Influence of Aging Temperature on Metallurgy, Impact Toughness and Pitting Behavior of Flux-Cored Arc Welded 2205 Duplex Stainless Steel Joint

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This study systematically investigated the effect of aging temperature on microstructure, impact toughness and pitting behavior of flux-cored arc welded 2205 duplex stainless steel joint. 600°C–900°C aging treatments were carried out with 1-hour insulation. The results showed that the ferrite (α) underwent serial metallurgical transformations with increasing the aging temperature. The formation of the sigma (σ) phase was detected in the weld zone (WZ) under all aging temperatures due to eutectoid reaction of the α phase, while the peak content of the σ phase was reached at 800°C aging. A similar content variation of the σ phase was detected in the heat-affected zone (HAZ). Secondary austenite (γ_2) and chi (χ) phases were only observed in the WZ and HAZ below 800°C aging. The impact toughness and the pitting resistance of the welded joints were strongly related to the σ phase content, that the 800°C aging specimen possessed the lowest impact toughness and the worst corrosion resistance due to the maximum content of the σ content in both the WZ and the HAZ. [doi:10.2320/matertrans.MT-M2020331]

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1. Introduction

Duplex stainless steels (DSSs), which consist of a nearly equal proportion of ferrite (α) and austenite (γ) phases, are usually used in chemical, petrochemical, nuclear, food industries, etc. due to their superior performance of higher strength, toughness and high corrosion resistance than conventional austenitic stainless steels.^{1–3)} Compared with the austenitic stainless steels, the hot cracking tendency during welding of DSSs is weaker.⁴⁾ In comparison to ferritic stainless steel, the DSSs possess lower ductile-brittle transition temperature, better resistance to intercrystalline corrosion and pitting corrosion. In addition, the DSSs retain high thermal conductivity and small linear expansion coefficient, benefitting to excellent weldability.^{3,5)}

For DSSs, the role of various alloying elements such as chromium (Cr), nickel (Ni) and nitrogen (N) plays a critical role in their weldability via adjusting the α/γ phase balance, due to high alloying contributed precipitation of intermetallic phases like carbides and nitrides of Cr (Cr_{23}C_6 and Cr_2N), chi (χ) and sigma (σ).^{6–11)} It was reported that the σ phase occurred in binary Fe–Cr system, while the χ phase could be found only in ternary Fe–Cr–Mo and quaternary Fe–Cr–Ni–Mo and Fe–Cr–Ni–Ti systems.^{7,12)} These precipitates affect the properties of DSSs in varying degrees. Thereinto, the σ phase is the largest amount of precipitation and leads to the greatest harm to the material properties such as ductility, toughness, and corrosion resistance.^{9,13,14)}

Welded joints play a crucial role during the application of DSSs, which can determine the whole structural performance to a certain extent. However, improper high-temperature exposures or heat treatment procedures of the welds can occur during their unanticipated service processes, which will produce undesirable precipitations and result in the decrease of the joint performance.¹⁵⁾ Therefore, the research of the influence of postwelding aging treatment on the metallurgy is considerable to comprehend the evolution of joint properties.

However, most of the current studies paid close attention to the effects of aging on the intermetallic precipitation, while systematic studies on the evolution of comprehensive properties such as impact toughness and corrosion resistance are relatively few in DSSs joints, especially in flux-cored arc welded joint.^{15,16)} Therefore, the present work aims to systematically investigate the influence of aging temperature on the evolution of metallurgy, impact toughness and pitting behavior of flux-cored arc welded 2205 DSS joint.

2. Experimental Procedure

2205 duplex stainless steel with a thickness of 12 mm was used as the base material (BM) and ER2209 flux-cored wire with 1.2 mm diameter was adopted during flux-cored arc welding (FCAW). The main chemical composition of the base metal and the welding wire is shown in Table 1. It can be seen that the Ni content of the welding wire is significantly higher than that of the base material. Before welding, a single V groove was machined with an angle of 60° and both sides of the groove were polished by a stainless steel grinding wheel. As sketched in Fig. 1, the distance between the two plates for welding was 2 mm, and the length of the root face was 1 mm. Acetone was used to clean oxidation film, grease and impurities on the plate surfaces.

A digital inverter welder (YD-350FR) was used to supply welding power and a direct current (DC) reverse connection method was adopted during welding. Detailed welding parameters are shown in Table 2. CO_2 with a purity of 99.99% was employed as the shielding gas to protect the

Table 1 Chemical composition of the 2205 DSS and ER 2209 (mass%).

Materials	C	Si	Mn	Cr	Ni	Mo	N	Fe
2205	0.015	0.37	1.39	22.35	5.66	3.22	0.16	Bal.
ER 2209	0.021	0.63	1.07	22.56	7.98	4.33	0.15	Bal.

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Fig. 1 Schematic of the weld configuration.

Table 2 Detailed welding parameters during the FCAW.

Filler metal	Welding current	Welding voltage	Welding speed
	(A)	(V)	(cm min ⁻¹)
ER2209	180	26	23

weld. During the formation of the joint, multi-layer and multi-pass process was taken on and the interpass temperature was not more than 150°C. After welding, the welded joints were subjected to aging treatment under 600°C, 700°C, 800°C and 900°C respectively. After 1-hour heat preservation, the samples were quenched in water for cooling to the room temperature.

The microstructures of the welded joints were observed via a scanning electron microscope (SEM) equipped with energy dispersive spectroscopy (EDS). X-ray diffraction (XRD) was used to acquire the joint phase variation characteristics during the aging treatment. Subsize impact specimen with V-notch and specimen size of 7.5 mm × 10 mm × 55 mm following ASTM E23-16b standard¹⁷⁾ was used for testing the impact properties of the WZ metal under -40°C. Pitting corrosion test was conducted according to ASTM G48-11(2015)

standard.¹⁸⁾ The joint samples prepared by metallographic polishing were etched with 6% ferric chloride solution. The spot corrosion rate was calculated by the weight loss per unit area and per unit time of the sample before and after etching in units of g/(m²h).

3. Results and Discussion

3.1 Microstructure evolution

Figure 2 shows the microstructural characteristics of the as-welded joint. The BM consisted of an $\alpha + \gamma$ duplex structure. The α phase was distributed in the γ phase matrix with band or island shape (Fig. 2(a)). Similarly, the weld zone (WZ) also showed an $\alpha + \gamma$ duplex structure, while its phase size was slightly coarser than that of the BM (Fig. 2(b)). As shown in Fig. 2(c), the heat-affected zone (HAZ) also showed α and γ phases, but its duplex structure was larger than that of the WZ. Besides, a few carbides were observed in the α phase. These two phenomena were due to the high-temperature exposure induced grain growth and precipitation of the second phase during the FCAW.

Figure 3 shows the microstructural characteristics of the welded joint aged at 600°C. As can be seen in Fig. 3(a), the WZ consisted of a duplex structure of $\alpha + \gamma$ without obvious precipitation of the second phase such as σ and χ , which was similar to the as-welded joint. However, two more precipitated phases of σ and χ phases were found in the HAZ except for the $\alpha + \gamma$ matrix phases (Fig. 3(b)). In the magnified views of the precipitated phases, the σ phase and χ phase were observed at the boundary of ferrite and austenite, and a few formed in the interior of the α phase (Fig. 3(c), (d)). It is generally considered that the σ phase was formed during the eutectoid reaction of $\alpha \rightarrow \sigma + \gamma_2$, where γ_2 indicates secondary austenite.^{14,19)}

Fig. 2 SEM microstructures of the as-welded joint.

Fig. 3 SEM microstructures of the welded joint aged at 600°C.

In terms of the morphologies of the precipitated phases in the SEM image, the strip-like χ phase was whiter than the sheet-like σ phase but showed a smaller size surrounded by the σ phase.⁷⁾ The χ phase is a kind of metastable phase, which can further transform into the σ phase with increasing the aging time. That is to say that the χ phase forms before the precipitation of the σ phase and disappears once the σ phase starts to precipitate, as also declared by Escriba *et al.*⁷⁾ and Okafor *et al.*¹²⁾ In addition, Llorca-Isern *et al.*²⁰⁾ pointed out that there was a great difference in alloy content for the σ phase and the χ phase. The σ phase possesses higher content of Cr element than χ phase even than the matrix phases of the α and γ , while the χ phase owns a higher Mo element than the other phases.

Figure 4(a) shows microstructures of the WZ after 700°C aging treatment. It can be seen that the σ phase ratio was significantly higher than the WZ and HAZ microstructures after the 600°C aging treatment. Except for the σ phase, the γ_2 phase was also found in the α phase, which was another product of the eutectoid reaction.

In the HAZ, σ and γ_2 phases were also found, while most of the σ phases were located at the phase boundaries of the α and γ , and the γ_2 phase was few and located in the α phase with rod shape (Fig. 4(b), (c)). Due to the diffusion coefficient of alloy element across the phase boundary was higher than that at the grain boundary, the Cr and Mo were first enriched near the phase boundary where the σ phase preferred to nucleate and grow up.²¹⁾

Besides, carbides were also formed inside the α phase with dark particles (Fig. 4(c)), which were generally considered as $M_{23}C_6$ type.¹⁹⁾ The precipitation of the carbides would cause Cr deficiency areas near the grain boundaries, which could promote the migration of the original phase boundaries and contribute to the formation of the

intermetallic phase. No significant χ phase was observed in the HAZ, which was due to the rapid diffusion rate of alloy elements of Cr and Mo with the increase of temperature, while the χ phase was basically transformed into the σ phase as a metastable phase.

After aging treatment of 800°C, the microstructures of the WZ showed an increasing amount of σ phase compared with that of the 700°C treatment, and the banded σ phases were more widely distributed in the γ matrix with smaller size and smoother phase boundaries (Fig. 5(a)). After aging treatment at 800°C, the α phase disappeared in the WZ, and the phase composition was almost composed of $\sigma + \gamma$. For verifying the formation of the σ phase, element distribution was measured via EDS shown that the concerned phase possessed high content of Cr and Mo (Fig. 6), which was the mean feature of the σ phase.²⁰⁾ As shown in Fig. 5(b), large block σ phases formed in the HAZ and were coarser than that of the 700°C HAZ. In addition, no carbide particle was found inside the α phase in the HAZ, indicating that the carbide was hard to precipitate at 800°C.

After 900°C aging treatment, the σ phase in the WZ gathered around the γ phase while its proportion was significantly decreased (Fig. 7(a)). The size of the γ phase increased during the high-temperature duration. Similar to the 800°C treated joint, the block σ phase was found in the HAZ, and no carbide particle and was observed in the α phase. However, the content of the block σ phase was lower than that of the 800°C HAZ (Fig. 7(b)).

Figure 8 shows the XRD results of the WZs after 600°C, 700°C, 800°C, and 900°C aging treatment, respectively. For 700°C, 800°C and 900°C parameters, the intensity of the α phase was very low and the phases were almost composed of γ and σ , indicating that the significant transformation of the α phase during the eutectoid reaction. Figure 9 shows the

Fig. 4 SEM images of the welded joints aged at 700°C.

Fig. 5 SEM images of the welded joints aged at 800°C.

XRD results of the HAZs. It was clear that almost no σ phase was observed in the HAZ after 600°C treatment, which was not consistent with the SEM results of Fig. 3. This was due to the extremely low content of the σ phase in the sample. Under the 700°C and 900°C parameters, the diffraction peaks of the σ phase became pronounced, and the medium aging temperature of 800°C promoted the highest peak intensity of the σ phase, which was consistent with the SEM results.

Because the content was very low, the quantitative analysis of the σ phase was hard to perform in this study. Thus, we calculated the σ phase variation for the WZ metal by using Thermo-Calc software. As shown in Fig. 10, the σ phase content of the WZ increased with increasing aging temperature and reached the maximum of 25.8% over 700°C (735°C). It was clear that the aging temperature of the peak σ phase content was not consistent with the experimental one (800°C), which could be related to the calculation deviation and was acceptable in common consideration. In addition, during the eutectoid reaction of $\alpha \rightarrow \sigma + \gamma_2$, Cr element

diffused quickly in α phase and gathered at grain boundaries and α/γ phase boundaries, resulting in the formation of the σ phase. With the temperature increasing, dynamic equilibrium of Cr element was reached between α phase interior and α/γ phase boundaries, which inhibited the eutectoid reaction and precipitation of σ phase. Hence, the σ phase was significantly decreased at 900°C aging temperature.

3.2 Weld impact property

The impact testing results of the aged WZ metals indicated a valley-distribution of the impact work as the temperature increase. The lowest impact work of 9 J was obtained for the 800°C sample, while 33 J, 24 J and 21 J were measured for the 600°C, 700°C and 900°C samples, respectively. The valley-distributed impact work was due to the dramatically increased σ phase content and the limitative precipitation of the σ phase under 600°C, 700°C, and 900°C. It is well known that the σ phase was a kind of hard and brittle carbide which can strongly decrease materials impact toughness, which

Fig. 6 SEM image combined with EDS line scan results showing the distribution of Cr, Mo, Ni and Fe in the WZ aged at 800°C.

Fig. 7 SEM microstructures of welded joints aged at 900°C.

leads to premature crack initiation during impact deformation and causes the slip plane gliding fracture phenomena of quasi cleavage.^{13,22)}

Figure 11 shows the impact fracture morphologies of the WZ metals at different aging temperatures. According to Fig. 11(a), after 600°C aging, the fracture showed a large number of dimples with small and shallow morphology, and few cleavage surfaces were found here. This was corresponding with superior toughness and high impact work. However, the dimple was significantly shallow for WZ metals aging

after 700°C and 900°C, displaying lower impact toughness than the 600°C sample (Fig. 11(b), (d)). Therefore, the 700°C and 900°C HAZs belonged to the mixed fracture mechanism of ductile fracture and brittle fracture. After 800°C aging, the dimple of the HAZ was relatively larger than that of other HAZs. Some dimples were elongated with an opening in a certain direction, which was an embodiment of tearing dimples (Fig. 11(c)). This was a characteristic of transgranular fracture, so the WZ metal showed brittle fracture with low impact work.

Fig. 8 XRD patterns of the WZs obtained at different aging temperatures.

Fig. 9 XRD patterns of the HAZs obtained at different aging temperatures.

Fig. 10 σ phase variation in the WZ metal by using Thermo-Calc software.

3.3 Pitting corrosion resistance

The corrosion rate of the welded joint samples at different aging temperatures was shown in Table 3. After 600°C aging, the corrosion rate of the HAZ and the BM was lower than 2 g/(m²h) while the corrosion rate of the WZ was higher. This indicated that the pitting resistance of the HAZ and the BM was better than that of the WZ, which was also found in other aging specimens. After 700°C aging, the corrosion rate of the WZ and the HAZ suddenly increased due to the increased σ phase content. Large numbers of the σ phases which were full of alloy elements would dilute the surrounding alloying elements such as Cr. Therefore, the pitting resistance in these areas was reduced eventually. After 800°C aging, the corrosion rate of the WZ and the HAZ reached the maximum value, up to 86.063 g/(m²h) and 31.842 g/(m²h) respectively, indicating the maximum content of the σ content in these areas accompanied by the worst corrosion resistance. Figure 12 shows the SEM images of the WZ metal after pitting at different aging temperatures. It was obvious that the 800°C aging specimen owned the worst

Fig. 11 SEM images of the fractural welds treated at different aging temperatures.

Table 3 Average corrosion rate of welded joints under different aging treatment (g/(m²h)).

Aging temperatures	600 °C	700 °C	800 °C	900 °C
WZ	4.625	68.921	86.063	76.417
HAZ	1.684	20.679	31.842	20.975
BM	1.492	16.854	29.7	20.171

corrosion resistance with high density of corrosion pits, which was consistent with the corrosion rate testing results mentioned above.

4. Conclusions

In the present work, a comprehensive study for the metallurgy, impact toughness and pitting behavior of flux-cored arc welded 2205 duplex stainless steel joint was adopted, and several conclusions could be made:

- (1) The aging temperature determined the microstructures of the phase composition of the welds. With the temperature increasing from 600°C, the σ phase content of the welded WZ and HAZ increased gradually and reached the maximum at 800°C due to the eutectoid reaction of the α phase. After 900°C aging, The σ content decreased.
- (2) The aging temperatures could make a significant effect on the impact toughness of 2205 duplex stainless steel welded joints, which was strongly related to the σ phase content. The 800°C aging specimen possessed the lowest impact toughness because of the maximum proportion of the σ phase.
- (3) The aging temperatures also make a significant effect on the pitting resistance of the joints. The 800°C aging specimen owned the maximum weight loss and worst corrosion resistance, which was corresponding with the maximum content of the σ content in the corresponding areas.

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Fig. 12 SEM appearance images of the WZ metal after pitting at different aging temperatures.