Failure analysis of dissimilar steel welded joints in a 3033t/h USC boiler

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Abstract

In this work, macroscopic and microscopic examinations on the failure specimen were conducted on T91/HR3C dissimilar steel welded joints for the high temperature reheater in a 3033t/h USC boiler. The cracking is along the circumferential direction of tube, initiated in the CGHAZ (course grain HAZ) and terminated at the FGHAZ (fine grain HAZ) of the outer wall. The high hardness and unstability microstructure in the HAZ are responsible for the cracking failure. Post welding heat treatment (PWHT) is suggested for improving the ductility and reducing the residual stress for the welded joints. The delta ferrite with lower hardness was observed near the fusion line of T91 side. However, it was not related to the failure.

Keywords: T91 steel; HR3C steel; dissimilar steel welding; delta ferrite; USC boiler.

1. Introduction

In order to improve the efficiency of thermal power generation and protect the environment, China is now developing Ultra-Supercritical (USC) coal-fired power plants with steam temperature up to 600 °C and pressure exceeding 25 MPa. Many new austenitic stainless steels, such as Super304H, HR3C have been widely used for the superheaters and reheaters, which have the severest service environment in USC boilers [1-2]. Commonly, T91 and

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T92 martensitic steels were used as the tubes linking superheaters and reheaters, and inevitably there are dissimilar welds with Super304H or HR3C. Until now, many researches have mainly focused on the properties of new martensitic and austenitic stainless steels [3-5]. As for the martensitic/ austenitic dissimilar steel joints, however, there was little report about it.

In the current work, the T91/HR3C dissimilar steels welded joints for the high temperature reheater were found leakage after the 168h experimental operation and the following water pressurized test in a 3033t/h USC boiler, and then the failure mechanism was analyzed.

2. Experimental

2.1. Materials and welding procedure

Two base materials, namely, T91 and HR3C (50.8 mm OD×3.5 mm thickness) steels are used. The T91/HR3C weld joints were welded by gas tungsten arc welding with pure argon gas (Ar) as the shielding gas and AWS ERNiCr-3 (Corresponding to INCONEL 82) wire with 2.4 mm diameter as the filler material. The arc voltage and the arc current used in welding process were 16-22V and 100-170A, respectively. Local post weld heat treatment is carried out at 760°C/90min.

2.2. Test methods

Macroscopic and microscopic examinations were conducted on the failure welded joints. The specimens were mechanically polished and etched, and then the microstructure was observed by Leica DMI 3000M optical microscopy.

The Vickers hardness distribution of welded joint was measures by the Wolpert 401MVD Vickers micro hardness tester with 100g load.

Charpy impact tests were performed in the room temperature on Charpy V-notch (CVN) bars with a 2.5×10×55-mm size, which were machined from the welding joint, including the HAZ.

3. Results and discussions

3.1. Macroscopic examinations

Fig.1 shows the appearance of the cracking in the leaky welded tube joint. There is no obvious deformation near the cracking, which occurs along the circular direction of the tube in the T91 side. The cracking initiated at the heated-affected zone of the outer wall welded joint tube, propagated along the fusion line, and terminated at the 4 mm away from the fusion line. Hence, it can be inferred that the failure of the T91/HR3C dissimilar steels welded joint occurred under the axial tensile stress.

Fig. 1. Appearance of the crack in the leaky welded tube joint (a) Outer wall; (b) cross section.
3.2. Microstructural observation

Fig. 2 shows the microstructure of T91/HR3C welded joint. The cracking morphology was shown in Fig.2 a-d. The coarse microstructure was found near the initial segment of the cracking shown in Fig.2 a, and can not be deduced whether the cracking propagated along the grain boundaries or inside grain due to the serious oxide. However, it is clear that the cracking propagated along the grain boundaries in the end of the cracking, as shown in Fig.2 b, the main and branch cracking morphology in the end segment, where the grain was very fine, indicated the cracking terminated at the fine-grained heated-affected zone (FGHAZ). The coarse column grain was observed in the weld metal. The T91 and HR3C base metal consist of the fine tempered martensite and not uniform austenite, respectively.

Fig. 2. Cracking morphology and microstructures of the welded joint (a) main cracking (initial segment); (b) branch cracking (the end segment); (c) T91 base metal; (d) HR3C base metal.

Fig. 3. Images of interface in the welded joint (a) interface at WM/T91; (b) interface at WM/HR3C.

The two interfaces including T91 HAZ/ weld metal and HR3C HAZ/ weld metal both bonded well and the fusing line is very clear, seen in Fig.3. The HAZ along T91 side exhibited the typical martensite, however, it should be noted that the bright dense polygonal microstructure formed along the fusing line of T91 side, with the lower hardness of 161-165HV than that of the surrounding martensite (319-383HV). These bright microstructures should be δ-Fe according with their morphology and hardness. No unusual microstructure was observed at the interface of
HR3C HAZ/ weld metal.

3.3. Hardness and impact tests

Fig.4 shows the Vickers hardness distribution of T91/HR3C dissimilar steels welded joint. It can be seen that the maximum hardness in the HAZ exceed 400HV and the high hardness zone (above 350HV) is about 4 mm. The weld metal exhibits the lowest hardness and the T91 base metal has the similar hardness as the HR3C base metal.

The toughness values in the HAZ with 1 mm away from the fusing line of T91 side at room temperature is given in Table 1. It suggests that a satisfactory toughness of the dissimilar welds can be obtained. It should be noted that the experimental values listed in Table 1 is higher than the real toughness in the HAZ due to the contribution of the high ductile weld metal when the cracking shifts to the neighbouring weld metal.

| Sample number | Measured absorbed energy (J) | Equal mean absorbed energy for standard sample(J) |
|---------------|------------------------------|-----------------------------------------------|
| 1             | 20                           |                                               |
| 2             | 21                           | 80                                            |
| 3             | 19                           |                                               |

3.4. Discussion

The failure of T91/HR3C dissimilar steels welded joint may be caused by the cold cracking because its initiated location was at coarse grain heated-affected zone (CGHAZ). However, the welded joint subjected to the lower restrain stress due to the thin wall thickness of the tube and was welded in the workship. Furthermore, the lower diffusion hydrogen content can be obtained in the weld joint as a result of gas tungsten arc welding with pure argon gas (Ar) as the shielding gas and the diffusion of hydrogen was difficult in the austenite weld metal with the Ni-based wire. Thus, the susceptible to the cold cracking of the T91/HR3C dissimilar steels welded joint is very low. The cold cracking often has the mixture fractographs of transgranular and intergranular fracture modes, but the fracture mode of the T91/HR3C dissimilar steels welded joint is intergranular fracture, as shown in Fig.2. The cold cracking should be found by the nondestructive examinations such as UT and RT before the experimental operation because it formed during or after welding. From above discusses, the failure of T91/HR3C dissimilar steels welded joint should not be caused by the cold cracking.

The obvious lath martensite with high hardness was observed near the fracture location, suggested that the post weld heat treatment(PWHT) was not enough, probably with the heating temperature below the required temperature (760°C). PWHT can reduce the crystals distortion of HAZ in T91 side due to the precipitation of carbides, and formed the stable microstructure, then improve the ductile and avoid the brittle fracture under creep. Furthermore, PWHT lowers the residual stress of the HAZ in T91 side and can improve the resistant to the fracture. In the case of
insufficient PWHT, the HAZ in T91 side of the T91/HR3C dissimilar steels welded joint has the high hardness and low ductile, is prone to intergranular fracture under the welding residual and structure stresses. The cracking in HAZ with the high hardness and low ductile easily propagated during the water pressurized test.

It should be noted that dense polygonal δ-Fe formed in the HAZ near the fusion line of T91 side. The presence of a δ-Fe phase is harmful both to creep strength and toughness of the high Cr ferrite steel [6]. The fracture position was 1-4 mm away from the band of the δ-Fe phase, and there is no second cracking in the δ-Fe phase zone, revealed that the delta ferrite with lower hardness was not related to the failure of the T91/HR3C dissimilar steels welded joint. However, the interface with δ-Fe phase was usually recognised as the weak zone of the welded joint, thus, we should pay more attention to the effect of δ-Fe phase on the failure of the T91/HR3C dissimilar steels welded joint in the long term service.

4. Conclusions

(1) Welded joint failures in the heat affected zone (HAZ) of the T91 side. The cracking is along the circumferential direction of tube, initiated in the CGHAZ (course grain HAZ) and terminated at the FGHAZ (fine grain HAZ) of the outer wall.

(2) The high hardness and unstable microstructure in the HAZ are responsible for the cracking failure. Post welding heat treatment (PWHT) is suggested for improving the ductility and reducing the residual stress for the welded joints.

(3) The delta ferrite with lower hardness was observed in the vicinity of fusion line of T91 side; however, it was not related to the failure.

Acknowledgements

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