experimental procedure

The chemical composition of the 8 mm thick A304 stainless steel tubes and the filler wire used for the HLAW process are shown in Table 3.1. Fig. 3.1 shows the hybrid arc laser welding equipment used to weld tubes. A diode-pumped Yb-YAG disk laser (TRUMPF, Trudisk 10003 laser) with a continuous wavelength of 1064 nm and maximum power of 10 kW was employed for the HLAW process. Since the ASS surface has high reflectivity once being exposed to laser beam, the laser head was inclined 6° with respect to its vertical axis to protect the focusing and collimating lenses from the back reflection. To avoid extra machining, the type of weld configuration was determined to be a butt-joint configuration. A CLOOS power source (Quinto GLC 403) was used for the GMAW process. To achieve the motion required along welding path, the wire feed nozzle and laser head were installed on a 6-axis high-precision KUKA KR 60-3 robot as shown in Fig. 3.1. A 98%Ar-2%CO₂ gas was used to shield the molten pool from the surrounding atmosphere. As mentioned in previous chapter, the main purpose of the CO₂ presence in the shielding gas was to stabilize the arc. As presented in Fig. 3.1, the gas protection design includes two configurations of post-shielding and root shielding.

| Table 3.1 Chemical composition of A304L steels and consumed wires. |
| ----------------- | --- | --- | --- | --- | --- | --- | --- | --- | --- | --- | --- |
| Designation      | C   | Mn  |P | S  | Cu | Si | Ni | Cr | Mo | N   | Fe  |
| SUS 304L         | 0.03 | 2  | 0.045 | 0.030 | -  | -  | 8  | 18 | -  | 0.1 | balance |
| ER 308LSi        | 0.01 | 1.8 | 0.030 | 0.030 | 0.1 | 0.9 | 10.5 | 19.9 | 0.15 | 0.1 | balance |
In order to attain a defect-free weld for tubes with the diameter of 280 mm, all experiments were conducted initially on plates. These plates were in the same condition as the tube in terms of the material type and thickness as well as the square butt-joint configuration. In addition to the welding speed, two heat source arrangements of the arc-leading ALHW (arc-pull) and laser-leading LAHW (arc push) processes, as presented schematically in Fig. 3.2, were studied to evaluate the effect of the position of heat source on the weld quality. The
obtained welding parameters extracted from the plates welding were directly applied on girth welding of the tube. A manual TIG welding process was used to facilitate the clamping of tubes and reduce the distortion of the welded tubes. The joint path was tack welded manually at equal intervals of 110 mm by (Fig. 3.2).

![Figure 3.2](image)

**Figure 3.2 (a) Real entire test built-up on the pipe, (b) tacking the joint by using manual TIG and (c) schematic positioning of the laser head with respect to the tube.**

To analyze the welded joint in terms of microstructure, the welded coupons were sectioned longitudinally and transversely by an abrasive water-jet cutting machine equipped with CNC. Then, the prepared cross-sections were polished and etched with a specific solution (an aqua mixture of 20 gr Ammonium Bi-fluoride, 0.5 gr Potassium meta-bisulfite, and 100 ml distilled water). An optical microscope and a scanning electron microscope (SEM) equipped with energy dispersive spectroscopy (EDS) chemical analysis were employed to observe and characterize chemically the microstructure of the weld cross sections. The micro-hardness test was conducted by using a Clark Micro-hardness tester with 1 kg force load under 15 seconds of dwell time. Tensile properties of the joints were examined by an Instron mechanical testing machine at the constant strain rate of 1mm/min. An X-ray diffractometer (XRD) was utilized
to determine the type of phases that were formed within the FZ. The X-ray source was Cu Kα generated at 25 kV and the fixed scan rate of 1°s⁻¹ was in the direction of the weld bead. According to ASTM E975-03 [30], a comparison method calculates the volume fraction of δ-ferrite (Cₐ) and γ-austenite (Cᵧ) within the weld of stainless steel. Based on this method, the integrated intensities of the diffraction peaks for (110) and (220) crystallographic planes of δ-ferrite and the (111), (200) and (220) diffraction peaks for γ-austenite were considered. Thus, the following equations were developed to calculate the Cₐ and Cᵧ [30]:

\[
C_a = \left[ 1 + \left( \frac{I_{\gamma(111)}}{182.8} + \frac{I_{\gamma(200)}}{81.6} + \frac{I_{\gamma(220)}}{44.4} \right) \right]^{-1} + \left( \frac{I_{\delta(110)}}{233.8} \right) + \left( \frac{I_{\delta(220)}}{31.9} \right) \tag{3.1}
\]

\[
C_y = \frac{I_{\gamma(111)}}{I_{\gamma(111)} + I_{\gamma(200)} + I_{\gamma(220)}} \times 100 \tag{3.2}
\]

where \( I_{\gamma(hkl)} \) and \( I_{\delta(hkl)} \) are the integrated intensities of a certain crystallographic plane (hkl) from the γ-austenite and δ-ferrite phases, respectively.

3.3. Results and discussion

3.3.1. Developing the processing parameters to mitigate the pores

The aim of the study was to generate a visually sound fully-penetrated weld bead coupled with a small and uniform back reinforcement carried out in single-pass HLAW. The chart of weld acceptance criteria is shown schematically in Fig. 3.3. Since there were quite a large number of variables coming in contact with each other during the HLAW process, the following solutions were implemented to mitigate the pores:

1. Faster travel speed and larger laser power was used to create a longer and enlarged weld pool. The solution resulted in a slower solidification rate and facilitated outflow of bubbles from the molten pool.

2. The best heat source arrangement was selected.
3. Laser ramping was applied at the end of the process

Based on the above approaches to produce defect-free weld, a number of experiments were performed under different welding speeds. The corresponding welding parameters in each condition are summarized in Table 3.2. Based on the type of shielding gas and wire feed rate, current and voltage were specified automatically by an algorithm provided by the CLOOS GMAW machine. The longitudinal macrograph of the welds at the weld face as well as weld root under various welding speeds are presented in Fig. 3.4. As can be observed, under lower welding speed and lower laser power the reinforcement root was not uniform. To consider the effect of net heat input that was obtained by both arc and laser the following equation was utilized [14]:

\[ HI_t = \frac{\eta_{PL} + \eta_{AVI}}{\nu} \]  

(3.3)

where \( P_L \) is the laser power, \( V \) is the arc voltage, \( I \) is arc current, \( \nu \) is the welding speed, and \( \eta_A \) and \( \eta_P \) are the efficiencies of the GMAW and laser, respectively. Based on the literature for ASS, \( \eta_A \) was assumed to be 0.8 and \( \eta_P \) was to be 0.9 [14]. According to Table 3.2 and Fig. 3.4, the faster welding speed improved the stability of the molten pool and resulted in uniform penetration. Similar results were found in [13] when the heat input was not large enough, metal dripped periodically at the root side of the weld. This drip caused sagging at the front side. Scaling up the welding speed and the laser power solved the above issue. Under both medium and high welding speeds, an even root reinforcement was produced. Extra experiments at medium and high welding speeds were performed in the arc-push mode. These experiments were conducted to study the influence of heat source positioning on weld quality. In regard to visual inspection, it was noticed that welds produced through arc-push were shinier, wider and smoother so that the rippling along the bead edge was reduced significantly (Fig. 3.5).

The weld beads were inspected in terms of pore formation, the main issue concerning the laser welding of ASS steels. The welds that were in a visually finalized condition were
sectioned along the bead. Fig. 3.6 presents the pore formation inside the weld bead. The formation might be caused by two probable reasons. The small spherical pores with a diameter of less than 0.25 mm could be formed due to the entrapment of the gasses inside the molten pool. These types of pores were located adjacent to the top surface of the weld. The large and irregular shape of the pores with a diameter larger than 0.25 mm were formed at the middle and root part of the bead. This type of pore formation could be related to the collapse of the keyhole. An average number of pores per length of the weld and average pore diameter accompanied by suggested pore mitigation solutions are indicated in Table 3.3. As can be observed, increasing the welding speed reduced the number of irregular-shaped pores that could be related to the instability of the keyhole. On the other hand, scaling up the welding speed had no significant role in the mitigation of the small spherical pores. It was reported [31] that under the condition of a low welding speed, the keyhole rear wall was more unstable and larger bubbles were produced. However, the faster welding speed resulted in more stability of the keyhole and a lower formation of large bubbles. This result merits further research in the field of online molten-pool monitoring. Cross sections of the weld were examined to understand the effect of welding speed and heat-source arrangement on the shape of the molten pool, and how to reduce the porosity. The transverse cross sections of the weld with respect to the weld bead were prepared and presented in Fig. 3.7. As can be seen, the wine-cup shape was the main characteristic observed in all the welds. This shape disclosed that the laser dominated to full-penetration, while the arc that melted the wire was principally restricted to the upper part of the weld. In addition, a faster welding speed induced the reduction of the width of arc area for both heat source configurations. Chang et al. [32] reported that increasing the heat input coupled with the welding speed raised the turbulence intensity. In turn, more energy transferred from the top part of the pool to the bottom part of the pool due to faster heat transfer. The enhanced turbulence inside the molten pool would enlarge the volume of the lower part of
pool and hasten the pore elimination rate. Based on Chang’s hypothesis, the geometrical features of the weld could be used to specify the absorbed energy ratio of the lower region to upper region of the weld pool. The volume characteristic coefficient of the molten pool, φ can be defined by following simplified equation:

$$\varphi = \frac{S_{\text{lower}}}{S_{\text{upper}}}$$  \hspace{1cm} (3.4)

where $S_{\text{lower}}$ and $S_{\text{upper}}$ are the transversal cross-sectional areas of the lower and upper regions, respectively. The smaller difference between the upper and lower regions of the weld pool leads to the higher φ. It was investigated that once φ exceeds 0.52, the pore percentage reaches to less than 0.5% [33]. Based on the geometrical features of weld cross-sections, faster welding led to the larger value of φ, from 0.45 to 0.55, implying that a lower number of pores was generated.

Regarding Table 3.3, the heat source arrangement didn’t drastically influence the mitigation of large pores. Nevertheless, the arc-push mode scaled down the small pores that could be related to entrapment of gas bubbles. According to Fig. 3.7 for all welding speeds, the wider molten pool face was generated for the case of arc push. It was found that arc pressure expands the molten pool face and accelerates the outflow of gaseous bubbles that are close to the surface of the molten pool, thereby, decreasing the number of small pores [34].

To study the mechanical damage based on pores, a resistance to the load-carrying capacity (R-LCC) was defined [35]. Under different types of loading modes, such as impact and tensile the above concept can be beneficial to quantitatively determine the integrity of the entire joint. The integrity of the joint is determined with respect to the number and size of pores incorporated within geometrical features of the weld. For the total number of the pores that are formed along the longitudinal cross-section, the total R-LCC value of a weld can be calculated based on the following equation [35]:

$$109$$
\[ R - \text{LCC}_{\text{total}} = \sum_{1}^{n} \frac{2\sigma D_p \left( \frac{W_{\text{max}}}{2} - D_p \right)}{2 + 1.18(1 - \frac{D_p}{W_{\text{max}}})^4} \]  

where \( \sigma \) expresses a hypothetical stress value depending on the effective stress applied to the joint during service conditions (\( \sigma = 100 \) MPa). \( D_p \) indicates the diameter of the pore and \( W_{\text{max}} \) presents the maximum weld width, depending on the type of welding conditions. Under each welding condition, the average value \( R - \text{LCC}_{\text{total}} \) was calculated based on three times measurement. The lowest value of \( R - \text{LCC} \) was used as the criteria for the optimized welding condition [21]. In fact, smaller and fewer pores resulted in a lower \( R - \text{LCC} \) value and a more efficient role of the corresponding welding condition to scale down the pores. Thus, the high-welding speed coupled with the arc-push mode lead to a joint with the least amount of the pores. This result implied that the joint had the highest resistance to the load-carrying capacity.

The developed processing parameters for a flat plate welding in arc-push mode were transferred to the tube welding in girth butt-joint configuration. The main issue concerning tube welding was a large pore that collapsed at the keyhole. This issue occasionally occurred at the end of the welding process when the laser power sharply dropped to zero (Fig. 3.8.). It was shown that an abrupt reduction of laser power leads to a high solidification rate, preventing a complete filling of the keyhole and resulting in the formation of a large pore with the diameter larger than 1 mm [36]. Using the laser power ramping removed this type of pore. The ramping coincided with overlap area at a specific time at the end of the welding process. Laser ramping delayed the solidification process, thereby, increasing the liquid backfilling time and reducing the chance of a keyhole-induced pore. Besides, the relatively deep concave that was generated at the ending point of the weld was relatively filled. Therefore, the process made a uniform bead with respect to the surrounding area (Fig. 3.8).

Fig. 3.9 shows the appearance of the final girth welded joints that were produced in the arc-push mode. As could be expected, the face weld was reflective and wide and there was no sign of excessive reinforcement .According to the ASME Boiler and Pressure Vessels 2015
Section V Article 2 [37], the welded tubes were inspected by a non-destructive radiography examination to reveal probable presence of pores. No porosity or cracks were detected. This result confirmed the overall high quality of the welded tubes.

![Image of weld acceptance criteria](image)

**Figure 3.3 Flow of weld acceptance criteria**

**Table 3.2** Welding parameters for the hybrid laser-arc welding of 304L stainless steel under different welding speeds in arc-pull position

<table>
<thead>
<tr>
<th>Welding condition</th>
<th>Laser Power (kW)</th>
<th>Welding speed (mm/s)</th>
<th>Wire feed rate (mm/min)</th>
<th>Current (A)</th>
<th>Voltage (V)</th>
<th>Shielding gas flow rate (l/min)</th>
<th>Root Shielding gas flow rate (l/min)</th>
<th>Stand-off distance (mm)</th>
<th>Heat input induced by arc (kJ/m)</th>
<th>Heat input induced by laser (kJ/m)</th>
<th>Heat input (kJ/m)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Low speed</td>
<td>4</td>
<td>15</td>
<td>3.5</td>
<td>95</td>
<td>20.3</td>
<td>28</td>
<td>18</td>
<td>3-3.5</td>
<td>103</td>
<td>180</td>
<td>283</td>
</tr>
<tr>
<td>Medium speed</td>
<td>4</td>
<td>25</td>
<td>4</td>
<td>219</td>
<td>26.5</td>
<td>28</td>
<td>18</td>
<td>3-3.5</td>
<td>186</td>
<td>144</td>
<td>330</td>
</tr>
<tr>
<td>High speed</td>
<td>7</td>
<td>35</td>
<td>10</td>
<td>258</td>
<td>27.4</td>
<td>28</td>
<td>18</td>
<td>3-3.5</td>
<td>162</td>
<td>180</td>
<td>342</td>
</tr>
</tbody>
</table>
Figure 3.4 Effect of welding speed on weld appearance of bead at face and root side.

Medium welding speed: 25 mm/s

High welding speed: 35 mm/s

Figure 3.5 Effect of heat source arrangement on weld face in different welding speeds.
Figure 3.6 Typical longitudinal cross-sections of the welds produced under different conditions.

Table 3.3 Influence of welding speed and heat source arrangement on average number of porosity per length of the weld and average pore diameter

<table>
<thead>
<tr>
<th>Welding condition</th>
<th>Width of the weld face (mm)</th>
<th>Average number of large pores (&gt;0.25mm) per length of the plates (150 mm)</th>
<th>Average number of small pores (&lt;0.25mm) per length of the plates (150 mm)</th>
<th>Average pore diameter size (mm)</th>
<th>R-LCC_total</th>
</tr>
</thead>
<tbody>
<tr>
<td>Medium speed-Arc pull</td>
<td>6</td>
<td>6</td>
<td>7</td>
<td>0.39</td>
<td>1974</td>
</tr>
<tr>
<td>Medium speed- Arc push</td>
<td>8</td>
<td>4</td>
<td>2</td>
<td>0.23</td>
<td>1483</td>
</tr>
<tr>
<td>High speed-Arc pull</td>
<td>5</td>
<td>1</td>
<td>5</td>
<td>0.29</td>
<td>906</td>
</tr>
<tr>
<td>High speed-Arc push</td>
<td>7</td>
<td>0</td>
<td>2</td>
<td>0.15</td>
<td>435</td>
</tr>
</tbody>
</table>
Figure 3.7 Typical weld cross-sections produced under different conditions.

Figure 3.8 Typical formation of keyhole-collapsed pore at the end of process and pore elimination by using laser ramping within the overlapping area.
3.3.2. Numerical simulation of HLAW in two arrangements of arc-laser position

3.3.2.1. Governing equations

In this part the thermal analysis was employed to obtain the temperature field in a hybrid laser/arc welding process. The transient energy conservation was used as the governing equation to evaluate the thermal history during the HLAW process:

\[
\frac{\partial (\rho(T) \times C_p(T) \times T)}{\partial t} = \text{div} \left( k(T) \nabla T \right) + \dot{q}_{\text{laser}}(x, y, z, t) + \dot{q}_{\text{arc}}(x, y, z, t)
\]  \hspace{1cm} (3.9)

Where, \( t \) is time, \( \rho \ (T) \) is the temperature-dependent material density, \( C_p \) is temperature-dependent specific, and \( K \ (T) \) is the temperature dependent thermal conductivity. Although for simplicity the convection term was ignored, the thermal conductivity was modified at temperatures above the melting point to compensate for the convection flow of the molten
weld. The laser induced power distribution was expressed as the conical volumetric heat source equation as follows:

\[
\dot{q}_{\text{laser}}(x, y, z, t) = \eta \frac{2P_{\text{laser}}}{\pi r_0^2 Y} \exp\left(1 - \frac{(x-x_0)^2 + (z-vt \pm SD)^2}{r_0^2}\right)(1 - \frac{y}{Y})
\]  

(3.10)

Where \(\eta\) is the laser absorption efficiency, \(P_{\text{laser}}\) is the nominal power of laser beam, \(r_0\) is the effective radius of laser beam, \(Y\) is the thickness of material, \(v\) is the welding speed, \(x_0\) is the \(x\)-coordinate of the laser spot’s center point, and \(SD\) is the stand-off distance between laser and arc heat sources. The laser absorption coefficient for the welded material can be evaluated with respect to Bramson’s equation [43]:

\[
\eta = 0.365 \left(\frac{R}{\lambda}\right)^{1/2} - 0.0667 \left(\frac{R}{\lambda}\right) + 0.006 \left(\frac{R}{\lambda}\right)^{3/2}
\]

(3.11)

where \(\lambda\) is the wavelength of laser beam (1030 nm for disk laser), and \(R\) is the electrical resistivity. Given that \(R\) is temperature-dependent, the average value of 80\(\mu\Omega\) cm was used for 304 stainless steel [43].

To express the input power of the GMAW process the double-ellipsoidal heat source [44] was used. The distribution of heat in front and rear parts of heat equation are given as follows:

\[
\dot{q}_{\text{arc}}^f(x, y, z, t) = \frac{6\sqrt{3} f_r P_{\text{arc}}}{abc \pi \sqrt{\pi}} e^{-3(x-x_0)^2/a^2} e^{-3(y-Y)^2/b^2} e^{-3(z-vt)^2/c_f^2} \quad \text{for } z > vt
\]

(3.12)

\[
\dot{q}_{\text{arc}}^r(x, y, z, t) = \frac{6\sqrt{3} f_r P_{\text{arc}}}{abc \pi \sqrt{\pi}} e^{-3(x-x_0)^2/a^2} e^{-3(y-Y)^2/b^2} e^{-3(z-vt)^2/c_r^2} \quad \text{for } z < vt
\]

(3.13)

Where \(P_{\text{arc}}\) is the nominal arc power. \(a\), \(b\), \(c_f\), and \(c_r\) are the semi-axes of the double-ellipsoid represented by depth, front length, and rear length, respectively.

The initial condition of the specimen was considered at room temperature \((T (x,y,x,0)=20^\circ\text{C})\). The heat loss of the model surfaces through convection and radiation was
combined according to the Vinokurov’s empirical relationship [45]. This lumped heat transfer equation can be expressed as eq. (3.14)

\[
h_{eff} = 2.4 \times 10^{-3} \varepsilon T_{1.61}
\]  

(3.14)

Where \( h \) is the lumped effective heat transfer coefficient and \( \varepsilon \) is emissivity.

3.3.2.2. Heat source validation and thermal analysis

Due to hindrances in experimentally capturing the thermal history of tube welding process and numerically requiring higher computational time, an alternative approach was employed. Namely, to simplify the calibration process of the heat source’s parameters and validation of the thermal simulations results, the flat plates of the same material with the same processing parameters in a butt joint configuration were used. Fig. 3.10 presents the dimensions of the plate with the locations of four thermocouples. It should be noticed that the same geometry was used for both arrangements of arc-pull and arc-push. The FE software package, ANSYS, was used to obtain the temperature distribution in HLA welding of stainless steel plates. The moving heat source was determined by considering the heat distribution equations. These equations were defined by the user-written APDL subroutine. The model was partitioned away from the weld into three regions where the element size was increasing as the distance from centerline increased. These partitions provided the optimum number of elements. The element type for regions 1 and 3 was the 8-node hexahedral element, SOLID70. The element type for the connecting region was a 10-node tetrahedral element, SOLID87. The element size plays a crucial role in the accuracy of FE simulation. For this reason, the independency of thermal results with respect to the mesh size was investigated. The mesh refinement process was conducted by a series of FE simulations under the same conditions, except for the mesh size, to obtain independent results from the element size. The captured maximum temperature did not vary too much by increasing the number of elements above 81375 (Fig. 3.11) This
number of elements was expected to be the optimum condition to obtain precise results with the lowest computational time. The dimensions of the smallest brick elements in the weld area were 0.5×0.2×0.33 mm. The temperature dependent thermal properties for stainless steel plates and filler material are presented in Table 3.4.

Figure 3.10 Dimensions (mm) of finite element model, size of brick elements in fusion area, and location of thermocouples.

Figure 3.11 Influence of element number on maximum temperature
Table 3.4 Thermo-physical properties of 304L steel and filler material [50]

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>Density (kg m⁻¹)</th>
<th>Cp (J kg⁻¹°C⁻¹)</th>
<th>k (W m⁻²°C⁻¹)</th>
<th>εR</th>
<th>Density (kg m⁻¹)</th>
<th>Cp (J kg⁻¹°C⁻¹)</th>
<th>k (W m⁻²°C⁻¹)</th>
</tr>
</thead>
<tbody>
<tr>
<td>25</td>
<td>7900</td>
<td>462</td>
<td>14.6</td>
<td>0.1</td>
<td>7880</td>
<td>500</td>
<td>20</td>
</tr>
<tr>
<td>100</td>
<td>7880</td>
<td>496</td>
<td>15.1</td>
<td>0.11</td>
<td>7880</td>
<td>505</td>
<td>21</td>
</tr>
<tr>
<td>200</td>
<td>7830</td>
<td>512</td>
<td>16.1</td>
<td>0.12</td>
<td>7800</td>
<td>510</td>
<td>22</td>
</tr>
<tr>
<td>400</td>
<td>7750</td>
<td>540</td>
<td>18</td>
<td>0.13</td>
<td>7760</td>
<td>600</td>
<td>22.5</td>
</tr>
<tr>
<td>600</td>
<td>7660</td>
<td>577</td>
<td>20.8</td>
<td>0.15</td>
<td>7600</td>
<td>610</td>
<td>22.8</td>
</tr>
<tr>
<td>800</td>
<td>7560</td>
<td>604</td>
<td>23.9</td>
<td>0.17</td>
<td>7520</td>
<td>610</td>
<td>23</td>
</tr>
<tr>
<td>1000</td>
<td>7532</td>
<td>642</td>
<td>28.8</td>
<td>0.18</td>
<td>7390</td>
<td>610</td>
<td>23.8</td>
</tr>
<tr>
<td>1200</td>
<td>7370</td>
<td>676</td>
<td>32.2</td>
<td>0.19</td>
<td>7300</td>
<td>610</td>
<td>24</td>
</tr>
<tr>
<td>1300</td>
<td>7320</td>
<td>692</td>
<td>33.7</td>
<td>0.19</td>
<td>7250</td>
<td>610</td>
<td>24</td>
</tr>
<tr>
<td>1550</td>
<td>7300</td>
<td>700</td>
<td>120</td>
<td>0.2</td>
<td>7180</td>
<td>610</td>
<td>24</td>
</tr>
</tbody>
</table>

Based on the macro-graphical images of cross section of the welds, the appropriate values for heat source parameters were chosen. The values were expected to achieve numerical results close to the experimental observations. The accuracy of thermal analysis was verified by the weld-bead cross-sectional views and temperature histories during the welding and cooling processes. The comparison of numerically predicted weld pool geometries and experimentally obtained ones are shown in Fig. 3.12. The numerically-predicted temperature histories and thermocouple recorded data at corresponding positions TC1 to TC4 are given in Fig. 3.13. This figure depicts well the match between the FE analysis and experimental results in both welding arrangements. These reasonable agreements between the experimental and numerical results confirm the accuracy of the selected heat source distributions, their constant parameters, and the specified boundary conditions. In this modeling some simplifications were considered and the molten pool was treated as a solid phase to obtain the temperature distribution. Therefore, some inaccuracies were expected. As clearly shown in Fig. 3.13, some deviations were detected at the peak temperatures of some thermocouples. Other deviations
were along the shape of the fusion zone. These deviations can be attributed to the modeling assumptions.

Figure 3.12 Comparison of cross sectional views of fusion zone in modeled temperature counter and welded joints micrograph a) arc-push b) arc-pull.
Figure 3.13 Thermal history curves at TC1 to TC4 thermocouples a) arc-push b) arc-pull, measurement vs simulation.

The validated FE model was developed for the thermal analysis of stainless steel tubes with diameter of 280 mm as shown in Fig. 3.14. Due to the symmetry condition only half the welded tube was considered. The dimensions of elements in this geometry was exactly same as the size of elements in the model of flat plates. This 3D tube model contains a total number of 384420 nodes with associated 365116 elements. Similar to flat plates, the non-uniform meshing approach was implemented in which fine elements were used in welding region and
coarse sizes were used with increasing distance away from centerline of the weld. The rotation angle $\theta$ is defined by element length and incremental time steps to provide the same experimental welding speed. The heat source equations and their constants were then calculated with respect to the new Cartesian coordinate system $(x_0, y_0, z_0)$ in local position (Fig. 3.14).

**Figure 3.14 3D meshed finite element model**

The validated parameters of heat sources for arc-push welding mode were exactly implemented in tube thermal simulations. In this study, total time step was calculated based on the whole weld length over the element length with respect to the experimental welding speed. Fig. 3.15 demonstrates the thermal history as well as peak temperature at three specific location (90˚, 180˚, and 270˚) from the starting point. Due to the similar behavior of temperature over the time and minor fluctuation of its maximum value at different angles, it can be concluded that angular movement over the circumferential of tubes hasn’t considerable effect on the thermal history. The weld-pool features were also depicted in cross-sectional and top views at these selected positions in which validates the full penetration condition in welding process. The weld-bead width, molten pool length, and other characteristics can also be obtainable in this way.
The effect of the heat source arrangement on temperature gradient (G) was evaluated. An example of variation of this parameter with respect to a distance that begins within the FZ and proceeds toward the base metal for a typical cross section in the specific time-interval is shown in Fig. 3.16. As could be expected, for both conditions the temperature gradient inside the molten pool was much larger than within the base metal. This result confirmed the fact that the area close to the laser beam experiences a higher cooling rate. However, a highly elevated temperature gradient that was numerically calculated within the FZ of the arc-pull mode resulted in a higher cooling rate compared to the arc-push mode. The underlying reason behind
this behavior was illustrated by Cao et al. [24]. It was reported that during the arc-push mode, a larger amount of laser energy is dedicated to melting the base metal, which lowers the evaporative ejection of alloying elements and causes a smaller expulsion of the molten pool. The result is a moderately hot molten pool. In contrast, for arc-pull mode the trailing heat source (laser beam) may lead to the considerable expulsion of the molten pool, thereby, making molten pool hotter. Therefore, the temperature gradient from the molten pool/keyhole interface toward the fusion boundary was relatively steeper for the case of arc pull.

![Graph](image)

**Figure 3.16** The variations of temperature gradient (G) that were generated from numerical simulation from the FZ toward the base metal for arc pull and arc push arrangements.

### 3.3.3. Microstructural characterization

Fig. 3.17 shows the solidification microstructures in the arc-induced fusion zone for both the arc-pull and arc-push modes. As can be observed, the microstructure was mainly composed of bright columnar δ-ferrite dendrites. The dendrites were growing in epitaxial mode
from the fusion boundary toward the center of the bead in the opposite direction of the heat transfer. The δ-ferrite dendrites were uniformly distributed inside the dark austenite matrix for both cases.

The phase identification within the weld bead was studied by X-ray diffraction (XRD) analysis as shown in Fig. 3.18. Based on the XRD patterns, the weld bead in the two heat source configurations and in the base metal contained the γ-austenite as the main phase and δ-ferrite as the minor phase. The volume fraction of the δ-ferrite phase for all samples was calculated according to Eqs. (3.1) and (3.2) and they are presented in Table 3.5. The calculated amount of δ-ferrite for welds was larger than for the base metal. The amount of δ-ferrite could be contributed to the rapid solidification during the welding process that postponed the diffusion-based transformation of δ-ferrite to austenite. As a result, the calculated amount of retained untransformed δ-ferrite within the FZ was higher than that inside the base metal. Moreover, the amount of retained δ-ferrite inside the FZ for the arc-pull mode was slightly higher than in the arc-pull mode. This result could originate from the higher cooling rate of the arc-pull mode that was previously predicted by numerical simulation.

It is noteworthy that the microstructure of arc-pull mode was finer than that of the arc push mode. The distance between the dendrites ranged from 10 µm to 20 µm for the arc-pull mode. In contrast, this value for the arc-push mode was measured in the range of 20 µm to 40 µm. The underlying reason for this difference could be due to the cooling rates that were experienced by the molten pool in different conditions. Based on the obtained results of thermal analysis, it was predicted that the molten pool of the arc-pull mode experienced the steeper temperature gradient. The larger the temperature gradient, the higher the cooling rate. Thus, the smaller dendrite spacing was found in the arc-pull mode.
The other interesting point concerning the arc-induced solidified zone was the variation of δ-ferrite morphology from FZ interface toward the center of the molten pool. The lathy/vermicular morphology was found to be dominant for the area close to the FZ boundary. However, the skeletal type was recognized as a main morphology for the region far away from the interface. Kristensen et al. [38] reported that once the cooling rate was moderate only skeletal morphology was formed. However, when the cooling rate was high, the δ-ferrite lathy morphology would form due to the limited diffusion during the ferrite-austenite transformation. It was shown that once the diffusion lengths were restricted, it was more probable for the solid-state transformation to go forward as more tightly spaced laths. These lathes caused to ferrite to be retained. The retained ferrite then split the primary dendrite [28]. Actually, the base metal side at the fusion interface acted as an efficient heat sink and induced the higher cooling rate at the FZ interface compared to the center of the molten pool, resulting in the formation of δ-ferrite lathy morphology adjacent to the FZ boundary.

Figure 3.17 Microstructures of the arc-induced area of HLA weld joints for various heat source arrangements
Figure 3.18. X-ray diffraction spectrum of the base metal as well as the cross-section of the weld beads of various heat source arrangements.

Table 3.5 Volume fraction of δ-ferrite within the raw and welded samples.

<table>
<thead>
<tr>
<th>Stainless Steels</th>
<th>C_δ (%) Eq. (1)</th>
<th>C_γ (%) Eq. (2)</th>
</tr>
</thead>
<tbody>
<tr>
<td>AR 304L</td>
<td>23</td>
<td>69</td>
</tr>
<tr>
<td>Arc-Push weld</td>
<td>27</td>
<td>60</td>
</tr>
<tr>
<td>Arc-Pull weld</td>
<td>29</td>
<td>57</td>
</tr>
</tbody>
</table>

Fig. 3.19 shows the microstructure of laser-induced solidified zone at the FZ/base interface. The width of HAZ composed of a coarse grain size was less than 200 µm, which is in accordance with similar studies in the field of HLAW [2-39]. The obtained feature is the typical positive effect associated with a highly concentrated heat source induced by laser. The weld joint microstructures within the upper zone and the lower region of the laser-induced area for different heat source arrangements are presented in Fig. 3.20. It is evident that the heat
source positioning has no profound effect on microstructural features of the laser-induced zone. For both heat source arrangements, the dendrites are finer in the lower area in comparison to the coarser grains that were formed within the upper area. The higher cooling rate induced by pure laser in the lower part of weld refined the microstructure with respect to microstructure in the upper region of the weld. This upper region experienced a lower cooling rate caused by the combination of arc and laser [39].

The whole solidification structure of the weld was analyzed through the solidification physics and can be specified by the weld pool variables like grain growth rate (R) and temperature gradient (G) [40]. The evolution solidification map of the weld for FCC-based alloy is quantitatively presented in Fig. 3.21 [41]. The grain growth rate (R) can be derived from welding speed (V) according to the following equation:

\[ R = V \cdot \cos(\theta) \]  

(3.15)

where \( \theta \) is the angle between the welding direction and the normal of the FZ/base interface that can be considered as the isothermal surface. Regarding different inclination of FZ boundary with respect to the welding direction, the R-value has the range of 0 to 35 mm/s. To calculate a rough value of temperature gradient, the temperature within the center of the molten pool and FZ/base interface should be considered. The laser-induced plume in the vaporized column of the high-temperature ionizing metal filled the keyhole. For the iron-argon system, it was reported that the maximum temperature within the core of vaporized column was 15000-19000 K [42]. On the other hand, the temperature of the FZ boundary could be expected to be equal to the iron melting point (1774 K). As a result, the temperature gradient (G) should be around \( 3.3 \times 10^6 \) to \( 4.4 \times 10^6 \) K/m. Thus, the high ratio of G/R for HLAW within the solidification map resulted in pure columnar dendrite growth inside the weld. The results were in accordance with the observed microstructure.
Figure 3.19 Optical microstructure of the FZ interface at the laser-induced region.

Figure 3.20. Optical microstructure of the FZ at upper and lower regions of the laser-induced part.
3.3.4. Effect of relative position of arc with respect to laser on mechanical properties

3.3.4.1. Micro-hardness

Micro-hardness measurements were conducted on the top and bottom part of the weld joint for both heat source arrangements (Fig. 3.22.). As can be seen for both conditions, the micro-hardness at the bottom part was slightly larger than at the top part. The higher micro-hardness could transpire from a finer microstructure in the laser-induced part of fusion zone. The weld obtained at the upper part was a result of the interaction of arc and laser. However, the bottom part was the product of a pure laser. Due to highly concentrated nature of the laser, it induced a higher cooling rate in the bottom of the weld, whereas, arc slowed down the cooling rate in the upper zone, resulting in the slight rise of the hardness at the root area of weld. The similar results were found by Casalino et al. [47]. Moreover, according to Fig. 3.20 the primary dendrite arm spacing of the columnar structure is lower at the bottom part of the weld. Based
on the Hall-Petch equation, hardness increases by reduction of dendrite arm spacing [48]. Unlike the micro-hardness profiles that were obtained for hybrid welding of high-strength steels in previous research [13], the hardness profile shows no significant change inside the FZ with respect to the substrate, resulting in no sign of softening behavior for weldments of ASS.

![Micro-hardness profiles](image)

**Figure 3.22 Micro-hardness profiles of the joint cross-sections for ALHW and LAHW conditions.**

### 3.3.4.2. Tensile properties

In order to examine the strength of the welding joint and study the influence of relative position of arc with respect to the laser beam on the joint performance the tensile test was performed. To achieve this purpose, the transverse tensile behavior of the weld was measured three times for each condition. The engineering stress versus elongation curves for ALHW and LAHW conditions accompanied by corresponding fracture regions are presented in Fig. 3.23. The average tensile strengths of ALHW and HLAH were almost identical and were close to the strength for the as-received material (Table 3.6). There was only a small difference between the elongations of the tensile coupons that could be negligible. Also, it is clearly shown that the fracture of coupons for both conditions occurred far away from the joint inside the base metal. This result shows that the tensile properties of the weld were much superior to those of the base metal. Sun et al. [15] reported the same fracture location at the base metal for welded
coupons of ASS304 that were generated by a cold-wire assisted laser welding. To analyze the fracture mode of the tensile specimen, the fracture surface was observed by the SEM (Fig. 3.24). Based on captured images at low magnification, surface slip and the necking down area were found for fractured surfaces. This finding took into account the typical feature of the cup-cone-shaped fracture surface. At higher magnification, the fracture surface was composed of two parts: the fibrous region in the middle and the surrounding smaller shear slip region. A small number of tearing ridges as well as the relatively small sized dimples dispersed around the coarse dimples were observed throughout the fibrous part of the fractured surface. The detected features proved that the joint fails in a ductile mode under tensile loading.

**Table 3.6** Tensile properties of the as-received tube and welded coupons from the tube joint at different conditions

<table>
<thead>
<tr>
<th>Type of joint</th>
<th>Hardness (HV)</th>
<th>Tensile properties</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Maximum in FZ</td>
<td>Minimum in HAZ</td>
</tr>
<tr>
<td>As-received</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>tube</td>
<td></td>
<td></td>
</tr>
<tr>
<td>ALHW-308LSi</td>
<td>217</td>
<td>178</td>
</tr>
<tr>
<td>LAHW-308LSi</td>
<td>219</td>
<td>187</td>
</tr>
</tbody>
</table>
Figure 3.23 Engineering stress-strain curves of weld coupons obtained by ALHW and LAWH processes.

Figure 3.24 SEM fractograph microstructure of raptured specimen of (a) ALHW and (b) LAHW in different magnifications.
3.4. Conclusions

In the current study, application of single-pass high-speed hybrid laser-arc welding process resulted in a pore-free fully-penetrated weld as well as the formation of uniform reinforcement at the root side of the weld joint. The main outcomes of this investigation have been drawn:

1. Using the arc-push mode configuration caused smoother, wider, and shinier weld beads with a lower number of gas-induced pores since the arc pressure expanded the molten pool face, resulting in accelerating the outflow of gasses.

2. The temperature evolution and molten pool geometry for different heat source arrangements were studied by using FEM. The higher cooling rate predicted by FEM for the arc-pull mode showed a good match with the observed microstructures.

4. The columnar growth of austenite dendrites was observed as the predominant microstructural feature of FZ that was consistent with the morphology obtained by the solidification map. The width of the HAZ was less than 200 µm that was a solid evidence for the highly focused heat input as well as the high cooling rate caused by a high welding speed.

5. Micro-hardness profile of optimal weld for both heat source arrangements was mostly uniform. No sharp fluctuation of either hardening or softening occurred inside the FZ and HAZ, that could be attributed to the slight refinement of microstructure and non-formation of hard phase inside the FZ.

6. Weld joints produced by HLAW showed higher strength with respect to the base material so that the failure location of tensile coupons happened outside of the joint. Fibrous fracture surface confirmed the ductile fracture mode as well as excessive plastic deformation during yielding that could correspond to the larger elongation value of the failed joints.
REFERENCES


4.1. Introduction

Austenitic chromium-nickel corrosion resistant steel (CRS) is one of the most popular materials that has been applied extensively in the fields of petrochemical, power generation, liquefied natural gas (LNG) interior structures, and in the medical industries. CRS has provided outstanding weldability, formability, and excellent general corrosion resistance against a hostile environment [1]. In practice, to join structural parts, welding has been used as one of the most important manufacturing processes. However, during the most commonly used types of fusion welding processes; i.e., gas metal arc welding (GMAW) and submerged arc welding (SAW), a large heat input by arc will negatively influence the microstructural homogeneity [2]. It was observed that the formation of coarse grains coupled with chromium carbides along the grain boundaries within the heat affected zone (HAZ) significantly dwindle the mechanical integrity of the final joint [3,4]. Moreover, because of the high thermal expansion coefficient and inherently large mushy zone (the difference between the solidus and liquidus temperatures), the fused weld of CRS is highly vulnerable to solidification cracking [5].

By introducing a laser characterized by highly focused, monochromatic, and directional properties, this type of heat source has been vastly developed to weld CRS. The advantages
offered by autogenous laser welding (ALW), are high penetrability and a narrow heat affected zone. Despite these benefits a precise fit-up requirement with a very narrow gap tolerance has restricted the ALW application particularly for thicker sections of CRS [6]. As an alternative technology, a hybrid laser arc welding process (HLAW) has been widely expanded. HLAW not only enables the welding process under fast speed and reduces the number of weld passes but also removes the requirement for a tight fit-up [7]. The benefits of HLAW transpiration occur during the physical interaction of laser and arc. It has been shown that laser plasma stabilizes the arc plasma and vice versa [8]. Such synergetic effect between these two heat sources improves the efficiency of energy absorption, thereby, increasing the weld penetration. Furthermore, it has been proved that the moderate cooling rate of HLAW restricts the formation of brittle phase that gives rise to hot cracking susceptibility [9].

Pitting corrosion is considered as the main cause of premature failure of CRS in corrosive industrial media. Due to the intrinsic localized aspects of pitting corrosion, pit generation is limited to much smaller regions with respect to the entire exposed surface. After the integrity of protective Cr$_2$O$_3$ film is disrupted very small low-depth pits are formed. These pits accelerate the rate of corrosive ion attack toward the bulk of material until complete devastation of the weld occurs [10]. It is generally accepted that solidification-induced micro-segregation leads to formation of a local galvanic cell between Cr-rich and Cr-depleted zones, and results in the initiation of micron-scaled pits [11]. One of the most critical concerns regarding the welding of CRS is metallurgical heterogeneity and microstructure alteration during the cooling phase. It has been revealed that the pitting corrosion resistance of laser-welded CRS, particularly in an aggressive halide environment, may exacerbate from micro-segregation, undesirable phase content, and loss of Cr by vaporization. Cr being the main alloying element that forms a protective Cr$_2$O$_3$ film on the surface [12].
Numerous variables have a drastic effect on the pitting corrosion behavior of the weld. The final alloying elements in the fusion zone (FZ) affect the solidification mode, and type and amount of phases inside the weld metal [13]. It was reported that the formation of δ-ferrite phase has a crucial role in reducing the hot cracking sensitivity of the final weld. However, an excessive amount of that phase has a detrimental effect on pitting corrosion resistance. Furthermore, it was observed that δ-ferrite/γ-austenite interface was the best place for nucleation of stable pits [14]. It was found that the addition of Cr, N, and Mo elements could retain the localized electrochemical resistance of the CRS weld [15]. Lu et al. [16] claimed that the welding processes associated with large heat input are more likely to cause the micro-segregation of main alloying elements and formation of Cr-depleted areas in an S304L weld. Agha Ali et al. [17] reported that repeated repair welding of AISI316L increased the sensitivity of both FZ and HAZ. Consequently, the resistance of entire joint to pitting and crevice corrosion deteriorated. Kwok et al. [12] pointed out that the pitting corrosion behavior of the laser welds of AISI316 was far superior to the common AISI304 type. This difference could be attributed to the presence of a larger amount of Mo inside the S316 weld.

One of the interesting points concerning HLAW is a capability to render chemical modification of FZ by adding the filler wire through arc melting. Such a possibility may come into consideration as a potential way to improve the pitting corrosion of CRS welds. Currently, the study of HLAW of CRS generally centers around two critical aspects. One area is the experimental optimization of processing parameters to get a fully-penetrated weld. The second area of study is arc/laser interaction and its effect on maximize energy absorption [18,19]. Both areas are trying to increase productivity beyond what can be attained by just laser or arc welding alone. Rarely investigated, however is a rational analysis of the HLAW merit of improving the pitting corrosion resistance of the final weld. In this regard, the goal of the present work is to study the effect of wire chemical composition on microstructural evolution of the AISI304L
 CRS joint that were made by HLAW process coupled with its corrosion behavior in 3.5% NaCl halide media. The microstructural alteration of welded specimens during electrochemical analysis were characterized to explain the formation and growth of corrosion pits mechanisms. The current work would be considered an introductory baseline study for the subsequent investigations in the field of CRS laser welding.

4.2. Experimental procedure

The material conducted in the current study was an austenitic AISI304L CRS in the form of a 8-mm-thick tube with the outside diameter of 280 mm. The filler wire of different austenitic CRS grades with a 1.2 mm diameter were used. The chemical composition of the parent material and consumed wires are shown in Table 4.1. Based on the results of the previous chapter, the HLAW welding parameters were determined for flat plates and developed variables were transferred to weld the tubes of A304L in a girth configuration (Table 4.2). The experimental configuration is presented in Fig. 4.1. The hybrid welding process was conducted by a diode-pumped Yb-YAG disk laser with a continuous wavelength of 1030 nm and maximum power of 10 kW. The tubes were placed cut-side down on a lathe in order to remove any traces of the cutting process. Then, the prepared tubes were installed on a three-jaw chuck (rotary positioner). To facilitate clamping of the two tubes, the tack-welds were employed in equal intervals in lengths of 110 mm with a manual Tungsten Inert gas welding (TIG). The HLAW welding of tubes was performed at a seven-axis positioning system that consisted of a six-axis KUKA robot and one rotatory axis.

For metallographic inspection, the welded samples were cut perpendicular to the welding direction, hot mounted, polished, and etched with a mixture reagent of 20gr Ammonium Bi-fluoride, 0.5 gr Potassium meta-bisulfite, and 100 ml distilled water. The etched weld cross-sections were observed by means of an optical microscope (OLYMPUS
BX51M). The microstructure of the welded samples was further analyzed by using thermal field emission scanning electron microscopy (FEI Nova Nano SEM 230). The electron microscopy was equipped with an electron dispersive spectroscopy (EDS). To identify the probable phase within the fusion zone, an X-ray diffractometer (XRD) was performed on the planar surface of the welds produced by different wires. A Philips diffractometer (40 kV) with Cu-Kα radiation (λ=0.12406 nm) was used. XRD patterns were recorded in the 2θ range of 10-100˚ (step size 1°s⁻¹ with a scanning direction along the weld). To determine the volume fraction of δ-ferrite (Cδ) and γ-austenite (Cγ) within the weld of CRS, a comparison method based on ASTM E975-03 was used. According to this procedure, the integrated intensities of the diffraction peaks for (110) and (220) crystallographic planes of δ-ferrite and the (111), (200), and (220) diffraction peaks for γ-austenite were taken into account. The calculation was done based on the following expression [20]:

\[
C_\delta = \left[ 1 + \left( \frac{I_\gamma(111)}{182.8} + \frac{I_\gamma(200)}{81.6} + \frac{I_\gamma(220)}{44.4} \right) - 1 \right] \times 100 \\
C_\gamma = \frac{I_\gamma(111)}{I_\gamma(111) + I_\gamma(200) + I_\gamma(220)} \times 100
\]

(4.1)

(4.2)

where \(I_\gamma(hkl)\) and \(I_\delta(hkl)\) are the integrated intensities of certain crystallographic planes (hkl) from the γ-austenite and δ-ferrite phases, respectively.

To study the corrosion behavior of the welds and base metal, the cyclic potentiodynamic polarization test (CPPT) was performed based on ASTM Standard G61-86 [21]. All the welded joints were cut longitudinally. The weld surface was cold mounted in the epoxy matrix with an exposure area of 1 cm². Then, the mounted coupons were ground and polished to get a smooth mirror-like surface. The testing electrolyte was 3.5wt.% NaCl solution made by ACS NaCl and distilled water. To remove the dissolved oxygen, the electrolyte solution was injected with Ar for 30 min before starting the experiment. The three-electrode
electrochemical cell, as shown in Fig. 4.2, was used. The cell consisted of saturated calomel electrode (SCE) as the reference electrode, graphite as the counter electrode, and test sample as the working electrode. In order to stabilize the surface of test sample, all CPPTs were commenced after 2 h of immersion in an open-circuit potential condition. Then, the potentiodynamic polarization test was performed at the scan rate of 1 mVs\(^{-1}\). The test was continued until the current density reached 5 mAcm\(^{-2}\). Then the scan was inverted and extended until the loop terminated at the protection potential. To achieve reliability for each condition, the tests were repeated at least three times. After scanning the cyclic curves, the Tafel extrapolation procedure was used to calculate the value of circuit potential (\(E_{\text{corr}}\)) and corrosion current density (\(i_{\text{corr}}\)). Pitting potential (\(E_p\)) and re-passivation potential (\(E_{\text{rp}}\)) were determined as characteristic parameters to define the pitting behavior in a corrosive medium.

Table 4.1 Chemical composition of AISI304L steels and consumed wires.

<table>
<thead>
<tr>
<th>Designation</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cu</th>
<th>Si</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>N</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>AISI 304L</td>
<td>0.03</td>
<td>2</td>
<td>0.045</td>
<td>0.030</td>
<td>-</td>
<td>-</td>
<td>8</td>
<td>18</td>
<td>-</td>
<td>0.1</td>
<td>balance</td>
</tr>
<tr>
<td>ER 308LSi</td>
<td>0.01</td>
<td>1.8</td>
<td>0.030</td>
<td>0.030</td>
<td>0.1</td>
<td>0.9</td>
<td>10.5</td>
<td>19.9</td>
<td>0.15</td>
<td>0.05</td>
<td>balance</td>
</tr>
<tr>
<td>ER 316LSi</td>
<td>0.01</td>
<td>1.8</td>
<td>0.030</td>
<td>0.030</td>
<td>0.12</td>
<td>0.9</td>
<td>12.2</td>
<td>18.4</td>
<td>2.6</td>
<td>0.05</td>
<td>balance</td>
</tr>
</tbody>
</table>
Figure 4.1 (a) HLAW set-up, visual bead face at (b) top and (c) root side of the weld under optimized condition.

Table 4.2 Welding parameters for the hybrid laser-arc welding of 304L stainless steel.

<table>
<thead>
<tr>
<th>Main welding parameters</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser Power (kW)</td>
</tr>
<tr>
<td>------------------------------</td>
</tr>
<tr>
<td>7</td>
</tr>
</tbody>
</table>
4.3. Results and discussion

4.3.1. Microstructural analysis

In order to determine the solidification mode of a CRS weld joint under the optimized conditions, the pseudo-binary sections of the Fe-Cr-Ni ternary diagram at 70% Fe was utilized (Fig. 4.3) [22]. Because of the relatively high cooling rates of HLAW, this type of diagram has been widely employed to predict non-equilibrium microstructures that are produced by the laser welding process. To predict the final phases after solidification and solid-state transformations first it is required to calculate the ratio of the equivalent chrome $[\text{Cr}]_{eq}$ with respect to equivalent nickel $[\text{Ni}]_{eq}$ based on the following equation [23]:

$$[\text{Cr}]_{eq} = \text{Cr} + \text{Mo}$$ (4.3)

$$[\text{Ni}]_{eq} = \text{Ni} + 35\% \text{C} + 20\% \text{N}$$ (4.4)
Details of various solidification types with corresponding \( (\text{Cr}/\text{Ni})_{\text{eq}} \) ratios, reactions and microstructures are summarized in Table 4.3. The equivalent chrome \([\text{Cr}]_{\text{eq}}\) and equivalent nickel \([\text{Ni}]_{\text{eq}}\) were specified by the chemical compositions of the base metal and filler wires. The \( (\text{Cr}/\text{Ni})_{\text{eq}} \) was calculated to be 1.55, 1.69 and 1.79 for the ER316LSi, ER308LSi and base metal, respectively. Regardless of the dilution percentage, the predicted solidification mode of either filler wires or base metal was found to be the ferrite-austenite (FA) mode. In other words, the solidification of the weld area was initiated with the formation of the primary \( \delta \)-ferrite phase. Then, austenite nucleates from the liquid and \( \delta \)-ferrite phase were formed by eutectic and peritectic reactions, respectively. Eventually, the remaining molten pool decomposed into austenite. To validate the final type of phases that were predicated by pseudo-binary diagram the welded joint was investigated under metallography. Fig. 4.4 shows the solidification microstructure of the typical A308LSi weld metal. As can be observed, the microstructure was composed mainly of bright columnar \( \delta \)-ferrite dendrite growing in an epitaxial mode from the fusion interface toward the center of the bead in an opposite direction of the heat transfer. The \( \delta \)-ferrite dendrite was distributed uniformly inside the dark austenite matrix. The generated microstructure was consistent with the one that was predicted by the pseudo-binary diagram.

In order to identify the probable phase within the FZ, the weld bead was studied by X-ray diffraction (XRD) analysis as shown in Fig. 4.5. Based on the XRD patterns, the weld beads that were made of two wires and the base metal contained the \( \gamma \)-austenite as the matrix phase and \( \delta \)-ferrite as the secondary phase. The volume fractions of \( \delta \)-ferrite for all samples were calculated according to Eqs. (1) and (2) and they are presented in Table 4.4. For the welds, the amount of \( \delta \)-ferrite within the FZ was increased compared to the base metal. Also this value was much higher than the value predicted by the Schaeffler diagram [22,23]. To explain this phenomenon, it should be noticed that this type of diagram was prepared in the equilibrium condition. However, the nature of HLAW is highly non-equilibrium. Furthermore, regarding
the final reaction associated with the FA solidification mode \((L + \delta + (\gamma + \delta)_{per/eut} \rightarrow \gamma + \delta)\), part of final austenite was formed through solid-state transformation of \(\delta \rightarrow \gamma + \delta\) that is a highly diffusion-controlled process. The fast cooling rate in the laser-based welding process did not allow enough time to proceed with the phase transformation completely. Consequently, the presence of the large portion of initial \(\delta\)-ferrite was inevitable. It has been found that the ferrite content in the range of 20-30% reduces the hot cracking sensitivity of the joint. Thus, residual \(\delta\)-ferrite could be favorable to improve the mechanical properties [24].

Figure 4.3 (a) WRC 1992 constitution diagram with 304L and 308LSi compositions [22] and (b) Pseudo-binary diagram of the Fe-Cr-Ni ternary diagram at 70% Fe, showing solidification modes [23].
Table 4.3 Effect of $\frac{Cr_{eq}}{Ni_{eq}}$ ratio on solidification mode, type of reaction and final microstructure [24]

<table>
<thead>
<tr>
<th>Solidification mode</th>
<th>Equivalent ratio</th>
<th>Reaction</th>
<th>Final Microstructure</th>
</tr>
</thead>
<tbody>
<tr>
<td>(A): Fully Austenitic</td>
<td>$\frac{Cr_{eq}}{Ni_{eq}} &lt; 1.25$</td>
<td>$L \rightarrow L + \gamma \rightarrow \gamma$</td>
<td>Fully Austenitic</td>
</tr>
<tr>
<td>(AF): Austenitic Ferrite</td>
<td>$1.25 &lt; \frac{Cr_{eq}}{Ni_{eq}} &lt; 1.48$</td>
<td>$L \rightarrow L + \gamma \rightarrow L + \gamma + (\gamma + \delta)<em>{eut} \rightarrow \gamma + \delta</em>{eut}$</td>
<td>Austenite matrix with grain boundary Ferrite</td>
</tr>
<tr>
<td>(FA): Ferritic Austenitic</td>
<td>$1.48 &lt; \frac{Cr_{eq}}{Ni_{eq}} &lt; 1.95$</td>
<td>$L \rightarrow L + \delta \rightarrow L + \delta + (\gamma + \delta)_{per/eut} \rightarrow \gamma + \delta$</td>
<td>Skeletal and/or lathy Ferrite resulting from Ferrite to Austenite transformation</td>
</tr>
<tr>
<td>(F): Fully Ferritic</td>
<td>$1.95 &lt; \frac{Cr_{eq}}{Ni_{eq}}$</td>
<td>$L \rightarrow L + \delta \rightarrow \gamma + \delta$</td>
<td>Ferrite matrix with grain boundary Austenite</td>
</tr>
</tbody>
</table>

Weld A308LSi

Figure 4.4 (a) Cross-section of the joint made by A308LSi wire and (b) Microstructures of the weld within the arc zone area in different magnifications.
Figure 4.5 X-ray diffraction spectrum of the base metal as well as the cross-section of the weld beads made of different wires.

Table 4.4 Volume fraction of δ-ferrite in as-received (AR) and welded samples.

<table>
<thead>
<tr>
<th>Stainless Steels</th>
<th>$C_\delta$ (%) Eq. (1)</th>
<th>$C_\gamma$ (%) Eq. (2)</th>
</tr>
</thead>
<tbody>
<tr>
<td>AISI 304L</td>
<td>23</td>
<td>69</td>
</tr>
<tr>
<td>308LSi weld</td>
<td>27</td>
<td>57</td>
</tr>
<tr>
<td>316LSi weld</td>
<td>25</td>
<td>58</td>
</tr>
</tbody>
</table>

The high cooling rate of the laser-based welding process made micro-segregation. Furthermore, it may have negatively affected the corrosion resistance of the final joint. The amount of weld-induced segregation was studied inside the FZ throughout the analysis of alloying element. The alloying element distributions of ferrite and austenite phases within this area were measured by means of EDS as indicated in Fig. 4.6. As could be expected, inside the δ-ferrite dendrite the weight extent of Cr (20.56) was slightly higher than in the austenite matrix (18.34). On the contrary, the opposite results were found for the Ni distribution through
dendrite and inter-dendritic region. Moreover, to observe the transition of Cr, the substantial alloying element to improve corrosion resistance, EDS line scanning was done through the dendrite core of δ-ferrite toward the austenite (Fig. 4.7). The gradual transition of Cr from δ-ferrite phase to austenite showed no significant thermally-induced segregation.

![Average chemical composition of solidified regions](image)

**Figure 4.6** (a) SEM micrograph of HLAW weld (b) EDS average chemical composition of dendritic (δ-ferrite) and inter-dendritic (γ-austenite) areas

![EDS line scanning of Cr element across the δ/γ grain boundary.](image)

**Figure 4.7** EDS line scanning of Cr element across the δ/γ grain boundary.
4.3.2. Effect of filler wire on general and pitting corrosion behavior of the weld

The high Cr content of the stainless steel leads to the formation of Cr$_2$O$_3$ based compounds on the surface, as barrier layers. These compounds are the main reason for the remarkable general corrosion resistance of this type of alloy against the harsh environment. However, due to micro segregation emanated from welding process, it was observed that once the weld of stainless steel was exposed to the halide environment pitting corrosion was activated. This pitting reduced the physical and mechanical integrity of the joint [25].

The effect of type of wire on both the general and localized corrosion of the weld metals was studied. A cyclic potentiodynamic polarization test (CPPT) was employed in 1.5%wt NaCl under ambient temperature. This evaluation renders the valuable data that would be of interest to the industry. Thus, the selection of proper filler wire could be considered as a useful output of this electrochemical study to improve the corrosion resistance of the final weld against the Cl-ion-containing medium. Some researchers have developed theoretical equations to show a relationship between the pitting corrosion resistance and the chemical composition of stainless steels [26,27]. It is generally accepted that elements such as C, P, S, and non-metallic inclusions negatively affect the pitting resistance. However, Cr, N, and Mo slow down the rate of nucleation and growth of pits on the surface [27].

A pitting resistance equivalent number (PREN) has been developed as a common predictive procedure to rank the resistance against pitting corrosion. Generally, better pitting resistance is in accordance with a larger PREN. According to the element weight percentage of three most commonly presented elements (Cr, Mo, and nitrogen), the PREN can be calculated as follows:

\[
\text{PREN} = \text{wt}\%\text{Cr} + 3.3\text{wt}\%\text{Mo} + 16\text{wt}\%\text{N} \quad (4.5)
\]
The overall PREN numbers were calculated and are listed for un-weld and weld conditions in Table 5. From theoretical approach, the consumed wires have the capability to give a significant amelioration to the localized corrosion resistance of the final weld over the base metal. The results of CPPTs of as-received A304L stainless steel and HLAW welds that were produced by different types of wire are exhibited in Fig. 4.8. Also, Table 4.5. and Fig 4.9 represent the extracted data from the CPPT.

The corrosion behavior of samples was quantified by considering a number of concepts. The general thermodynamic stability of tested materials under the electrochemical corrosive condition was defined as corrosion potential ($E_{corr}$) [28]. The more positive, $E_{corr}$ the higher the uniform corrosion resistance. The $E_{corr}$ was compared for different conditions. The general corrosion behavior of the AR 304L base metal was slightly better than for the other materials. Furthermore, all of the examined alloys were spontaneously passive under the halide environment that was associated with the high corrosion resistance.

More accurate criteria to rank the general corrosion is the polarization resistance ($R_p$) that can be calculated in terms of the anodic/cathodic Tafel slopes ($\beta_{anodic}$ and $\beta_{cathodic}$) and the corrosion current density ($i_{corr}$) based on the following Equation [29]:

$$R_p = \frac{\beta_a \times \beta_c}{2.303 \times i_{corr}(\beta_a + \beta_c)}$$ (4.6)

$R_p$ is associated with the corrosive process between the metal and solution, which happens through the protective passivation film. Comparing the samples, the AR 304L base metal showed higher values of $R_p$. The improvement of corrosion resistance was coincidental with the decrease in $i_{corr}$. The ranking of general corrosion resistance of the tested-coupons based on $R_p$ is as follows:

Weld 308LSi < Weld 316LSi < AR 304L
Table 4.5 Average values for corrosion parameters of as-received and hybrid welded stainless steels obtained by the cyclic potentiodynamic corrosion test

<table>
<thead>
<tr>
<th>Specimens</th>
<th>$E_{corr}$ (mV)</th>
<th>$i_{corr}$ (µA/cm$^2$)</th>
<th>$E_{pit}$ (mV)</th>
<th>$E_{rp}$ (mV)</th>
<th>PREN (mV)</th>
<th>$\beta_{anodic}$ (mV)</th>
<th>$\beta_{cathodic}$ (mV)</th>
<th>$R_p$ (kΩ)</th>
</tr>
</thead>
<tbody>
<tr>
<td>AR 304L</td>
<td>-292</td>
<td>0.108</td>
<td>270</td>
<td>-78</td>
<td>19.6</td>
<td>732</td>
<td>143</td>
<td>481</td>
</tr>
<tr>
<td>Weld 308LSi</td>
<td>-313</td>
<td>0.153</td>
<td>325</td>
<td>-165</td>
<td>20.1</td>
<td>1156</td>
<td>188</td>
<td>459</td>
</tr>
<tr>
<td>Weld 316LSi</td>
<td>-297</td>
<td>0.071</td>
<td>497</td>
<td>244</td>
<td>28.6</td>
<td>131</td>
<td>184</td>
<td>468</td>
</tr>
</tbody>
</table>

The characteristic feature of CPTT scans is the study of the localized corrosion behaviors of corrosion resistant alloys. The localized corrosion behavior of the alloy is examined by breakdown (pitting) potential ($E_p$) that can be obtained from the potentiodynamic curves. The pitting potential is specified as the potential where the rate of current density is scaled up quickly with a slight increase in potential [25]. In fact, $E_p$ refers to the potential condition where stable pits begin to form and grow at an excessively high rate. The higher tendency to pitting nucleation is associated with the smaller range between the breakdown potential and corrosion potential ($E_b - E_{corr}$). This smaller range implies a narrower passivation region. As can be seen, the 316LSi weld metal showed the highest resistance against pit formation over the other samples. Regarding potentiodynamic curves of 304L base metal and 308LSi weld metal, sharp current density spikes were randomly observed in the passive potential regime.

Such localized current transients could be attributed to repeated alteration between passive and active states (Fig. 4.10). This phenomenon might be due to the formation of metastable pits, temporary growth, and annihilation (or localized breakdown and subsequent repassivation) [27]. Hakiki et al [30] explained that the existence of anodic inhibitor in the solution, as well as the formation of the double film of $\text{Cr}_2\text{O}_3$ composed of iron oxides and
hydroxides, were the main mechanisms responsible for re-passivation. The absence of local current oscillations in the passive potential range of the 316LSi weld metal can be considered as solid evidence for the relatively high $E_p$ measured for this alloy. It was reported that the growth rate of expanding pits is a function of the nucleation rate of metastable pits (or the rate of oscillations in polarization test). Therefore, the survival possibility of metastable pits drops to zero when $E_p$ reaches a certain value [31]. Thus, the 316LSi weld metal without current spikes exhibited the lowest susceptibility against pit formation.

Figure 4.8 Cyclic Potentiodynamic polarization graphs of the raw material AISI304 stainless steel and the HLAW weldments made of ER 308LSi and ER 316LSi.
Figure 4.9 Cumulative cyclic potentiodynamic data of corrosion indicators of each coupon.

Figure 4.10 Examples of current density transients on Cyclic Potentiodynamic polarization graphs as an indicator of pit initiation.
It seems that there was a good consistency between PREN predictive data and obtained results from CPTT. The good corrosion resistance of the 316LSi weld could be attributed to the presence of the Mo element. It is generally believed that Mo is a required element to recover the corrosion resistance of the weld that is already reduced due to compositional heterogeneity such as micro-segregation that is emanated from rapid melting and solidification [32]. It was revealed that Mo acts synergistically with Cr and N, thereby, promoting the durability of passive films formed on the CRS. It was proposed that the increased passivation property in the presence of Mo was related to the reduction of active dissolution of metal through adsorption by localized formation of molybdates or generation of molybdenum hydroxide. These compounds reduced the rate of anodic metal dissolution [33]. It has also been reported that Mo extremely decreases the number of de-passivation events, as confirmed by current oscillations. In addition, the combination of Mo and N synergistically remove such transients [34].

Based on Fig. 4.8, all the samples presented a positive hysteresis loop. This loop was characterized by passivation breakdown on the potential rising to the positive direction and re-passivation at or close to their associated corrosion potential $E_{corr}$. After the potential was inversed to the negative direction, the hysteresis loop in the CPTT curve showed a retardation mode in re-passivation of an existing pit. The potential where the forward and reverse scan cross was defined as re-passivation potential ($E_{rp}$). This potential divided the passivation regime into the stable and unstable passivity [27]. In other words, $E_{rp}$ refers to a point below which the metal becomes stable-passive. The potential range between $E_p$ and $E_{rp}$ did not allow the nucleation of new pits, but those pits which already formed were able to develop. In fact, the value of $E_p - E_{rp}$ determines the growth rate of existing pores. A larger value means a higher growth rate of already formed pits. Thus, the larger the hysteresis loop, the more the pitting corrosion sensitivity, and the lower inclination to re-passivation. The AISI304L base metal and
308LSi weld metal showed $E_{np}$ close to their $E_{corr}$. This result implies that these metals did not have a good re-passivation tendency. This result was more noticeable for the 308LSi weld metal. For this weld after the scan mode was changed to the negative direction, the current density still scaled up, which was the first sign that confirmed the poor re-passivation property of 308LSi weld metal. Among the characterized samples, the 316LSi weld metals exhibited the smallest hysteresis loop associated with the lowest unstable passive region.

In order to evaluate the pitting initiation mechanism, a number of CPPTs were interrupted after the applied potential reaches close to the pitting potential. Subsequently, the surfaces of the corroded coupons were observed through the optical microscope and SEM, as depicted in Fig. 4.11 and 4.12. The observation shows that before reaching the applied potential to $E_p$ a number of compounds decomposed on the surface. EDS line scanning analysis showed that those corrosion products were generated inside the Cr-depleted zone of the passive layer. The cathodic effect of the surrounding Cr-rich area against the embedded Cr-depleted compound leads to the formation of a defective passive oxide layer. This layer reduces the film resistance against the corrosive environment. In addition, the passivity breakdown could be accelerated in the presence of Cl-ions. It was reported that locally absorbed chlorine at the oxide layer surface accelerated anodic dissolution at the oxide layer/solution interface [35]. At the same time, reduction of the concentration of oxygen vacancies at the oxide layer/solution boundary leads to the upheaval in the cation migration rate from the metal/oxide layer interface toward the oxide layer/solution interface. An extremely high rate of cation diffusion will form a huge number of metal vacancies at the metal/film boundary. These vacancies will create a void and lead to a local collapse in the shape of a pit [36]. As a result, formation of the corrosion product accompanied by accumulation of chlorine close to such compounds might be the best reason for pore initiation.
Figure 4.11 SEM micrograph of the A316LSi weld metal and EDS line scanning analysis of Cr element at the corrosion product before de-passivation during CPPT.

Figure 4.12 SEM micrograph of the A316LSi weld metal and EDS line scanning analysis of Cr element at the corrosion product after de-passivation during CPPT.
After the CPPT was terminated, the pitting propagation was studied. As can be seen (Fig. 4.12), large pores were surrounded by a great many small pores. Moreover, in some areas, the corrosion products were not delaminated from the surface and covered the pores. EDS line scanning analysis of such corrosion products revealed that all of the pores were already formed within Cr-depleted areas that lost their protective characteristic of passivation. Migration of small pores toward the large ones and their coalescence could be the main mechanism that was responsible for pore propagation. The obtained results are in good accordance with prior study [14] where the Cr-depleted zone was considered as a preferential location of pitting corrosion for destructive Cl-ion attacks in CRS steels

4.4. Conclusions

This investigation detailed the microstructural changes and corrosion behavior of A304L joints that were produced by hybrid laser arc welding of CRS using A308LSi and A316LSi filler wires. The major conclusions based on the presented study are as follows:

1. XRD patterns of welds and microstructural observation revealed that all welds were mainly austenitic with the presence of a higher amount of δ-ferrite with respect to the base metal, confirming the ferrite-austenite solidification mode that was predicted by Pseudo-binary diagram. The weld experienced a higher cooling rate that hindered the solid-state diffusion-controlled base of ferrite-to-austenite transformation and larger amount of δ-ferrite was formed within welding region.

2. Based on EDS analysis inside the fusion zone, relatively uniform distribution of Cr element from γ to δ and a small difference of chemical composition between dendritic cores and interdendritic region revealed that a slight micro-segregation occurred during solidification.
3. Regarding general corrosion behavior, the AISI304L base metal showed the highest corrosion resistance compared to the 316LSi and 308LSi welds, as evidenced by the lower $R_p$ value.

4. According to the results of potentiodynamic polarization scans, all laser welds and the base metal exhibited the passivation behavior in 3.5% NaCl solution. Among all alloys, the lowest susceptibility against pitting was obtained for the 316LSi weld joint, as proved by the higher difference of pitting and corrosion potentials. Concerning pitting corrosion sensitivity and the tendency to re-passivation, it was revealed that the 316LSi weld with the smallest hysteresis loop had the highest inclination to re-passivation that was in accordance with the lowest unstable passive region.

5. Pore growth analysis showed that the Cr-depleted zone was the most vulnerable area for the attack of Cl-aggressive ions to form a pit. Propagation of small pores toward the large one was considered as the main mechanism of pit growth.
REFERENCES


Chapter 5

SUMMARY AND FUTURE WORK

5.1. Summary of welding techniques of difficult-to-weld steels

Welding of thick gauge sections of alloyed steels is a major step in production. Application of currently-developed arc-based welding methods are accompanied by a number of pre- and post-processing steps that negatively affect the efficiency and productivity. Besides, due to low penetrability of the arc, single-pass welding is barely possible and the multi-pass joining is followed by excessively large heat input. The latter issue is a critical matter for difficult-to-weld steels and results in undesired side effects such as a cracking and distortion. Compared with conventional arc welding procedures, hybrid laser/arc welding (HLAW) process has been developed to respond to the limitations that raised from individual arc and autogenous laser welding techniques. Arc-induced wider molten pool and laser-induced deeper penetration makes this process attractive to produce single-weld for thick gauge sections of steels. Furthermore, moderate cooling rate and localized heat input are the benefits that make HLAW a good alternative to join difficult-to-weld steels.

5.2. Summary of hybrid laser/arc welding of thick plates of high-strength quenched and tempered steels in various configurations

Single-pass welding feasibility of high strength quenched and tempered steels were studied by hybrid laser/arc welding process in butt- and T-joint configurations. It was observed that under high welding speed, sound full-penetrated weld was generated in butt-joint configuration. However, for the T-joint configuration the welding speed should drop off significantly to produce uniform weld bead at face and root side. Regardless of joint
configuration and optimal condition, the joint of high strength steels that were fabricated by the hybrid laser/arc welding technique was prone to formation of softened area next to fusion boundary. The softened area negatively affected the tensile properties in way that tensile coupons failed from the welding region. To predict a residual stress distribution and distortion, the experimentally-verified thermomechanical simulation was developed based on commercial SYSWELD software. It was observed that the isotherm profiles that shaped fusion zone and heat affected zone matched well with molten pool geometry and size of softened area that was measured based on the micro-hardness results. Mechanical analysis revealed a low prediction errors and relatively good agreement between the measured and predicted residual stress and permanent out-of-plane distortion. Relatively narrow tensile residual stress area induced by hybrid laser/arc welding process was the main output of mechanical analysis.

5.3. Summary of hybrid laser/arc welding of 304AISI stainless steel tubes in orbital joint configuration

Effect of hybrid laser/arc welding variables to generate the free-pore uniform weld bead of AISI304 stainless steels under single pass in orbital joint configuration was investigated. It was found that using arc-push heat source configuration coupled with faster welding speed made weld beads smoother, wider, and shinier with a lower number of gas-induced pores. The temperature distribution for different heat source arrangements were studied by using Finite Element method. It was revealed that higher cooling rate anticipated by FEM showed a good match with observed microstructures. In regard to microstructural characterization, the columnar growth of austenite dendrites was observed as the dominant phase that was well consistent with the morphology predicted by solidification map. From mechanical inspection point of view, weld joints generated by HLAW showed higher strength than the base metal so that the fracture location of tensile coupons occurred outside of the joint. Also, micro hardness profile of weld under optimal condition for both heat source arrangements was mostly even.
Relatively smooth hardness transition from welding region toward the base metal was associated with the slight refinement of microstructure and non-formation of strengthening phase within the FZ.

5.4. Summary of corrosion behavior of hybrid laser/arc joints of AISI304 stainless steel produced by different filler wires.

The corrosion resistance and microstructural changes of AISI304L stainless steel joints produced by hybrid laser arc welding under different filler wire compositions were studied. XRD patterns of joint revealed that all welds were mainly composed of austenite phase with the presence of higher amount of δ-ferrite as compared to the base metal. Due to higher cooling rate experienced by welding region, the solid-state diffusion-controlled base of ferrite-to-austenite transformation was hindered and a larger amount of δ-ferrite was decomposed. Based on the EDS analysis within FZ, relatively consistent distribution of Cr element from γ to δ showed a slight micro-segregation throughout solidification. In regard to general corrosion behavior, the AISI304L base metal showed the most superior corrosion resistance with respect to the all welds, as evidenced by the lower $R_p$ value. Concerning pitting corrosion resistance, the results of cyclic potentiodynamic polarization scans revealed that the 316LSi weld joint exhibited the lowest susceptibility against pitting, as proved by the larger difference between pitting and corrosion potential. Regarding pitting corrosion sensitivity and the tendency to repassivation, the 316LSi weld with the smallest hysteresis loop had the highest inclination to repassivation that was in agreement with the lowest unstable passive region. SEM analysis revealed that Cr-depleted zone was the most sensitive region for the attack of Cl-aggressive ions to form a pit. Merging and combination of small pits was determined as the main mechanism of pit growth.

5.5. Future work
The future study may concentrate on several aspects such as:

- Implement the optimization of the main welding variables to generate the processing parameter windows for hybrid laser/arc welding process to make it more robust particularly for large structures that are widely used in mass production.

- Integration of the on-line sensing and control equipment with hybrid laser/arc welding systems to monitor the penetration depth in real time and adjust welding variables accordingly.

- Numerical simulation of the hybrid laser/arc welding process by considering the effect of fluid effect on shape of molten pool and heat transfer to improve the accuracy of thermal and residual stress distributions.

5.6. Publications


