REPAIR OF HOT CRACKING IN PIPING: EXPERIMENTAL INVESTIGATION INTO GTAW CLAD WELDING OF AISI 304L STAINLESS STEEL USING INCONEL 52M

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This study investigated the cladding of AISI 304L stainless steel (SS) with four layers of Inconel 52M using gas tungsten arc welding (GTAW). Our results revealed an increase in the hardness of the SS substrate at the interface with the first cladding layer. However, the hardness of the Inconel 52M region within the weld decreased with the number of cladding layers, due to the effects of dilution. This had the effect of altering the thermal efficiency during the solidification process, resulting in the formation of various solidification structures. In addition, the cladding process showed to increase the susceptibility of the 304L SS substrate to hot cracking due to an increase in the precipitation of \( \delta \)-ferrite phase at the interface between the substrate and the first Inconel 52M cladding layer, the effects of which appear to increased susceptibility.

I. INTRODUCTION

Despite environmental concerns in some quarters, nuclear power is widely favored as a sustainable energy source with minimum emissions. As a result, nuclear power reactors are now deployed in many countries around the world. However, many of these power plants have been in operation for more than 30 years and various flaws have been detected in their piping systems, leading to concern regarding their long-term safety. The piping systems in nuclear power plants are generally fabricated from austenitic stainless steel (SS) alloys due to their favorable workability and mechanical properties and their superior resistance to rust, corrosion, oxidation and heat [1-3]. However, many of the pipe joints comprise dissimilar sections, and are therefore susceptible to deterioration in the welds (e.g., cracking, distortion and porosity) following long-term exposure to high-temperature, high-pressure and radioactive environments [4-7]. In recent years, cracks have been discovered in dissimilar welds connecting the reactor nozzle to the Class 1 piping system in several plants [8]. Moreover, many instances of failure associated with dissimilar alloy weldments have been reported [9-11]. As a result, the structural integrity evaluation and repair of the piping systems in the reactor vessels of nuclear power plants have attracted significant interest in recent years. To suppress crack growth and increase the thickness and corrosion resistance of the welds used in the piping systems of nuclear power plants, the welds are generally clad with a dissimilar metal using a gas tungsten arc welding (GTAW) process [12, 13]. However, when using Inconel 52M to clad AISI 316L SS piping, hot cracking readily occurs in the surface layer of the cladding weld due to changes in the materials, environmental conditions, and operating parameters of the cladding process [14, 15]. The defects associated with hot cracking can be attributed to the composition of alloys, and are largely due to thermal contractions resulting from differences in the regional heating and cooling rates of the dissimilar metals during the welding process [16, 17]. It has been suggested that hot cracking may be prompted by a diffusion reaction at the interface between the two dissimilar metals, which produces a discontinuity in the strength distribution across the interface and leads to a weakening of the joint after welding as a result [16, 18]. Consequently, suppressing the hot cracking effect during the clad welding of SS alloy components is essential in ensuring the long-term safety of nuclear power plants. Accordingly, the present study performs an experimental investigation into the hardness and hot cracking propensity of AISI 304L SS plates clad with four layers of Inconel 52M by means of a GTAW process. The hardness and hot cracking properties of the clad specimens are interpreted in terms of the microstructural and compositional changes prompted by the heating and cooling cycles of the welding process. We measured the total length of hot cracks in the heat-affected zone (HAZ) and used the total crack length (TCL) as an index for evaluating susceptibility to hot cracking. It is anticipated that the results presented in this study will serve as a useful reference for the cladding repair of piping systems in nuclear power plants and other industries.

II. MATERIALS AND METHODS

1. Experimental Materials

The welding trials were performed using AISI 304L SS plates. According to the manufacturer’s specification, the chemical composition of the AISI 304L SS plates was as follows:


3. Microstructural Characterization Observations

The weld cladding shape was examined using an optical microscope (Olympus SZ51). In addition, the microstructures of the various cladding zones were analyzed using an optical microscope (OM, Olympus BH-2) and a field emission scanning electron microscope (FE-SEM, JSM-6390 LA). Finally, the chemical composition and element distribution within the various specimens were examined via energy dispersive X-ray spectrometry (EDS) were sectioned and embedded in resin at room temperature and were then mechanically polished and chemically etched in an acetic picric solution. The specimen for metallographic analysis was sectioned into planes perpendicular to the direction of welding and parallel to the weld crown. The sectioned specimens were then prepared in accordance with standard metallographic practice and etched with micrograph reagents in order to enable the separate identification of the base metal (BM), fusion zone (FZ), and heat affected zone (HAZ) regions of the weld.

4. Mechanical Properties

i. Hardness test

Hardness measurements were performed on transverse cross-sections of the welded specimens using a Vickers hardness tester (Matsuzawa, Seiki Mv-1). The hardness measurements were obtained at intervals of 0.2 mm under a load of 300 gf for 15 s, as shown in Fig. 1. (Note that the reported hardness values represent the average measurement obtained over five separate tests).

ii. Hot cracking susceptibility test

The hot cracking susceptibility of the clad specimens was investigated using a multiple spot Varestraint tester. For each specimen, tests were performed with augmentation strains of 3%, 4% and 5%, respectively, in order to imitate the residual stress in the fusion zone following each pass of the cladding process, as shown in Fig. 1. Following completion of the Varestraint test, an optical microscope (Olympus SZ51) was used to observe the hot cracking phenomenon in each specimen. In addition, the total crack length (TCL) in the HAZ was measured using commercial image-processing software (Photoshop CS4,
III. RESULTS AND DISCUSSIONS

1. Hardness Analysis

Fig. 2 illustrates variations in the measured hardness values across the AISI 304L SS substrate and the four Inconel 52M cladding layers. The highest hardness values (approximately 210 ± 5 Hv) were obtained at the interface between the substrate and first cladding layer. Hardness values reduced with an increase in the number of cladding layers, falling to a roughly consistent value of 180 ± 5 Hv in the third and fourth layers.

2. Microstructure Analysis

i. Microstructure characterization

Fig. 3 presents the microstructure of the AISI 304L SS substrate and four layers of Inconel 52M cladding. The microstructure of the AISI 304L SS substrate remains austenitic following the cladding process, indicating that full polymorphic transformation took place during cooling. However, due to the presence of 0.5~1% Nb in the Inconel 52M, the substrate stabilized at a higher temperature. It should be noted that Nb has a strong tendency to raise the bulk solidification temperature range. In addition, the microstructure of the fourth cladding layer consists of cellular grains rather than dendrites, due to the fact that Nb reduces the constitutional undercooling effect \[16, 20, 21\]. In other words, additional layers of Inconel 52M cladding may increase the constitutional cooling rate. The increasing constitutional cooling rate associated with the addition of Nb may result in the gradual refinement of dendritic and cellular grains.

The images presented in Figs. 3 (a) and (d) show that the solidified morphology changed with an increase in the number of cladding layers due to the dilution effect associated with the overlapping of cladding layers. This finding is consistent with those of previous studies \[22\], in which it was reported that the thermal conductivity of alloys varies as a function of Cr, Nb content, with a subsequent effect on constitutional supercooling that either promotes or inhibits the formation of grains, leading to a change in the solidification morphology. As shown in Fig. 3 (d), the microstructure of the fourth cladding layer contains refined cellular grains. Thus, it can be inferred that no dilution from the AISI 304L SS substrate occurred and the constitutional supercooling effect remained unaffected by the substrate. In contrast, in the second and third cladding layers, the microstructure consists predominantly of columnar grains (Figs. 3 (b) and (c)) due to the fact that solidification behavior in these layers is affected by the composition of the AISI 304L SS substrate and Inconel 52. Finally, the microstructure of the first cladding layer contains a mixture of dendritic and cellular grains (Fig. 3 (a)) due to the dilution effect of the AISI 304L SS substrate on the Inconel 52M cladding layer, particularly at the interface where the microstructure comprises bulk grains. Generally, when
welding dissimilar metals, the region near the fusion boundary differs significantly in composition from the bulk weld metal, and therefore has a different microstructure and properties. This region is variously referred to as the unmixed zone (UZ), i.e., the filler-metal-depleted zone, the partially-mixed zone, the intermediate mixed zone, and the hard zone. The UZ assumes the form of a laminar layer, in which a small fraction of the BM completely melts and re-solidifies without undergoing dilution with the filler metal [23, 24]. Moreover, the UZ exists distinctly on both sides of the interface (i.e., the AISI 304L SS substrate and Inconel 52M cladding layer in the present study). In other words, the direction of growth in the interfacial region between the substrate and first cladding layer was perpendicular to the FL due to the large temperature differential near the fusion line. Between the application of the first and fourth cladding layers, these grains became increasingly coarse. The density in the interfacial region near the FL of the bottom was greater than that observed in the fourth cladding layer, due to the higher rates of cooling and constitutional cooling. The cooling rate depends on the heat input, preheat temperature, and plate thickness. When the weld beads of the first layer were deposited, the preheat temperature was at its lowest. The cooling rate in the first layer was higher than that of the other layers. Constitutional cooling depends on differences in composition. The root had the highest constitutional cooling rate because the greatest difference in composition occurred in the area near the FL.

In addition, a small portion of the BM just within the HAZ exceeded the solidus temperature but never quite reached the liquidus temperature. This region is known as the partially melted zone and is noted for the formation of bulk grains. It can be seen in Fig. 3 (a) that in the present specimens, the partially melted zone appears wider on the Inconel 52M side of the interface than on the substrate side. The tendency of the microstructure to form a mixture of dendritic and cellular grains melting into the Inconel 52M The use of heterogeneous materials in the welding process in which Inconel 52 was overlay welded produced dilution effects. This affected the solidification behavior, thereby altering the solidification morphology, precipitates, and mechanical properties. These, in turn, influenced weld formability, as shown in Fig. 4. The dilution ratio was obtained using the formula (1):

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

In the components affected by overlay welding of the first cladding layer to the AISI 304L SS, the dilution ratios of the bottom and bead crown welds were 62.3% and 4.5%, respectively. On the basis of these results, we predict that the dilution effect should gradually decrease with an increase in the number of cladding layers. We observed that the interface region contained both small and large δ-ferrite precipitates. In addition, during cladding of the first layer, the microstructure took the form of a mixture of dendritic and refined cellular grains.

As described above, the cladding process results in a dilution between the AISI 304L SS substrate and the Inconel 52M cladding layers. Fig. 5 presents the SEM/EDX analysis results of chemical composition gradients within AISI 304L SS substrate and four Inconel 52M cladding layers regarding the distribution of the Fe, Cr and Ni content in the substrate and cladding layers. It is clear that for each element, the greatest change in the composition gradient occurred at the interface between the substrate and the first cladding layer. (Note the distribution of major elements (Ni, Cr, Fe) across the BM (i.e., AISI 304L SS substrate) and FZ (i.e., cladding regions). The composition profiles in the two dilution regions were similar. However, variations in the composition of the high dilution
Fig. 6 SEM micrographs of AISI 304L SS substrate and Inconel 52M cladding layers: (a) Interface between the substrate and first cladding layer; (b) interface between the first and second cladding layers; (c) interface between the second and third cladding layers; and (d) the fourth cladding layer.

It should be noted that the Ni content at the interface decreased, whereas Fe and Cr content increased. In contrast, the composition gradients in the second and third cladding layers varied slightly. The composition gradient for each element in the third cladding layer was close to zero, indicating a composition nearly identical to that of the original Inconel 52M. Changes in the solidified morphology of the cladding layers can be attributed to the dilution effect and may have a direct impact on the hardness of the cladding weld. For example, as shown in Fig. 2, the greatest change in hardness was observed at the interface between the AISI 304L SS substrate and the first cladding layer; i.e., the region of the cladding weld characterized by the greatest change in composition (see Fig. 5). Similarly, the gradual reduction and stabilization in the hardness value following an increase in the number of cladding layers reflects a gradual restoration of the composition in cladding layer to that of the original Inconel 52M filler metal (185 ± 4 H_{V}).

ii. Interfacial microstructure characterization

Fig. 6 presents SEM images of the AISI 304L SS substrate and four Inconel 52M cladding layers. It can be seen that a very large UZ in the form of a laminar layer formed between the substrate and the first cladding layer along the fusion line (FL) (see Figs. 3 (a) and 5 (a)). When the melting range of the first cladding layer (Inconel 52M) is similar to or higher than that of the AISI 304L SS substrate, only a small fraction of the AISI 304L SS melted. As a result, no dilution occurred in the re-solidification stage, such that a UZ formed between the two regions, this observation is consistent with that of previous studies [19, 24, 25]. However, considerable grain growth can be observed in the HAZ of the AISI 304L SS substrate due to variations in the welding temperature. The interfacial region between the substrate and first cladding layer contains a small amount of δ-ferrite precipitates (see Fig. 6 (a)). In other words, in the earliest stages of solidifica-
ination, the phase in the interfacial region between the substrate and first cladding layer has a structure that includes δ-ferrite precipitate, which grows in size and decreases in number, resulting in non-uniform distribution. This phenomenon is thought to be directly related to the effects of hardening. The drop in hardness near the FL can be attributed to the dilution effect resulting in the formation of less precipitate near the FL. This observation is consistent with that of previous studies [26-28]. Fig. 6 (b) and (c) illustrate the interfacial regions between the first and second cladding layers and second and third cladding layers, respectively. A UZ similar to that observed at the interface between the AISI 304L SS substrate and first cladding layer formed in both interfacial regions; however, no UZ is observed in the second or third cladding layers, which show signs of considerable dilution. This can be attributed to similarities in the melting temperature and chemical compositions of the second and third cladding layers. In addition, columnar grain growth was observed near the FL. Fig. 6 (d) presents the interfacial region between the third cladding and fourth cladding layers, which appears to be similar to that between the second and third cladding layers.

3. Susceptibility to Hot Cracking Analysis

In this experiment, we divided the HAZ of the weld metal (cladding layers), and that of the AISI 304L on the basis of the HAZ position, as shown in Fig. 7. We then tested them under various amounts of augmentation strains and a various number of thermal cycles. The TCL of the stainless steel in the cladding layers as a function of augmentation strain and heating duration. Test results show that hot cracking of the weld metal (WM) HAZ occurred mainly at the AISI 304L HAZ; however, the TCL in this region was significantly short. At strains of 3% and 4%, the TCL decreased with an increase in welding time (4 and 5 sec, respectively). In contrast, at a welding time of 6 sec, the TCL showed a slight increase. Under strain of 5%, the TCL increased with welding time. The fact that welding time did not affect the susceptibility of the cladding layer to hot cracking under strain of 3% and 4% clearly indicates a relationship between TCL and welding time.

It has been reported that hot cracking is caused by the liquidation of grain boundaries due to the segregation of elements [29]. Microstructural observations in Fig. 6 reveal significant element segregation that occurred during the weld cladding process. Nonetheless, a small amount of δ-ferrite formed at the interface between the AISI 304L SS substrate and the first layer of Inconel 52M cladding [25, 30]. It appears that δ-ferrite precipitate inhibited the growth of cracks, thereby reducing susceptibility to hot cracking [18, 22]. However, much of the small δ-ferrite formed at the interface between the AISI 304L and the Inconel 52M cladding layer. This helped reduce TCL growth, which would otherwise have occurred as a result of hot cracking. Previous studies have reported that the formation of 3-5% δ-ferrite in a stainless steel system helps reduce its susceptibility to hot cracking [21, 22]. Our results were, therefore, consistent with the findings reported in such studies. Interestingly, when the welding time was 6 sec, the TCL tended to increase under strains of 3% and 4%. One possible reason would be the melting of grain boundaries (the last to solidify) or a change in the δ-ferrite resulting in phase transformation, both of which could lead to an increase in the TCL. Under 5% strain, the TCL increased significantly with an increase in welding time. One reason to explain the acceleration in TCL growth is that the stainless steel in the Inconel 52M cladding layer was unable to withstand high strain or lengthy exposure to a high-temperature environment. Another possibility is that, as the strain increased, the solidification process following the cladding welding required a greater quantity of molten metal to be added to the grain boundaries. If the supply were insufficient, cracking could easily occur at the interface. These experimental results are comparable to those found in the literature [2, 12]. They demonstrate that the application of Inconel 52M cladding layer on the AISI 304L stainless steel substrate can significantly reduce susceptibility to hot cracking. The fracture surface was produced on the hot cracking susceptibility test specimens, as shown in Fig. 8. The fractures were observed at low mag-

![Fig. 7](image_url)
nification to identify the fracture morphology of several different regions, and at high magnification to identify the fine-scale features of the fracture surfaces. The acquired images offer interesting results, as there are definite differences in the fractured surfaces under various augmentation strains (3%, 4%, and 5%), which allowed us to evaluate the efficacy of various welding times (4, 5, and 6 sec). The specimens presented similar fracture zones characterized by equiaxed ductile fractures. We therefore surmise that crack formation is due to the segregation or dendritic arm spacing that occurs in the grain boundary when the molten austenite stainless steel solidifies, and a film with a lower freezing point and lower strength is produced.

Austenite stainless steels possess higher coefficients of thermal expansion, lower heat conductivity coefficients, and superior high temperature strength. For this reason, extreme contraction stress, which increases as the weldment cools, can form at very high temperatures and is greater than the film can bear, results in cracks along the gaps between grains and dendritic arms.

Fig. 9 illustrates the influence of multi-thermal cycles on the susceptibility of HAZ to hot cracking in cases of augmentation strain of 1%, 3% and 5%. Cracks in specimens subjected to three thermal cycles (3C) are longer than those in specimens subjected to one thermal cycle (1C). In all of the specimens, the length of hot cracks in the AISI 304L SS HAZ was significantly longer than those in the weld metal HAZ. It is believed that this is the result of the extended heating, which can result in melting at grain boundaries (the last area to solidify) or a change in the interface between the AISI 304L SS substrate and δ-ferrite in the first cladding layer, resulting in phase transformation [19]. In specimens subjected to augmentation strain of 5%, the TCL also increased significantly with an increase in the number of thermal cycles. This may be due either to the inability of the Inconel 52M cladding layer to withstand high strain or the fact that extended exposure to high temperatures accelerated the growth of cracks. This can be attributed to the formation of a lot of δ-ferrite precipitate at the interface between the AISI 304L SS substrate and the Inconel 52M layer following repeated thermal cycles. Further, the δ-ferrite precipitate in the welds exhibited secondary-phase strengthening with a concomitant decrease in ductility and toughness [31]. Alternatively, it may be that under increased strain, the solidification process requires additional molten metal at the grain boundaries, a lack of which resulted in cracking at the interface. Most of the experiment results in Fig. 9 are consistent with those of previous studies [22], showing that Inconel 52M cladding is unable to alleviate the susceptibility of AISI 304L SS to hot cracking. This conclusion is based on these results, and reasonable inferences concerning the effects of cladding parameters on the dilution of weld specimens. During the cladding of AISI 304L SS specimens, the molten components are transferred individually to the liquid alloy and mixed thoroughly through thermal agitation. Fe, Ni, and Cr become dominant in the solidification process because their atoms are solutes in the alloy system. Following a reduction in the temperature of the specimen, the atoms lose their solubility and precipitate in the preferred crystalline structure.
such that the concentration and flux of the solutes cannot be predicted. The radius of the atoms, the crystalline structure, and the melting point of the elements all influence the mean free path of the diffusing atoms, which when increased can lead to the formation of small δ-ferrite precipitates at the interface between the AISI 304L SS substrate and the first layer of Inconel 52M cladding.

The fracture surface was produced on the hot cracking susceptibility test specimens, as shown in Fig. 10. The fractures were observed at low magnification to identify the fracture morphology of several different regions, and at high magnification to identify the fine-scale features of the fracture surfaces. The acquired images offer interesting results, as there are definite differences in the fractured surfaces under various augmentation strains, which allowed us to evaluate the efficacy of multi-thermal cycles. The specimens presented similar AISI 304 L SS zone (i.e., BM), weld metal zone (i.e., Inconel 52M), and interface characterized by dendrite ductile fractures. We therefore surmise that crack formation is due to the segregation or dendritic arm spacing that occurs in the grain boundary when the molten austenite stainless steel solidifies and a film with a lower freezing point and lower strength is produced. Austenite stainless steels possess higher coefficients of thermal expansion, lower heat conductivity coefficients, and superior high temperature strength. For this reason, extreme contraction stress, which increases as the weldment cools, can form at very high temperatures and is greater than the film can bear, results in cracks along the gaps between dendritic arms.

In summation, the objective of this study was to produce effective welds between Inconel 52 and AISI 304L SS. The microstructure and mechanical properties were examined in detail in order to interpret the relationship between the structure of the weldment and resulting physical and mechanical properties. Our findings provide a valuable resource for industries employing this form of bimetallic combination.

IV. CONCLUSIONS

This study investigated the hardness characteristics and susceptibility of AISI 304L SS to hot cracking after being clad with four layers of Inconel 52M using GTAW. Our experiment results support the following conclusions:

1. In AISI 304L SS clad with four layers of Inconel 52M, the highest hardness values (210 HV) were obtained at the interface between the substrate and first cladding layer. The hardness within the cladding region of the weld decreased with an increase in the number of cladding layers, due to changes in microstructure. Observation of the microstructure revealed that an increase in the number of cladding layers led to the formation of various solidified morphologies. Specifically, overlap between the cladding layers prompted an alteration in the composition of the alloy (i.e., through dilution effect) during the cladding process. This resulted in the formation of regions with different thermal efficiencies, which led to the formation of different solidification structures.

2. Hot cracking susceptibility tests demonstrated that AISI 304L SS substrate clad with a Inconel 52M cladding significantly reduced hot cracking because δ-ferrite precipitates formed at the interface between the AISI 304L SS substrate and Inconel 52M reduced susceptibility to hot cracking. In addition, considering to the hot cracking susceptibility of the specimens subjected to multi-thermal cycles, it is shown that the cladding process unable resisted the susceptibility of AISI 304L SS to hot cracking following multi-thermal cycling increased significantly after being clad with Inconel 52M. This can be attributed to an increase in the formation of δ-ferrite at the interface between the substrate and first cladding layer.

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