Strain Effect On Microstructural Properties Of SUS 304L Joint By PAW+GTAW

Juan Wang (✉ jwang@sdu.edu.cn)

Edison Welding Institute

Yajiang Li

Edison Welding Institute

Qi Wang

Edison Welding Institute

Kun Liu

Edison Welding Institute

Hang Li

Edison Welding Institute

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Abstract

304L stainless steel was joined by a combination welding process of plasma arc welding and gas tungsten arc welding (PAW + GTAW). Then pre-strain treatment on 304L welded joint by 9% was carried out using uniaxial static tensile at room temperature. Effect of strain on microstructure evolution in joint was analyzed by field emission scanning electron microscope (FESEM) and on mechanical properties was also studied. The results indicated that the strain rate of 304L joint showed inhomogeneity including 3% in the weld metal and 13% in the heat affected zone (HAZ), which induced martensitic transformation occurring in HAZ. Tensile strength of the joint increased from 700MPa as welded to 804MPa after strain at room temperature and it reached 1700MPa from 1480MPa at low temperature of -196°C. Impact toughness in HAZ was the least among the whole joint, but it was still 94J at -196°C after strain. The fracture surface showed large number of cleavage steps with elongated parabolic dimples.

Introduction

SUS 304L austenitic stainless steel is widely used for pressure vessel at low temperature due to its good comprehensive mechanical properties and excellent corrosion resistance. The permissible stress of 304L steel is usually designed to use by its yield strength considering current safety factor. Therefore, the thickness of pressure vessel is large as some heavy equipment resulting in serious waste of materials owing to its low yield strength and low bending strength ratio, which leads to high cost of manufacturing and transportation. The yield strength of 304L steel can be effectively improved by pre-strain strengthening treatment to make it partially plastic deformed on the premise that the original mechanical properties of 304L steel are not greatly affected[1–3]. The thickness of vessels designed by new yield strength after strain strengthening can be reduced by 20%-30%, which is favorable to save materials, reduce costs and energy consumption in transportation[4].

Welding is the key technology in the manufacture of austenitic stainless steel pressure vessel. Submerged arc welding (SAW) and gas tungsten arc weld (GTAW) are commonly used in medium and heavy plate stainless steel vessels [5, 6]. The heat affected zone (HAZ) in joint is wide due to the larger welding heat input by SAW, which makes the performance of HAZ greatly reduced after strain [7]. GTAW is having a major application in the important joint due to its high welding quality [8]. But its low productivity is restricted to join thick 304L steel because of low welding speed. Plasma arc welding (PAW), which is a kind of fusion welding process with keyhole, is effective in joining 304L steel with a thickness of 8-12mm due to its high energy density heat beam [9, 10]. But undercut and little corrosion resistance is prone to occur in weld metal because there is no filled metal during PAW. So some research on the combination of PAW and GTAW (PAW + GTAW) has been conducted to weld 304L and a good formation and mechanical properties in joint are obtained [11]. In addition, PAW + GTAW has several advantages, such as energy concentration, high operating efficiency and high welding speed as compared with single PAW or GTAW.

304L steel joint should have high quality when it is necessary to be pre-strained. Microstructure and mechanical properties of 304L joint with a thickness of 4mm by PAW is hardly changed and there is still composed of austenite and δ -ferrite without martensitic transformation in joint after strain by 4% [12]. Effect of strain rate on microstructural properties of 304L steel PAW joint with a thickness of 8mm has also been studied and it indicated that the number of dislocations and deformation twins were increasing with hardening stress and α' -martensite (α' -M) occurred in the joint [13]. There was some dislocation with high density in the fusion zone and a lot of α' -M with twins in base metal. Impact toughness of the welded joint at low temperature was still more than the maximum value from the national standard and the impact fracture showed ductile fracture with dimple after strain strengthening [14].

In this paper, study on microstructure characteristics and mechanical properties in 304L steel joint with a thickness of 12mm experiencing pre-strain by 9% has been carried out to investigate the effect of strain on tensile strength and impact toughness at -196°C in the whole joint by PAW + GTAW.

Experimental Design

In this work, SUS 304L austenitic stainless steel with a thickness of 12mm was used as the base metal for welding. The plate prepared for joining is 500mm×300mm. 308LSi filler wire with a diameter of 1.0 mm is filled in GTAW for covering layer. The chemical compositions of base metal and wire are listed in Table 1.

Before welding, the base metal surface has been mechanically polished to remove the oxide and oil stain. The plate was provided with a 60 ° V-type groove by a blunt edge of 7mm. Plasma arc welding (PAW) was used to weld 304L steel for the first layer by EWM-422 plasma arc welding machine of German IDA. There was no metal filled in the welding process and the plasma gas was a mixture of 97%Ar + 3%H₂. The welding parameters are given in Table 2. GTAW cover welding with filler wire was carried out after the completion of penetration welding by PAW. The wire feeding speed was 280cm/min and the pure argon was adopted with a flow rate of 18L/min.

High pressure water jet was used to cut some samples for metallographic observation, pre-strain treatment and mechanical properties test along the direction perpendicular to the weld. A hydraulic universal testing machine (WE-600A) was used to conduct static tension strain by 9% on joint at room temperature, as shown in Fig. 1. The loading rate was 0.1kN/s and the strain rate was 3×10^{-4} /s.

Figure 2 shows V-notch impact test samples with dimensions of 55mm×10mm×10mm extracted transverse to each weld. The locations notched are at the weld metal (WM), the heat affected zone (HAZ) and base metal (BM). The fractured joint samples are presented in Fig. 2c. Tensile strength and impact toughness were tested at 25°C and -196°C in the liquid N₂ according to the ISO148-1. The microstructure of the joint was analyzed by XJP-6A metallographic microscope and thermal field emission scanning electron microscope (SU-70).

Results And Discussion

3.1 Joint morphology and strain feature

Figure 3 shows The morphology of 304L stainless steel joint obtained by plasma arc and argon tungsten arc welding (PAW + GTAW) process. It is found that the front GTAW weld ripple is smooth and neat and the weld pool height is uniform. There is a good fusion between both sides of base metal. Back-shaping is also got by single weld with PAW in the help of keyhole effect and the back bead is fully penetrated with a good appearance. Figure 3b shows the cross-section feature of PAW + GTAW joint. The internal quality of the filled metal is uniform with deep penetration characteristics of PAW bead and a complete fusion between GTAW and PAW weld bead.

A sample with a length of 300mm including filled metal and base metal was prepared for strain treatment and the total strain was 9%. The surface of specimen was marked at an equal distance of 10mm, as shown in Fig. 4a. After strain, the distance between the every two original marked lines was measured and the result is shown in Fig. 4b. Strain varies in different region of the whole 304L joint including BM, WM and HAZ. It suggests that ability to resist deformation in BM, WM and HAZ is different although the total average strain is 9%.

The strain in the weld metal (WM) has a minimum value of about 3% and it has a peak value of 13% in HAZ. It is mainly related to the microstructure characteristics with different crystal lattice. Figure 5a shows microstructure in 304L joint got by PAW + GTAW. It can be obviously seen that there is a large number of delta ferrite (δ -F) with body-centered cubic (bcc) crystal structure in WM and austenite (A) with face-centered cubic (fcc) crystal structure in HAZ as welded.

The strain condition is known to exert a significant influence on the dislocation substructure and density [15]. There is less δ -F slip systems in bcc, which makes it easier to stack dislocations and increase deformation resistance [16]. Therefore, the strain is only 3% in WM with a lot of δ -F. The larger deformation in HAZ is attributed to the weakening of austenitic grain boundary experiencing strain, as shown in Fig. 5b. Meanwhile, there are more austenite slip systems in fcc system and martensite shearing occurs in grains, which makes stress concentration relax and reduces slip resistance and it is conducive to deformation in different orientations of austenite. These results indicate that the generation of dislocations in the HAZ provides a greater strengthening effect than that in the weld metal.

3.2. Microstructure evolution

Figure 6 shows the microstructure comparison of 304L steel joint as welded and after strain. The distribution of delta ferrite (δ -F) in GTAW filled metal is relatively concentrated and the large part is skeletal δ -F with some interfacial characteristics such as migrating grain boundary (MGB), solidified grain boundary (SGB) and solidified sub grain boundary (SSGB), as shown in Fig. 6a. The skeletal δ -F provides a greater resistance to plastic flow during strain. The microstructure in weld metal by GTAW becomes fine and dense with decreasing ferrite lath gap along the deformation, refer Fig. 6b. The interfacial characteristics with solidification grain boundary are not obvious. The structure of delta ferrite is slender

and irregular. The fine delta ferrite introduces an increased number of dislocation sources, which act as barriers to the mobile dislocations [15].

Figure 6c shows the microstructure in PAW backing layer as weld and it is mainly composed of austenite (A) and delta ferrite (δ -F). Temperature in molten pool is high and the temperature gradient is large under the action of high energy plasma arc. Therefore, Austenite is generated by eutectic reaction of ferrite cell boundary and dendrite boundary. The distribution of austenite and delta ferrite in the weld metal by plasma arc is not uniform and orientation of austenite is obvious. This is because the transformation from reaction in Fe-Ni system to eutectic reaction in Fe-Cr-Ni system is greatly affected by the diffusion of solute atoms in the front of solid-liquid interface.

Figure 6d shows the microstructure characteristic in the weld metal for PAW after strain. It is found that the grain in WM has been elongated along the deformation direction. The epitaxial growth of weld grain leads to the transformation from cellular dendrite to columnar grain, and there are fewer dense non-uniform shear deformation bands after strain. Some studies [13, 15] have shown the effect of different strain rates on the microstructure transformation of 304 stainless steel, it is revealed that these fine spacing shear bands are the product of changing the microstructure of austenitic stainless steel in the process of strain strengthening, and the dislocation in the weld is prone to pile up. In the present study, it is assumed that the difference between the closely packed lattice planes in the delta ferrite and those in the austenite does not represent the source of micro-shear band formation and martensitic nucleation. Consequently, there are fewer formations of martensite and micro-shear bands in the WM.

Some inhomogeneity of microstructure is shown in narrow HAZ affected by the welding thermal cycling shown in Fig. 7a. The grain size of HAZ is large and grain growth occurs near the high temperature zone due to the concentrated energy by plasma arc. The current high-resolution lattice image display that presence of dislocation is limited to the austenite structures shown in Fig. 7b. There is also a small amount of ferrite can be observed in the HAZ, and there is a trend of extending from the fusion zone to the HAZ along the direction of large temperature gradient. The ferrite in HAZ has no liquid-solid phase transformation but only solid phase transformation.

Figure 7c shows that there are α' -martensite (α' -M) formations and micro-shear bands in the HAZ of the weldments after strain. The formation of micro-shear bands and martensite appears to be influenced by the high density dislocation in austenite grain. The stacking fault energy of austenite is low, which provides the driving force for martensitic transformation [17]. The martensitic lath are parallel or cross distributed in austenite grains, as shown in Fig. 7d. Since the α' -M formations with different orientation overlap within the austenite, an increased volume of martensitic transformation causes a corresponding reduction of austenite, which not only influences the motion of the dislocations, but also reduces the number of twin formations. Meanwhile, More stacking faults in austenite promote the accumulation of solute atoms at the stacking faults, which increase the resistance of deformation dislocation expansion to play a strain strengthening effect [14]. The deformation dislocation, micro-shear band and their intersection location will become the preferential nucleation area of martensitic transformation [15]. The

original grains in HAZ are elongated and the grains are broken with the grain boundaries disappeared after strain. A similar observation was reported by Lee and Lin [18], who stated that dense deformation slip bands and finer lath martensite were found in the grains under the larger strain.

The formation of martensite depends on stacking fault energy. With the increase of stacking fault energy, stacking fault overlap becomes more and more irregular to make martensite nucleation difficult [19]. Figure 8 shows the α' -M formation in austenite during strain. It is noted that origin of α' -M is dislocation and then occurrence of dislocation increment is found in the deformed microstructure under strain, refer Fig. 8b. When dislocations are more and more, the different dislocation will meet and merge each other, refer Fig. 8c. The dynamic mechanical behavior at a specific strain can be associated with the work-hardening stress, which in turn is related to the dislocation density. At this time, some stacking faults form and they are parallel in austenite grain, which provides driving force to α' -M formation, refer Fig. 8d. It suggests that the stacking faults act as precursor to the deformation-induced formation of α' -M at larger strain of 13%. Meanwhile, some dislocations and even stacking faults with different orientation occur and they slip to form martensite lath with deformation, refer Fig. 8e. Finally, the martensite lath with network is shown in austenite grain, refer Fig. 8f.

The microstructure evolutions in the HAZ of 304L joint reveal that the α' -M structures and micro-shear bands are both formed in the austenite grain experiencing strain by 13%. This is attributed to the motion of different orientation dislocation under the condition of deformation. It is also noted that the larger deformation is in the joint, the finer martensite lath can be formed.

3.3. Effect on mechanical properties

The present microstructural observations provide insights into the evolution of the dislocation slip and martensite transformation after strain. However, it is also desirable to study the effect of strain on mechanical properties of weldment in order to better understand the characteristics of the 304L steel joint by PAW + GTAW under high strain deformation. Figure 9 depicts that tensile strength is greatly different between the joint as welded (AW) and that after strain (AS). It is also noted that there is a significant effect of strain on tensile strength and fracture location of 304L steel weldment at both room temperature 25°C and low temperature of -196°C.

It can be seen that the tensile strength of 304L steel weldment as welded is 700MPa and the fracture occurs in the HAZ at 25°C. The tensile strength increases to 804MPa after strain. This suggests that joint has an obvious strengthening of 104MPa induced by strain and it is more apparent at -196°C. The tensile strength increases from 1480MPa to 1700MPa and the fracture occurs on the base metal, refer Fig. 9b. It is related to transformation of α' -M with high hardness phase at larger strain of 13% in HAZ. Meanwhile, there are in a large number of dislocations and twins in the austenite grains. Compared to 304L steel joint as welded fracturing in HAZ, deformation of in the weld metal and base metal is smaller.

In addition, the dislocation accumulation and increment in austenitic grain may form cellular structure, and the deformation twins produced in the process of strain can also induce strengthening effect, which improves the strength of 304L steel weldment. There is obvious necking phenomenon in the tensile

fracture at 25°C, as shown in Fig. 10a. However the fracture at low temperature of -196°C is relatively flat, and the martensitic transformation structure shows greater brittleness under low temperature stress, as shown in Fig. 10b. The tensile fracture morphology at -196°C is obvious brittle feature with much more cleavage step pattern.

Figure 11 shows the impact absorption energy of 304L weld joint at low temperature (-196°C). It is noted that the impact toughness in base metal (BM), the heat affected zone (HAZ) and weld metal (WM) decreases after strain treatment. Especially, the impact toughness in HAZ reduces to 94J from 116J as welded. This can be explained by the microstructure characteristics after strain. There are a lot of ferrite in the WM which can limit the sharp decreasing of ductility and toughness. The grain boundary can prevent the dislocation from slipping resulted to higher impact toughness than that in HAZ. While the less impact toughness in HAZ is mainly due to the formation of deformed martensite with high strength, high hardness and low toughness. The crystal distortion and dislocation density increase in austenite grain after strain strengthening, which leads to the decrease of impact toughness at -196°C. These deformed martensite lath and dislocations accumulated at grain boundaries result in stress concentration and impact toughness reduction after strain. The fracture morphology with more deep cleavages is also shown in Fig. 11b compared to that as welded.

Figure 12 shows the representative impact fracture surfaces of the specimens both notched at the weld metal (WM) and the heat affected zone (HAZ) under the condition of as welded and after strain. As shown in Fig. 12a and Fig. 12b, most of pattern is dimple zone in fracture surface of WM as welded and there are fewer tearing with elongated dimple shown in WM after strain. The stress-fracture regions as welded are covered with ~ 5µm-diameter and ~ 50µm diameter dimples. The small-dimple area in the specimen as welded shows deep holes and the corresponding impact toughness is up to 122J. A comparison of the fracture surfaces of the specimens strained shows that the area of dimple is less with more obvious tearing feature than that in as welded specimens. The diameter of the large dimples increased with decreasing small-dimple area. The strain conducted has decreased the average size of the dimples in WM.

Figure 12c shows the fracture morphology of impact toughness tested specimens in HAZ as welded. It is a mixture of dimple and few quasi-cleavages, which reflects an interaction between the cleavage cracks and grains suffered plastic deformation [20]. Dimple fracture is a known kind of ductile fracture which occurs during high energy absorption process. However, quasi-cleavage is a kind of brittle fracture. It can be concluded that the fracture of the as welded specimen exhibits a mixed-mode of brittle and ductile rupture. Large numbered cleavage steps occur in the impact fracture surface after strain specimens as shown in Fig. 12d. In comparison to Fig. 12c, the cleavage area is found to be larger, which indicated an identical brittle fracture feature with less toughness after strain. This is distributed in the true strain treated by 13% deformation in HAZ. There are a lot of micro defects such as dislocations in the grain caused by martensite sheared in HAZ after strain, which leads to the stress concentration and the formation of microcracks in the process of impact loading. The micro-crack accelerates the crack growth and is unfavorable to the impact toughness.

Conclusion

In the present work, the effects of strain on the microstructure evolution, tensile properties and impact toughness at a low temperature of -196°C of SUS 304L stainless steel weldments were investigated using pre-strained specimens containing the weld metal (WM), the heat affected zone (HAZ) and base metal (BM), which was achieved by PAW + GTAW process. The conclusions can be summarized as follows:

- (1) The deformation of 304L joint shows inhomogeneity at an average strain by 9%. The deformation of the weld metal is only 3% and it is 13% in the heat affected zone.
- (2) The skeletal delta ferrite in the weld metal for GTAW covering layer becomes fine and dense with decreasing ferrite lath gap along the deformation and crystal orientation of austenite in PAW backing layer is elongated along the deformation direction after strain.
- (3) There are martensite formations and micro-shear bands occurring in the HAZ regions of the weldments after strain and the finer martensite lath are cross distributed in austenite under the larger strain of 13%.
- (4) The origin of α' -M transformation in HAZ is dislocation and then occurrence of dislocation increment is found in the deformed microstructure. Subsequently, stacking faults act as precursor to the deformation-induced formation of α' -M at strain and some stacking faults with different orientation occur and they slip to form martensite lath with larger deformation.
- (5) There is a significant effect of strain on tensile strength. The tensile strength of 304L steel weldment as welded is 700MPa and it increases to 804MPa after strain. Meanwhile, strengthening is more apparent at -196°C and the tensile strength increases from 1480MPa to 1700MPa in specimen after strain.
- (6) The impact toughness at -196°C in base metal, the heat affected zone and weld metal decreases after strain treatment, especially the impact toughness in HAZ sharply reduces to 94J from 116J as welded. Correspondingly, a mixed-mode of brittle and ductile rupture as welded is also changed into large numbered cleavage steps after strain.

Declarations

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Ethical Approval

The submitted work is original and not has been published elsewhere in any form or language. The manuscript is not submitted to more than one journal for simultaneous consideration.

Consent to Participate

Not applicable.

Consent to Publish

Not applicable.

Authors Contributions

All authors read and approved the final manuscript.

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Competing Interests

Not applicable.

Availability of data and materials

The data and materials used and/or analyzed during the current study are included in this article and they are available from the corresponding author on reasonable request.

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Tables

Table 1 Chemical compositions of 304L steel and 308LSi filler wire (%).

Element	C	S	P	Si	Mn	Mo	Cu	Cr	Ni	Fe
304L	0.06	0.012	0.026	0.40	1.62	—	—	18.03	8.14	Bal.
308LSi	0.03	0.03	0.03	0.9	1,5	0.7	0.7	21.5	10.5	Bal.

Table 2 Welding parameters of PAW+GTAW process.

Groove	Process	Current <i>I</i> /A	Welding speed <i>v</i> /cm•s ⁻¹	Heat input <i>E</i> /kJ•cm ⁻¹	Plasma gas /shielding gas
PAW	280	0.35	24.0	97%Ar+3%H ₂
	GTAW	270	0.35	22.5	100%Ar

Figures

Figure 1

Static tension strain by 9% on 304L joint.

304L joint by PAW+GTAW: (a) 304L joint morphology and (b) cross-section feature.

Figure 4

Strain measurement for joint: (a) measure location and (b) strain in different region of joint.

Figure 5

Microstructure in 304L joint (a) as welded and (b) after strain.

Figure 6

Microstructure characteristic in the weld metal: (a) GTAW as welded, (b) GTAW after strain, (c) PAW as welded, and (d) PAW after strain.

Figure 7

Microstructure characteristic in the HAZ: (a) as welded, (b) grain feature, (c) after strain, and (d) α' -M.

Figure 8

α' -M formation in austenite during strain: (a) Dislocation (b) increment (c) merge (d) stacking fault (e) slip, and (f) network.

Figure 9

Tensile strength of 304L joint: (a) effect of strain and (b) fracture location.

Figure 10

Fracture morphology of joint after strain: (a) at 25°C and (b) at -196°C.

Figure 11

Impact toughness of 304L joint at -196°C: (a) effect of strain and (b) fracture morphology.

Figure 12

Fracture morphology for impact specimens: (a) WM as welded, (b) WM after strain, (c) HAZ as welded, and (d) HAZ after strain.