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The Effect of Welding Process on the Chi Phase Precipitation in As-welded 317L Weld Metals

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Three multipass 317L welds were prepared using submerged arc welding (SAW), flux-cored arc welding (FCAW) and manual metal arc welding (MMA), respectively. A microstructural study was undertaken by trepanning 3 mm discs from specific positions in the weld gap, and examining the thinned specimens by transmission electron microscopy (TEM). It was found that, in addition to Chi phase precipitation detected in the regions where overlapping of the welding passes occurred, the large heat input for each welding pass can also induce Chi phase precipitation during the cooling of the welding pool. The mechanical properties showed that there were few differences resulting from the three processes. The intermetallics which formed due to high heat input passes did not appear to have a particularly harmful effect on the mechanical properties.

KEY WORDS: Chi phase; delta ferrite; heat input; transmission electron microscopy; welding.

1. Introduction

Stainless steel welds can suffer from the following problems: solidification cracking, hydrogen cracking, poor HAZ toughness and the effects of precipitation reactions. Within the majority of the austenitic stainless steels, solidification cracking can be counteracted by controlling the volume fraction of delta ferrite in the weld structure.[1)] This can be achieved by correct consumable selection. The reason for the beneficial effects of delta ferrite are well documented.[2)] Hydrogen cracking and HAZ toughness problems are not specific to austenitic grade stainless steels,[3)] and tend to occur also in ferritic stainless steels. Precipitation reactions at one time caused harmful effects such as weld decay,[4)] and the precipitation of certain intermetallic phases has been associated with a decreased corrosion resistance in some austenitic stainless steels.[5)] Normally the presence of Sigma and Chi phases, which result from delta ferrite decomposition, has been responsible for the decrease in corrosion resistance.

Previous investigations[6–10)] have shown that the intermetallic phases can only precipitate in the weld structure after ageing at elevated temperatures (650–800°C) for several hours. Recent work,[11)] however, using analytical TEM identified the Chi phase in multipass 316L/317L weld metals in the as-welded condition. It was also established that Chi phase precipitation in 316L/317L welds was closely related to the region where overlapping welding passes occurred. In fact, the weld structure in as-welded multipass 316L/317L welds could be complex due to the effect of microsegregation, the varied heat input for each welding pass and the cooling rate. Furthermore the structure also contained areas that had been reheated and cooled down due to the lay-up of subsequent welding passes, which may display a remarkably different microstructure from that found in other regions.

The present work compares the weld structure of 317L welds produced by submerged arc welding (SAW), flux-cored arc welding (FCAW) and manual metal arc welding (MMA). An attempt was made to establish a correlation between intermetallic precipitation and the welding techniques, including the varying heat inputs of the welding passes, by examining the microstructure taken from specific positions within the weld gap.

2. Experimental Method

Three specimens, labelled SAW, FCAW and MMA respectively, were produced by using respectively submerged arc, flux-cored arc and manual metal arc welding techniques to weld 18–19 mm thick 316LN stainless steel plate. A solid 317L 3.2 mm diameter wire and a basic non-compensating flux were used for SAW, a flux cored 317L wire with 1.2 mm diameter was used for FCAW, and a 317L-140 stick electrode was used for MMA welding. The welding arrangement for each technique and the resulting weld structure are illustrated in **Fig. 1**, together with the heat input for each welding pass used in the three weldments. The average chemical compositions of the parent plate and of each of the overall welds are given in the **Table 1**.

A thin section (1 mm thick) of the welds was cut after welding, and small discs (3 mm diameter) were trepanned from specific positions on the welds, as indicated by the circles in Fig. 1. The discs were then ground to a thickness of approximately 100 microns and finally thinned with a twin-jet polisher using a mixture of perchloric acid, acetic acid and ethanol in ratio of 1:4:1. The electrolyte was maintained at 5–10°C with an applied voltage of 20 V. During the microstructural observations, the difference in the microstructure from area to area was evaluated from an estimated percentage of the delta ferrite and the intermetallic phase, by averaging data from 30–40 different fields of view. The measuring error for the delta ferrite phase was $\pm 1\%$ and for the intermetallic precipitates $\pm 0.1\%$, both of the measured value.

3. Results

In the microstructural study, residual delta ferrite and spherical particles were widely distributed in all positions of all three welds. The delta phase, which is not present in the parent plate, was introduced deliberately to counteract solidification cracking during the cooling of the welds. Randomly distributed spherical particles with a size range from 0.2 to 2 microns (the majority between 0.4 to 0.5 microns) are microslag inclusions formed during the welding process. Energy dispersive X-ray analysis (EDX) indicated that the microslag inclusions contained mainly Ti, Cr, Mn and Si. No significant difference was found in the concentration and the chemical composition of the microslag inclusions within the three welds. In addition to the delta ferrite phase and the microslag inclusions, precipitates of an intermetallic phase were also observed. These precipitates were present to a different extent from place to place, in each position of all three welds. The estimated percentages of the delta ferrite and the intermetallic phase for each particular position marked 1 to 5 in Fig. 1, are given in **Table 2**.

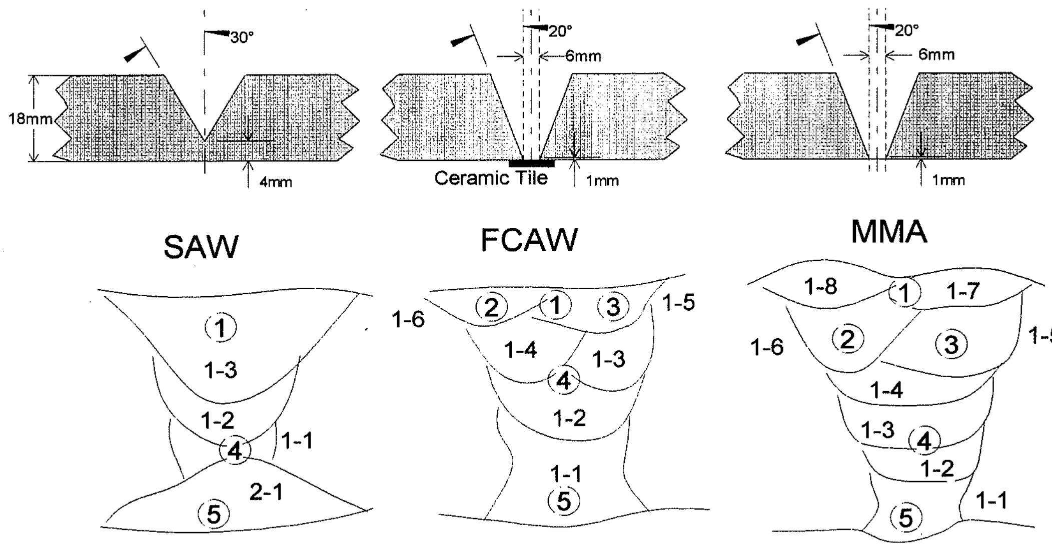
All the intermetallic precipitates observed in the investigation were nucleated at the delta ferrite/austenite boundaries and grew as individual particles. The results given in Table 2 indicate that the intermetallics were observed only in the reheated zones (overlapping area of welding passes), and in the welding passes which had

Table 1. Chemical composition of the parent plate and overall welds. (wt%)

	Cr	Ni	Mo	Mn	C	Si	Nb	N
Parent plate	17.73	10.74	2.80	1.42	0.019	0.44	--	0.16
SAW	18.40	12.70	3.60	1.52	0.007	0.65	0.076	0.06
FCAW	18.40	12.70	3.60	1.29	0.024	0.81	0.014	0.08
MMA	17.90	12.40	3.70	0.92	0.014	0.79	0.014	0.08

Table 2. Estimated percentage of delta ferrite phase and intermetallics in 317L welds.

Positions	1	2	3	4	5
% delta ferrite SAW % intermetallics	8.1 0.21	-- --	-- --	7.4 0.54	6.8 0.19
% delta ferrite FCAW % intermetallics	7.6 0.36	9.6 0	8.8 0	6.1 0.49	7.6 0.59
% delta ferrite MMA % intermetallics	8.3 0.39	6.8 0.13	7.5 0.16	7.9 0.57	6.3 0



Heat Input Tables (kJ/mm)

1-1	1.38
1-2	1.94
2-1	1.90

1-1	2.46
1-2	2.10
1-3	1.42
1-4	1.61
1-5	1.36
1-6	1.06

1-1	1.49
1-2	1.41
1-3	2.21
1-4	1.95
1-5	2.08
1-6	1.83
1-7	1.04
1-8	1.04

Fig. 1. Welding arrangement and resulting weld structures for three weldments used in the present study.

a relatively large heat input of more than 2 kJ/mm. In the SAW sample, more intermetallic precipitates were detected in position 4, where the region was reheated by a subsequent welding pass and then cooled down. However, a relatively smaller amount of intermetallics was also observed in positions 1 and 5, where neither positions are in the reheated region. The intermetallic precipitation in these regions could be due to the large heat input used for the welding passes. In the FCAW weld, considerable intermetallic precipitation was found in positions 1 and 4, but not in positions 2 and 3. An exceptional observation made in position 5 was due to the large heat input used in the pass together with an insulating effect exerted by a ceramic tile deliberately placed at the bottom of the weld gap. Similar observations were made in the sample of MMA, where intermetallics were observed in positions 1 and 4, but little or no precipitation was detected in positions 2, 3 and 5.

SAED (selected area electron diffraction) analysis revealed that most of the intermetallics at the delta ferrite/austenite boundaries were Chi phase, with a size range of 0.05–0.1 mm. EDAX data gave the average composition of the Chi phase as 53.5 % Fe, 28 % Cr, 5 % Ni and 13.5 % Mo. **Figures 2, 3** and **4** are the examples of the Chi phase observed in the samples of SAW, FCAW and MMA respectively. A schematic indexed solution is attached to all three diffraction patterns. In Figs. 2 and 4, the diffraction patterns are more diffuse and complicated due to more than one particle contributing to the pattern. Each has a slightly different orientation due to more than one particle contributing to the pattern. However, a close study found that the zone axes of the patterns are [111] and [014]. The lattice parameter of the cubic Chi phase obtained from the diffraction data is 8.91–8.92 Å. Details of the microstructures found in FCAW are given in another publication.[11]

The tensile testing results are given in **Table 3**. It is shown that the weld microstructures all have similar yield and ultimate strengths, but the ductility of the FCAW weld is poorer than that of the SAW and MMA welds.

4. Discussion

Previous investigations[6–10] on welded 316 stainless steel claimed that intermetallic phases, such as Sigma and Chi phases, were only observed after the welds were aged at elevated temperatures (650–800°C) for several hours. There are also suggestions[12] that the solidification

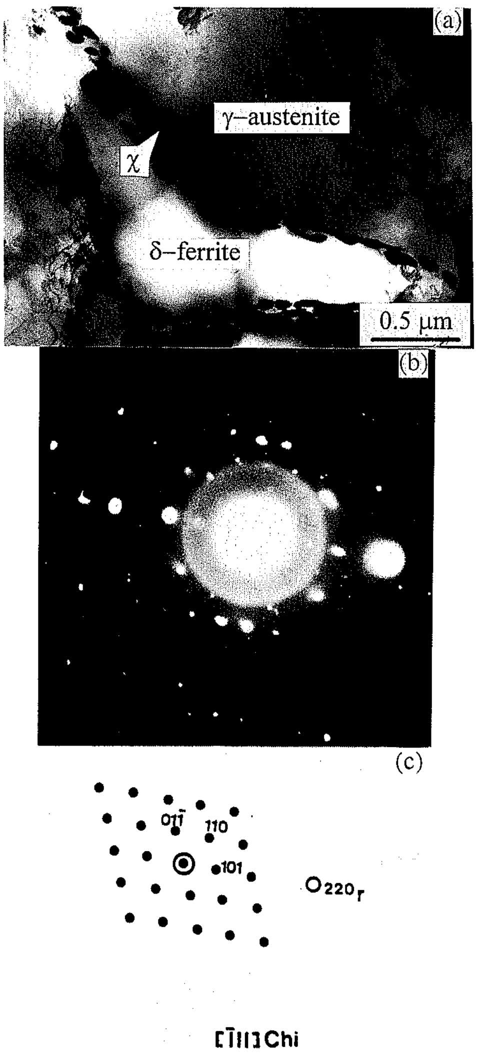


Fig. 2. Typical Chi phase precipitation observed in position 4 of SAW sample.

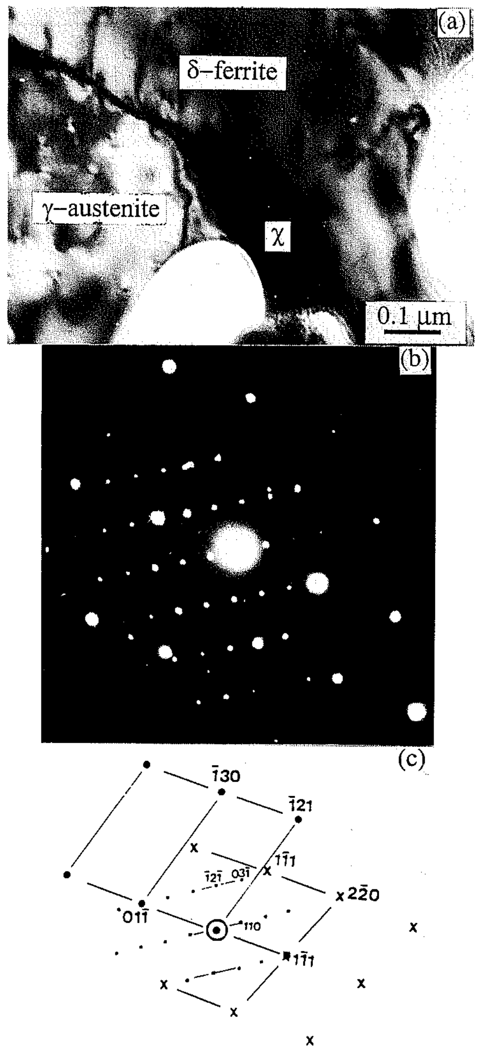


Fig. 3. Chi phase precipitation observed in position 1 of FCAW sample.

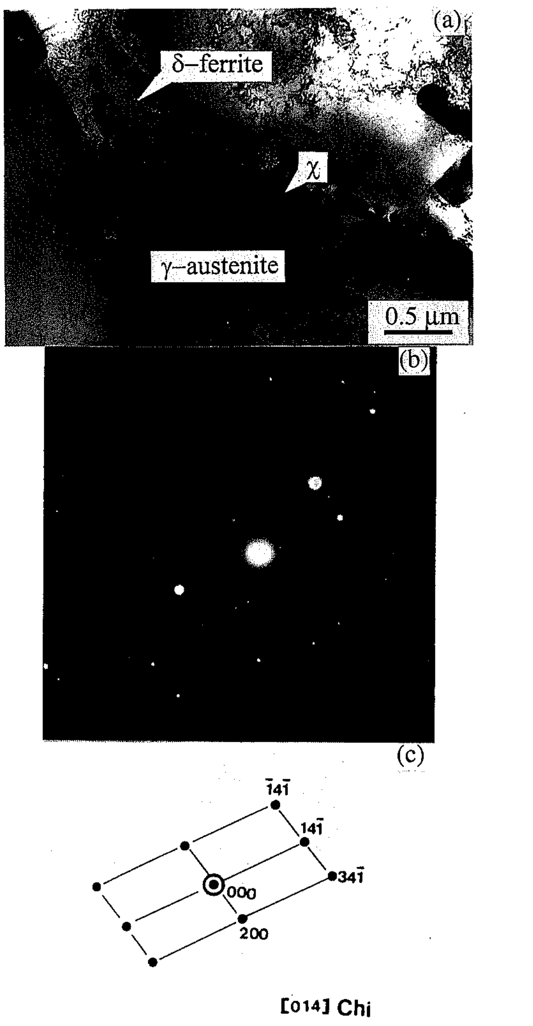


Fig. 4. Chi phase precipitation observed in position 4 of MMA sample.

Table 3. Mechanical properties of the as deposited weld metals.

Process	Y.S. ($N\ mm^{-2}$)	U.T.S. ($N\ mm^{-2}$)	EL. (%)	R.of A. (%)	Energy Absorbed at Temp.(°C) 20	-20	-60	-100
SAW	482	718	33.6	54.3	67	54	42	30
FCAW	501	699	31	37	68	63	47	38
MMA	483	672	40	63.5	78	68	54	33

mode of an austenitic stainless steel is a key factor in deciding whether or not an intermetallic phase will form. An exception, reported by Cieslak *et al.*[13] on an investigation of autogenous TIG 316 welds identified two kinds of Chi phase in the as-welded deposit. One kind was a solid state Chi phase, which formed during the solid state transformation of delta ferrite to austenite, and can only be observed in weld metal which has undergone a primary austenite solidification mode. The other kind was a eutectic Chi phase, which formed directly from the liquid phase, and can be observed in welds solidified as primary delta ferrite or primary austenite.

The carbon and silicon contents can also have an influence on the intermetallic precipitation in 316 weld metals. Bannister and Farrar[14] reported that a higher C content (0.1 %) and a lower Si content (0.25 %) promoted the formation of the carbide $M_{23}C_6$, but suppressed the formation of the intermetallic phase, whereas the lower C content (0.05 %) and higher Si content (0.8 %) enhanced the precipitation of the intermetallic phase, but with little carbide formation in the 316 weld metal. In addition, the lower nitrogen contents in the weld metal will lead to a greater stabilisation of the delta ferrite, which in turn would potentially increase the proportion of Chi phase within the weld metal. The materials used in the present study have a higher Mo content (3.6 %) than alloys used in most of the previous investigations (where the average Mo content was less than 2.5 %). It was suggested in Cieslak's work[13] that a high Mo content is the driving force for the precipitation of the Chi phase. In addition, the lower C content (0.02 %) and the higher Si content (0.7 %) in the present weld metal also favour the precipitation of intermetallic phases. Furthermore, during the present microstructural observations, the commonly reported $M_{23}C_6$ carbide, which has an extremely high Cr content, was not detected in any of the three welds. This observation is also consistent with the conclusion given by Bannister and Farrar.[14]

In the present study, the intermetallic precipitation of Chi phase in 317L welds is closely related to the specific position within the weld metal and the heat input for each welding pass, but it is not influenced by the particular welding technique applied. In the FCAW and the MMA samples, the weld gap was filled by 6–8 welding passes, while in SAW sample only 4 passes were used. This difference indicates that a smaller heat input (<1.5 kJ/mm) will avoid intermetallic precipitation in the weld pool during cooling. However, the reheating and cooling from the lay-up of subsequent weld passes will cause the precipitation of Chi phase in the reheated zones.

The welded microstructure from each process produced broadly similar mechanical properties with the exception of the ductility of the FCAW weld. The tensile specimen for the FCAW was taken from an area above position 4, (Fig. 1), towards the area containing low levels of intermetallic levels. For the MMA weld, the tensile sample was machined from regions near to positions 2 and 3 *i.e.* areas containing zero intermetallics phase. It is feasible that the presence of intermetallics can cause local deterioration in ductility.

The impact properties, although similar, for the three processes show the effect of the coarse grain structure of the SAW weld compared to the finer structures produced by the MMA and FCAW processes using the lower heat impact welding. There may be a beneficial contribution from the lower volume of intermetallic phase in the corresponding positions in FCAW and MMA welds.

5. Conclusions

Intermetallic precipitation of Chi phase was detected in the as-welded state in all three 317L welds made by SAW, FCAW and MMA welding techniques. Higher molybdenum (3.6%) and silicon (0.07%) contents and a lower carbon (0.02%) content in the present alloy probably account for the high precipitation kinetics of the Chi phase in the weld metal. As far as the microstructure is concerned, no significant difference was found within the three welds. However, in addition to the Chi phase precipitation detected in the reheated zones, it was also found in the weld passes, where a relatively large heat input was used during the welding. Mechanical test data indicates that the mechanical properties were similar for each weld and that slight differences could be explained by the position of the sample on the volume or the intermetallics within the weld metal.

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