Microstructures and mechanical properties of Ni/Fe dissimilar butt joint welded using the cold metal transfer

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Abstract

In this study, dissimilar butt joints were formed between as-rolled Inconel 718 and SUS 316 using cold metal transfer (CMT) with ERNiFeCr-2 filler metal at a welding speed of 5 mm s\(^{-1}\) and three different welding currents (130, 160 and 190 A). The morphology, microstructure, and mechanical properties of the joints and the mechanism by which the joint interface formed were studied. The results indicate that CMT welding parameters have an effect on the weld appearance and that the weld metal exhibits better wettability and spreadability at a welding current of 160 A. The interface between the weld metal and SUS 316 base metal was characterized by applying an optical microscopy (OM) and scanning electron microscopy (SEM) equipped with energy-dispersive X-ray spectrometry (EDS). The results revealed that no obvious diffusion of Ni and Fe ions was identified across the joint interface. The fractures that initiated in the joints during the tension testing formed in the heat-affected zone (HAZ) of the SUS 316 base metal, and the high-temperature tensile strength of the joint was observed to be 424 MPa, which is approximately 87.06\% of that of SUS 316 base metal (487 MPa) and 67.19\% of that of Inconel 718 base metal (631 MPa). In terms of the elongation, the joint (21.9\%) and Inconel 718 base metal (21.2\%) and SUS 316 base metal (22.3\%) elongations were similar. For comparison, the joint also was tested at room temperature, resulting in a tensile strength and elongation of 511 MPa and 27.5\%, respectively. A spherical elemental Ni segregation morphology was clearly observed at the high-temperature fracture. Hardness studies demonstrated that the weld metal hardness is higher than that of the SUS 316 base metal.

1. Introduction

Inconel 718 consists of an austenite matrix \(\gamma\) phase, which is strengthened by a principal strengthening phase \(\gamma’\) (Ni\(_3\)Nb, DO\(_{22}\)) and an auxiliary strengthening phase \(\gamma’’\) (Ni\(_3\)(Al, Ti), L\(_{12}\)), and the metastable \(\gamma’’\) is inclined to transform into a stable \(\delta\) phase (Ni\(_3\)Nb, DO\(_{22}\)) when exposed to a temperature of over 700 °C for a long time [1]. Inconel 718 has shown great potential in the aviation and aerospace industries. Due to its favourable characteristics, Inconel 718 is also an essential high-temperature structural material required for nuclear reactors, combustion chambers and chemical equipment. These include hot corrosion resistance, oxidation resistance and an excellent strength in high-temperature environments [2–5]. However, owing to the high cost of parts made of Inconel 718 superalloy, the joining of Inconel 718 with other metal materials is essential. Nevertheless, austenitic stainless steels (SUS 316) can be regarded as a substitute for Inconel 718 in low-risk parts. SUS 316 is a cheaper structural material and is used extensively in corrosive environments, steam generating power plants and nuclear reactors because of its outstanding performance in corrosion resistance as well as its high strength at high temperatures [6, 7]. Therefore, to replace Inconel 718 with SUS 316, the application of dissimilar metal welding is unavoidable. The concern and interest in the integrity of the dissimilar metal welds have been a focus of attention. The joining of materials with various properties into a multi-material hybrid structure can achieve economic efficiency and numerous engineering benefits [8]. The welding of Inconel
718 and SUS 316 into a bimetal structure can take advantage of the exceptional performance of the two materials for utilization in a wide range of applications, such as the fabrication of components applied in high temperature environments, especially in nuclear power plants and aerospace equipment [9, 10].

However, there are still many problems that exist in the welding process and the formation of joints. First, Inconel-718 is a Ni-based superalloy strengthened mainly by the intermetallic Ni₃Nb type γ′″ and partially by Ni₃Al type γ′ precipitation. However, during solidification, the presence of the Nb also makes the alloy susceptible to severe segregation. This results in the formation of a liquid film and the precipitation of a low-melting Laves phase in the inter-dendritic region, which makes it more prone to hot cracking and can largely degrade the comprehensive properties of the joint [11]. K. Devendranath Ramkumar et al reported that the welded joints were achieved between Inconel 718 and AISI 416 by employing CO₂ laser welding. The outcomes showed that the micro-segregation of the Laves phase increased in the Nb at the fusion zone, and trivial liquation cracks or microfissuring at the HAZ of Inconel 718 were observed [12]. Patterson et al joined alloy 625 and 304L stainless steel using autogenous gas tungsten arc welding, and the welds produced employing this technology were prone to weld solidification cracking [13]. P. Mithilesh et al also studied gas tungsten arc joining of Inconel 625 to AISI 304 using a Mo-rich filler wire, and with a low heat input employed in the gas tungsten arc welding (GTAW), the hot cracking tendency was completely avoided [14]. Second, there are some differences between Inconel 718 and SUS 316 in terms of their physical properties, such as thermal conductivity, linear expansion coefficient, and melting point. The inhomogeneity of chemical compositions and microstructures across the interface of dissimilar metal welded joints may lead to their complex crack propagation behavior, and the existence of cracks has a negative impact on the performance of welded joints [15]. Xinliang Mei et al joined Inconel 718 and 316L stainless steel by selective laser melting method. They investigated the interfacial characteristics and mechanical properties of joints, and the cracks were observed on the side of the 316L stainless steel close to the interfaces. This may be due to the thermal stress at interfaces, and lower tensile strength of 316L stainless steel [16].

Currently, tungsten inert gas (TIG), welding gas tungsten arc welding (GTAW), and metal inert gas welding (MIG) are the most commonly used methods for welding Ni-based alloys and stainless steel. Compared with other methods, the CMT welding process is modified by MIG based on the short-circuiting transfer process developed by Fronius of Austria. CMT provides a controlled method of material deposition and low thermal input by combining an innovative wire feed system coupled with high-speed digital control and a spatter-free welding process [17, 18]. To realize sufficient energy to melt both the base metal and filler wire, the wire feed rate and the cycle arcing phase are controlled [19]. Schierl reported that the droplet detachment mode of the CMT process is without the aid of the electromagnetic force, unlike in the conventional MIG process, so the spatter may decrease [20]. S. Selvi reported that CMT technology has revolutionized the welding of dissimilar metals and thicker materials by producing improved weld bead aesthetics with controlled metal deposition and low heat input [21]. Therefore, CMT is a better alternative to realize the connection between dissimilar metals. Zhang et al and Cao et al focused on the application of the process in dissimilar alloy welding with a low heat input, which inhibits the brittle intermetallic compound formation [22, 23]. Varghese et al reported that the CMT welding technique could produce crack and defect free claddings of Inconel 617M on SUS 316L with great metallurgical adherence on the substrate surface [24]. However, to my knowledge, the assessment of diverse sources in terms of dissimilar metal welding indicated that few studies have yielded regarding the dissimilar welding between Inconel 718 and SUS 316 employing CMT; thus, this study provides a test basis for engineering applications of nickel-based/stainless steel dissimilar metal joints.

In the present paper, Ni/Fe dissimilar butt joints are welded using the CMT welding process is discussed systematically with the characteristics of the weld appearance, microstructural, interfacial formation mechanisms and mechanical properties. The main objective of this study was to explore the characteristics of the microstructure, the weak area of the joint, and the mechanism of crack formation at high temperature. The results can provide theoretical support for its engineering application.

2. Experiments

In this research, the candidate base materials, as-rolled Inconel 718 superalloy and SUS 316, with dimension of 200 mm × 100 mm × 3 mm, were selected to develop the dissimilar joints; the filler wire was ERNiFeCr-2, 1.2 mm in diameter. The chemical compositions of Inconel 718, ERNiFeCr-2 and SUS 316 are given in table 1. Each as-rolled Inconel 718 and SUS 316 substrate was cut at a 45° angle to yield a 90° groove angle to ensure that the weld penetration, and the gap between the two substrates was maintained at approximately 2 mm. A schematic diagram of the butt joint is shown in figure 1.

Before the experiments, the surfaces and edges of the Inconel 718 and SUS 316 substrates were cleaned with alcohol to remove the contaminants, and the welding process was carried out using CMT, due to its ideal
properties, including stable forming, lack of spatter and low heat input required [19]. High-purity Ar (99.99%) was applied as the shielding gas. Figure 2 shows a schematic of the CMT welding of Inconel 718 and SUS 316. The welding parameters chosen in the experiments are listed in table 2. The wire feeding speed was determined by the welding equipment after welding, which is equivalent to the ratio of the total wire feed length to the welding time. After welding, the butt joints were sectioned and polished, and then etched in a solution containing 2 g of CuCl₂ + 40 ml of CH₃CH₂OH + 35 ml of HCl for microstructural analysis via OM and SEM equipped with EDS. The weld and grain sizes were measured by ImageJ metallographic analysis software. The hardness and tensile strength of the joints were checked to evaluate the mechanical properties of the joints. The microhardness distribution of the joint was characterized using a microhardness tester.

**Table 1.** Chemical compositions (wt%) of Inconel 718, ERNiFeCr-2 and SUS 316.

| Material   | Ni  | Mo  | Ti  | Cr  | C   | Al  | Nb  | Mn  | Si  | P   | Cu   | Co  | Fe  |
|------------|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|------|-----|-----|
| Inconel 718| 52.97 | 2.89 | 1.0 | 18.2 | 0.05 | 0.57 | 5.3 | 0.07 | 0.32 | 0.012 | 0.26 | 0.8 | Bal. |
| ERNiFeCr-2 | 51.7 | 3.1  | 0.98 | 17.7 | 0.05 | 0.45 | 5.45 | 0.08 | 0.33 | 0.01  | 0.24 | 0.7 | Bal. |
| SUS 316    | 10.0 | 2.07 | —   | 16.54 | 0.047 | —   | —   | 1.42 | 0.45 | 0.04  | —    | —   | Bal. |

**Table 2.** Welding parameters for dissimilar metal welds.

| Welding Type                  | CMT                                  |
|-------------------------------|---------------------------------------|
| Filler wire                   | ERNiFeCr-2                            |
| Filler wire Dia. (mm)         | 1.2                                   |
| Welding current (A)           | 130, 160, 190                         |
| Wire-feed speed (mm s⁻¹)      | 4.7, 6.25, 8.2                        |
| Welding speed (mm s⁻¹)        | 5                                     |
| Flow rate of shield gas (99.99% Ar) | 15 l min⁻¹                          |

The initial arc length was approximately 3 mm [25].
The sample was tested with a load of 200 gf and a dwell time of 15 s at ambient temperature. The schematic of the microhardness testing of the butt joint is illustrated in figure 3.

Before performing the tensile studies, the reinforcement of the Inconel 718/SUS 316 dissimilar joint was removed. Figure 4 shows a schematic of a tensile test sample. For purpose of researching the effect of a 45° inclination angle of the base metals and the CMT negligible dilution rate of the fusion zone by the base materials on the high-temperature tensile strength of joints. The high-temperature tensile tests were conducted using a Gleeble 3500. The tensile sample was heated to the test temperature at a rate of 20 °C s⁻¹, and then tested at a strain rate of 0.001 s⁻¹. The room temperature tensile properties of the joint were evaluated by means of a Zwick/Roell Z100 electronic universal material testing machine at a drawing speed of 1 mm min⁻¹. The microstructures were observed using SEM, and the chemical compositions and element distributions of the fractures were inspected by EDS.

3. Results and discussion

3.1. Weld appearance

The geometry of weld forming is an important basis for evaluating welding quality. Figure 5 exhibits the weld appearance of the CMT Inconel 718/SUS 316 dissimilar joints under different welding currents. The reinforcement, weld width, areas, and wetting angle of the weld metal were analysed. It can be seen that the smaller the reinforcement and wetting angle, the larger the weld width, and the better the wettability and spreadability of the weld metal, which can contribute to a good weld appearance. Figure 6 shows the weld geometry characteristics of the joints. When the welding currents is increased from 130 A to 190 A, the reinforcement of the weld and areas of the weld metal increase from 1.97 mm to 2.43 mm and 13.20 mm² to 28.52 mm², respectively. The weld width increases from 5.27 mm to 9.96 mm and then decreases to 8.52 mm, and the wetting angle decreases from 50° to 46° and then increases to 68°. When the welding current is 130 A, the weld is much narrower, and bonding defects can be observed near the fusion line (figure 5(a)). This may be attributed to insufficient heat input under the low welding current, which reduces the melting amount of the filler metal. Therefore, it is difficult for the filler metal to wet Inconel 718 and SUS316 surfaces. At a higher welding current of 160 A, the weld the appearance is improved greatly, the weld width is 9.96 mm, which is the largest width observed throughout the testing, and the wetting angle is the smallest at 46°. This indicates a better wettability and spreadability of the weld metal and that the weld formed uniformly, as shown in figure 5(b). No macroscopic defects are visible in the joint. However, when the welding current is further increased to 190 A, due to the higher heat input, the weld exhibits unsatisfactory characteristics. At this current, the weld lacks symmetry, and the wetting angle (68°) is the largest in comparison to the others observed. The wettability of the weld metal is poor, as seen in figure 5(c).
3.2. Microstructures

In figure 7(a) shows the microstructure of the SUS 316 base metal, and equiaxed grains of austenite can be observed. The average grain size in the SUS 316 base metal is $\sim 31 \, \mu m$. Similarly, the microstructure of Inconel 718 base metal is fine-grained. It consists of a $\gamma$ matrix and has an average grain size of $\sim 15 \, \mu m$. The grain boundaries of a few particles are shown in figure 7(b), and the EDS analysis results of point B are shown in table 3. It is evident from the analysis that the particles are Nb-rich carbides. The columnar dendritic microstructure and equiaxed crystal in the weld center can be seen in figure 7(c), in which the average grain size is $\sim 9 \, \mu m$. Micro-segregation occurred in the weld center. The content of Nb and Ti supplemented and Ni, Fe and Cr decreased at point C (micro-segregation region) comparing with the point C1 ($\gamma$ matrix), which confirmed the composition of the Laves phase. This indicates that Nb had redistributed into inter-dendritic

![Figure 5. Weld appearance of dissimilar joints under different welding currents (a) 130 A, (b) 160 A and (c) 190 A. Welding speed of 5 mm s$^{-1}$.](image)

![Figure 6. Influence of the welding currents on the (a) reinforcement and weld width; (b) areas and wetting angle of the weld metal.](image)
regions during solidification; the content of Nb in the liquid metal reached a specific component ratio and produced an Nb-enriched phase [2]. Figure 7(d) shows the typical $\gamma$ + Laves eutectic, which is distributed along the grain boundaries in a chain.

Figure 8(a) shows the microstructure of the HAZ of the SUS 316 base metal near the fusion line. The grain size of this region is larger than that of the SUS 316 base metal, and the average grain size is $\sim 43 \mu m$. This suggests that severe grain coarsening occurred in the HAZ of the SUS 316 base metal under the influence of the welding heat cycle. Figure 8(b) shows the microstructure of the HAZ away from the fusion line. The grain size shows a decreasing trend with the decrease in welding heat, with an average size of $\sim 38 \mu m$. This may affect the mechanical properties of the joint, which will be discussed in the following section.

Figure 9(a) shows the interfacial microstructures with a welding current of 160 A. The plane crystal appears in the weld near the fusion line, and a widespread dendritic microstructure is observed in the weld zone. The columnar dendritic microstructure with primary dendrites growing perpendicular to the fusion line can be observed clearly. To understand the diffusion of the main elements across the interface region between the weld metal and SUS 316, an EDS line scan test was employed to analyse the alloy element distribution. As shown by the black line in figure 9(b), the EDS line scan results show the distribution profile of Fe and Ni, and an increase in the Fe content and a decrease in the Ni content can be observed across the interface, as shown in figure 9(c). The width of the fusion line is approximately 5.1 $\mu m$. The EDS results are shown in table 4. The chemical composition at point $B_4$ in the weld metal is 24.06 wt% Fe, 47.53 wt% Ni and 17.95 wt% Cr, and the chemical composition at point $B_2$ on the fusion line is 61.25 wt% Fe, 9.69 wt% Ni and 17.58 wt% Cr, which is similar to that at point $B_3$ (66.27 wt% Fe, 10.13 wt% Ni, and 16.12 wt% Cr) in the Fe-based base metal. This indicates that

| Location | Nb  | Ni  | Fe  | Cr  | Ti  | C   |
|----------|-----|-----|-----|-----|-----|-----|
| A        | 4.33| 13.49| 62.07| 18.07| —  | —   |
| B_1      | 6.99| 55.81| 16.8 | 18.64| 0.87| —   |
| B_2      | 55.96| 4.82| 5.65 | 3.74 | 8.89| 19.87|
| C_1      | 3.22| 53.58| 19.29| 19.83| 0.75| —   |
| C_2      | 22.74| 43.71| 12.41| 14.47| 1.47| —   |
Figure 8. OM microstructure image of the HAZ located (a) near the fusion line (b) away from the fusion line. Welding current of 160 A.

Figure 9. (a) OM microstructure image of the interface between the weld metal and SUS 316; (b) SEM image of the analysed region and EDS line scan location; (c) EDS line scan results of the interface. Welding current of 160 A.

Table 4. EDS results of each position of the interface (wt%).

| Location | Cr  | Fe  | Ni  |
|----------|-----|-----|-----|
| B₁       | 17.95 | 24.06 | 47.53 |
| B₂       | 17.58 | 61.25 | 9.69  |
| B₃       | 16.12 | 66.27 | 10.13 |
the content of Fe, Ni and Cr in the vicinity of the fusion line changes less or that there is no obvious exchange of Ni and Fe ions near the fusion line.

Figure 10 shows the solidus and liquidus of the Ni-Fe alloy calculated using the equilibrium phase diagram, which indicates that the melting point of the Ni-based alloy (1,453 °C) is lower than that of the Fe-based alloy (1,538 °C); in the Ni-based alloy,

The melting point of the alloy decreases with the increasing of Fe, for the Fe-based alloy, the alloy melting point decreases as Ni increases. The melting point is at its lowest (1,442 °C) when the Ni:Fe ratio is 9:11.

In one period of the thermal cycle of the welding process, the surface of the Fe-based side was melted and came into contact with the Ni-based molten pool of filler metal. In this period, the Ni/Fe interface cooled at a faster rate and caused the elements to hardly diffuse.

3.3. Mechanical properties
Figure 11 illustrates the microhardness along the cross-sectional area of the Inconel 718/SUS 316 dissimilar joint. The hardness shows an overall decreasing trend from the Inconel 718 base metal towards the SUS 316 base metal. The hardness exhibits a major change across the interface of the weld metal and SUS 316. Inconel 718 base metal is a solution strengthened Ni-based superalloy with a higher microhardness (201–214 HV) than that of SUS 316 base metal (152–171 HV); the microhardness measured in the weld metal region (192–218 HV) is
approximately the same as that of Inconel 718 base metal and is higher than that of SUS 316 base metal. However, the hardness values of the weld metal vary, which may be attributed to the variation in the supersaturation of Nb and Mo strengthening elements in the $\gamma$-matrix [14]. The length of the weld metal is found to be approximately 2 mm.

To better understand the mechanical performance of the joints, a high-temperature tensile test was carried out on the joint and base metal samples. The Inconel 718/SUS 316 dissimilar joint was obtained at the welding current of 160 A. The outcomes of the room temperature tensile tests show that the joint fractured at the HAZ of the SUS 316 side, and the length of the upper region of the fractured location to the fusion line is approximately 0.35 mm, as shown in segment A in figure 13(a). However, the joint was tested at high-temperature; in this case, the fracture also occurred in the SUS 316 HAZ. The distance from the upper position of the fracture to the fusion line is nearly 0.32 mm, as shown in segment B in figure 14(a). The hardness values also match well with the tensile results of the joints in which tensile failures occurred in the SUS 316 base metal. Figure 12(a) represents a typical stress-strain curve of tensile tested base metals and Inconel 718/SUS 316 dissimilar joint; (b) the corresponding tensile properties of the samples tested at room and high temperature. Welding current of 160 A, RT: room temperature.

![Figure 12](image12.png)

Figure 12. (a) Type stress-strain curve of the tensile tested base metals and Inconel 718/SUS 316 dissimilar joint; (b) the corresponding tensile properties of the samples tested at room and high temperature. Welding current of 160 A, RT: room temperature.

![Figure 13](image13.png)

Figure 13. (a) Cross-section of the sample fractured in the HAZ of the SUS 316; (b) SEM image of the fractured zone. Welding current of 160 A, tested at room temperature.

approximately the same as that of Inconel 718 base metal and is higher than that of SUS 316 base metal. However, the hardness values of the weld metal vary, which may be attributed to the variation in the supersaturation of Nb and Mo strengthening elements in the $\gamma$-matrix [14]. The length of the weld metal is found to be approximately 2 mm.

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Figures 13(b) and 14(b) represent the SEM fractography of the tensile failure samples, and the tensile fracture mechanism of the samples is ductile. The fiber region exhibits an equiaxed dimple morphology. As a result of SUS 316 has a good plasticity and forms many large and deep dimples at the tensile fracture surface. The dimples on the fracture surface formed at room temperature are obviously bigger and shallower than those formed at high temperature, and the tearing edges between the dimples are obvious. Meanwhile, micro-voids are observed...
on the high-temperature fracture surface, which may be attribute to microscopic defects and the precipitated phase in the SUS 316 base metal. As shown in figure 14(c), an obvious spherical morphology can be observed on the high-temperature fracture surface. This spherical morphology, as examined by EDS results (figure 14(d)), is the segregation of elemental Ni. While performing the welding process, the content of Ni in the grain boundaries of the HAZ on the SUS 316 side increased, which consequently lowered the melting point of the grain boundaries. When conducting high-temperature tensile testing, it melted prior to the matrix under the influence of interface energy; this induces the HAZ cracks, and under the influence of tensile stress, concentrations of stress are formed, and fractures eventually occur. Then, the liquid film comes in contact with air and shrinks to form a spherical shape under surface tension.

4. Conclusions

In the present work, Inconel 718/SUS 316 butt joints were achieved by the cold metal transfer welding process employing ERNiFeCr-2 welding wire as filler metal. Through the analysis, the main conclusions obtained were as follows:

1. When the welding current is 160 A, no macroscopic defects can be observed in the weld and the wettability and spreadability of the weld metal are better in contrast to other weldments.

2. The weld microstructure consists of a dendritic structure. In inter-dendritic regions of the weld metal, some micro-segregation exists. The primary dendrites near the fusion line of base metal SUS 316 can be clearly observed. No obvious atomic diffusion is identified across the joint interface.

3. The weld exhibits a higher hardness than that of the SUS 316 base metal. The high-temperature tensile strength of the joint (424 MPa) can reach 87.06% of the SUS 316 base metal, 67.19% of the Inconel 718 base metal and 83.0% of the joint tensile strength tested at room temperature. The elongation of the joint (21.9%) is slightly higher than that of the Inconel 718 base metal (21.2%) and lower than that of the SUS 316 base metal (22.3%) and the joint (27.5%) tested at room temperature. The tensile fracture mechanism of the joints is ductile failure.

4. Spherical elemental Ni segregation morphology is observed on the high-temperature fracture surface. During the welding process, the grain boundaries in the HAZ of the SUS 316 had higher Ni, resulting in a
lower the melting point of the grain boundaries in this region. When tested at 750 °C, the grain boundaries melted prior to the matrix under the influence of interface energy; this induces the HAZ cracks, and lowers the mechanical properties of the joint at high temperature.

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