Effect of the Number of Passes on the Structure and Properties of Submerged Arc Welds of AISI Type 316L Stainless Steel

As the number of passes increased, the hardness and tensile strength increased, while the ductility and toughness decreased

KEY WORDS
SAW 316L Stainless
Effect of No. Passes
No. Passes and 316L
Effect on Structure
Effect on Properties
316L Weld Properties
316L Weld Hardness
Tensile Strength
316L Tensile/Ductile
316L Weld Toughness

ABSTRACT. AISI Type 316L stainless steel plates were submerged arc welded using 5, 9 and 13 passes. With an increase in the number of passes during welding, hardness and tensile strength properties of the welds increase systematically, while their ductility and toughness decrease progressively. These changes in the mechanical properties could be correlated with the observed microstructural characteristics, particularly the amount, morphology and distribution of the delta ferrite.

Introduction
Shielded metal arc welding (SMAW) and gas tungsten arc welding (GTAW) are used for joining many stainless steel components. However, these processes prove to be uneconomical if plates thicker than 10 mm (0.4 in.) are to be welded, since the number of passes involved would be considerably high. If the submerged arc welding (SAW) process can be used for such applications, then it could lead to a significant reduction in the number of passes required for completing the joints. In addition, the SAW process would offer high deposition rates, thermal efficiency and weld metal recovery (Ref. 1). The biggest handicap, however, with the application of SAW to welding of austenitic stainless steels is its high heat input and the consequent reduction in toughness and resistance to intergranular corrosion.

In recent years, low-carbon grades of austenitic stainless steels have been employed widely, due to their improved resistance to intergranular corrosion and intergranular stress corrosion cracking. Types 304L and 316L stainless steels with carbon levels as low as 0.01% are manufactured regularly, although the allowable maximum carbon content of these steels is 0.03%. The wider exploitation of low-carbon grades of stainless steels allows greater use of high-deposition rate processes like SAW for thick stainless steel sections (Ref. 2). However, there is little information available on the properties of submerged arc austenitic stainless steel weldments with multiple passes. Multiple-pass welds exhibit considerable heterogeneity in microstructure and significant differences in mechanical properties along the thickness of the weld (Ref. 3). This can be attributed to the complex nature of thermal cycling experienced by each pass as subsequent passes are deposited. In other words, the microstructures and mechanical properties of these welds are a result of the complex interplay between the effects of heat input and the number of passes used.

A systematic attempt has been made to investigate the effects of the number
Microstructural Characterization

Transverse sections of the welds were prepared for examination under an optical microscope. Etching with boiling Murakami's reagent, containing 10 g of KOH, 10 g of K$_3$Fe(CN)$_6$ and 100 mL of distilled water, for 4 min, revealed the delta-ferrite morphology. Delta-ferrite measurements were made across the thickness of the welds at equal intervals of ~1.3 mm (0.05 in.) by employing a Magne-Gage calibrated according to AWS A4.2-74.

Evaluation of Mechanical Properties

Vicker's hardness profiles were obtained at equal intervals of ~1 mm (0.04 in.) along the thickness of the weld by using a 10-kg load. All-weld metal tensile specimens, as shown in Fig. 2, were machined from the weldments and were tested in tension using a nominal strain rate of 3.2 x 10$^{-3}$ s$^{-1}$. The tensile tests were conducted both at room temperature and at 500°C (932°F).

Impact Testing

Subsize (5- X 10- X 55-mm) (0.2- X 0.4- X 2.2-in.) Charpy V-notch specimens with notches located in the center of the weldments were fabricated as shown in Fig. 3. The impact specimens were tested at room temperature using an instrumented impact testing machine calibrated according to ASTM E23. Plane-stress fracture toughness (K$_c$) was determined from the load-time and integrated energy-time plots (Ref. 4).

Results

Microstructural Observations

The general microstructure of all the submerged arc welds revealed a duplex structure containing delta-ferrite in the austenite matrix. The austenite matrix consisted of columnar grains which grew epitaxially from the fusion boundary across the interpass regions. The columnar grain size and the interdendritic spacing between the secondary arms of the delta-ferrite were found to increase with a decreasing number of passes. The absence of microfissuring can be attributed to the duplex microstructure (Refs. 5-8). The ferrite number (FN) and ferrite morphology varied significantly from the root region to the surface of the welds. The FN measured near the surface, i.e., in the final pass region, is about 10 FN, regardless of the total number of passes employed to make the weld. However, the FN value measured near the root region varied significantly with the number of weld passes. It increased from 2.5 FN for Weld C (5 passes) to 7 FN for Weld A (13 passes) — Fig. 4. The ferrite morphology near the root regions is found to be globular, while the ferrite morphology near the surface is found to be vermicular (Figs. 5, 6). Mean spacing between the secondary arms of the dendrites was found to be 8 and 12 μm for Welds A and C, respectively. These values are in good agreement with the results obtained by scanning electron microscopic observation of samples which had been subjected to selective dissolution of austenite (Ref. 9).

Hardness and Tensile Properties

Hardness surveys made across the thickness of the welds revealed maximum hardness values at the root region — Fig. 7. The maximum hardness values and the hardness at other regions
were found to increase with the increase in the number of passes.

Tensile properties of the all-weld metal specimens tested at room temperature and at 500°C (932°F) are summarized in Tables 3 and 4, respectively. The values of yield strength (0.2% offset) and tensile strength at both these test temperatures show a decreasing trend as the number of passes is decreased. The values of percent elongation and percent reduction in area show an increase as the number of passes are decreased. Furthermore, it is noticed that the strength and ductility values obtained at 500°C are much lower than those obtained by testing at room temperature.

Anisotropy of deformation occurred in the all-weld metal tensile tests, as seen by elliptical cross-sections of the fractured specimens—Fig. 8. The greatest diametrical contractions have occurred parallel to the columnar grain growth direction. Previous studies (Ref. 10) have also disclosed anisotropy of deformation in austenitic weld metal. The reasons for the anisotropic deformation could be 1) preferred crystallographic orientations, and 2) elongated substructure in the solidification direction.

Figure 9 reveals the microstructure of the tensile specimens sectioned parallel to the tensile axis and taken away from the region of necking. The secondary dendritic arm spacing is considerably higher for Weld C (5 passes) (Fig. 9B) than for Weld A (13 passes)—Fig. 9A. Figures 10A and 10B show the elongated ferrite morphology in the necked region of tensile specimens for Welds A and C, respectively. The delta-ferrite is elongated to a higher degree in Weld C (5 passes) than in Weld A (13 passes). This correlates well with the higher percentage of reduction in area undergone by the specimens of Weld C—Table 3.

Figures 11A and 11B show true stress/strain curves for Welds A and C, respectively. The stress-strain curve for Weld C is much lower than that for Weld A at both room temperature and 500°C. The reasons for the lower strength and ductility values at 500°C are similar to those discussed above for Welds A and C. The anisotropic deformation observed in the tensile tests is also present in the stress-strain curves. Figure 12 shows the true stress-strain curves for Welds A and C at room temperature and 500°C. The curves are similar to those shown in Fig. 11, but the yield strength and tensile strength at 500°C are much lower than those at room temperature.

Table 1—Chemical Composition of Base Metal and Filler Metal

<table>
<thead>
<tr>
<th>Element (wt-%)</th>
<th>Base Metal</th>
<th>Filler Metal</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
<td>0.029</td>
<td>0.014</td>
</tr>
<tr>
<td>Mn</td>
<td>0.93</td>
<td>1.40</td>
</tr>
<tr>
<td>Si</td>
<td>0.34</td>
<td>0.31</td>
</tr>
<tr>
<td>Ni</td>
<td>10.50</td>
<td>12.80</td>
</tr>
<tr>
<td>Cr</td>
<td>16.80</td>
<td>18.60</td>
</tr>
<tr>
<td>Mo</td>
<td>2.00</td>
<td>2.80</td>
</tr>
<tr>
<td>N</td>
<td>0.04</td>
<td>0.04</td>
</tr>
<tr>
<td>Fe</td>
<td>Balance</td>
<td>Balance</td>
</tr>
</tbody>
</table>

Fig. 4—Delta-ferrite profiles measured by Magne-Gage method, across the thickness of the welds.

Fig. 5—Microstructure of Weld A (13 passes). A—Near the surface of the weld (~10 FN); B—Near the root of the weld (~7 FN)

Fig. 6—Microstructure of Weld C (5 passes). A—Near the surface of the weld (~10 FN); B—Near the root of the weld (~2.5 FN)
true plastic strain plots for specimens tested at room temperature and at 500°C, respectively. It can be seen that the flow stress increases and uniform elongation decreases as the number of passes are varied from 5 to 13. The flow stress and the uniform elongation values obtained at 500°C are significantly lower than those obtained at room temperature. True stress/true plastic strain curves are found to follow Ludwigson's equation (Ref. 11), given by:

\[ \sigma = k_1 \varepsilon_p + \exp(k_2 + n_2 \varepsilon_p) \]  

(1)

where \( \sigma \) is true stress, \( \varepsilon_p \) is true plastic strain and \( k_1, n_1, k_2, n_2 \) are constants whose values depend on material composition, material condition, test temperature and strain rate. Significant variations in the above constants were observed for welds with different numbers of passes — Table 5. Differentiation of Equation 1 gives

\[ \frac{d\sigma}{d\varepsilon_p} = k_1 n_1 \varepsilon_p^{n_1 - 1} + n_2 \exp(k_2 + n_2 \varepsilon_p) \]  

(2)

where \( \frac{d\sigma}{d\varepsilon_p} \) is the work hardening rate. The work hardening rate decreases very sharply in the initial stages of the tests and then assumes a nearly constant value at high true stress levels. It is seen that while the initial work hardening behavior of these samples is significantly dependent on the number of passes during welding,
the saturation value of the work hardening rate is independent of the number of passes.

Impact Toughness

The results of the instrumented impact tests conducted at room temperature on subsize specimens are presented in Table 6. It is clear that the impact energies and the plane-stress fracture toughness ($K_c$) values increase as the number of passes are decreased from 13 to 5.

Discussions

Microstructure and Delta-Ferrite

As mentioned earlier, the variations in the number of passes was achieved by varying the heat input per pass. The cooling rate at the centerline of the weld can be estimated from a simple analytical solution derived by Rosenthal (Ref. 12) for three-dimensional heat flow.

\[
\frac{dT}{dt} = \frac{2\pi K (T - T_0)^2}{H}
\]

where $K$ = thermal conductivity (0.17 W/cm °C)

$T$ = temperature at which cooling rates were calculated (1425°C)

$T_0$ = initial temperature (25°C)

$H$ = heat input per pass (J/cm)

The cooling rates corresponding to the heat input values of 5.77 kJ/mm and 1.73 kJ/mm were calculated from Equation 3 as 36°C/s (97°F/s) and 120°C/s (248°F/s), respectively. This threefold increase in the cooling rate produced only a negligible variation in the delta-ferrite content of the final pass near the weld surface. The minor/negligible effect of heat input on FN has been reported by Delong and others (Refs. 13, 14). However, Suutala (Ref. 15) reported a decrease of 1 to 2 FN when the welding speed is increased from 2.5 cm/min (1 ipm) to 80 cm/min (32 ipm). High cooling rates and high welding speeds associated with laser beam welding processes were shown to reduce FN drastically, to levels less than 1 FN (Ref. 16).

The minor effect of heat input on FN can be explained as follows. During weld solidification and subsequent cooling, delta-ferrite partially transforms into austenite. 

<table>
<thead>
<tr>
<th>Weld</th>
<th>Number of Passes</th>
<th>Impact Energy (J)</th>
<th>Fracture Toughness, $K_c$ (MPa $\sqrt{m}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>13</td>
<td>29.7</td>
<td>230</td>
</tr>
<tr>
<td>B</td>
<td>9</td>
<td>36.9</td>
<td>259</td>
</tr>
<tr>
<td>C</td>
<td>5</td>
<td>45.9</td>
<td>283</td>
</tr>
</tbody>
</table>

Fig. 11 — True stress/true plastic strain curves of the all-weld metal tensile tests. A — At room temperature; B — At 500°C (932°F)
ite, leaving behind a network of skeletal ferrite. The skeletal delta-ferrite is enriched in ferrite-stabilizing elements such as Cr, Mo, etc. The kinetics of transformation of delta-ferrite into austenite are essentially dependent on the diffusion rates of these alloying elements. Higher heat input produces larger interdendritic arm spacing (larger diffusion distance) and slower cooling rates (increased diffusion time). The larger diffusion distance retards the redistribution of alloying elements, while the increased diffusion time has the opposite effect. Thus, the effects of larger interdendritic arm spacing and slower cooling rates are partially self-compensating. As a result, the FN observed in the submerged arc weld metal is not very sensitive to the heat input.

Although the variation in the heat input produced negligible change in the as-deposited FN in the final pass, it significantly affected the FN in the prior passes. The delta-ferrite is unstable at temperatures below the γ-solvus line. Thermal cycling due to multiple weld passes can lead to partial decomposition of the delta-ferrite, resulting in lower FN. The kinetics of transformation of delta-ferrite into austenite and other phases during aging is given by (Ref. 17):

\[
\ln(-\ln(1-x)) = 19.32 \cdot \frac{19,860 \cdot \ln t}{T} + K_D - \frac{m}{2}
\]

where \( x \) is the fraction transformed at temperature \( T \) after \( t \) hours. From Equation 4, it can be seen that temperature has a strong influence on \( x \), while the aging time has a comparatively minor effect. On the basis of experiments carried out using a weld thermal simulator, Lundin, et al. (Ref. 18), have shown that \( x \) increases as the peak temperature of thermal cycle is increased to 1204°C (2200°F). The \( m \) value increases still further as the number of thermal cycles is increased at a constant peak temperature of 1204°C. However, in the present investigation, it was found that FN values in the root region increased from 2.5 to 7 as the number of passes (thermal cycles) is increased from 5 to 13. The peak temperatures in the present investigation, unlike in the weld simulator experiment, were not constant, but varied according to the number of passes. The peak temperatures in the HAZ, i.e., in prior passes, increase as the heat input per pass is increased from 1.73 kJ/mm to 5.77 kJ/mm (corresponding to 13 and 5 passes, respectively). As a result of the increase in the peak temperatures, the fraction of the transformed ferrite \( (x) \) in samples welded with a higher number of passes would be small. This would lead to lower values of FN in the root regions of the samples welded with 5 passes than those samples welded with 13 passes. This is in agreement with the experimental results obtained.

Another factor responsible for the lower ferrite numbers in the root region is weld dilution with base material. Dilution is high near the root region and decreases progressively towards the weld surface. Increasing the heat input per pass also results in greater weld dilution with base material. Thus, decreasing the number of passes from 13 to 5 results in a reduction in the root FN values from 7 to 2.5.

Ferrite Morphology

Vermicular delta-ferrite with 10 FN was observed near the surface of the welds. The vermicular morphology results from incomplete \( \delta \rightarrow \gamma \) transformation (Ref. 19). The ferrite is located near or within the cores of the primary and secondary dendrite arms. Figures 5B and 6B show globular ferrite morphology at the root regions. The globular morphology of ferrite can be attributed to repeated thermal cycling of root regions, which breaks down the initial vermicular structure into small disconnected globules. David (Ref. 20) has attributed the dispersed morphology to the occurrence of shape instabilities along the length of thin ferrite needles during exposure to high temperatures for short durations.

Hardness Profiles

During welding and subsequent cooling, stresses are generated in weldments due to differential heating and cooling rates (Refs. 21, 22). During the cooling cycles, these stresses may exceed the yield strength, causing plastic deformation and work hardening in the weld metal. Since the root regions experience more thermal cycles than any other region, they undergo more deformation. Higher dislocation densities observed in the root regions (Ref. 3) support this view. The enhanced plastic deformation in the root regions is responsible for the maxima in the hardness values observed in the root. The increase in the root (maximum) hardness with an increase in the number of passes can be explained on the basis of similar arguments.

Tensile Properties

Tables 3 and 4 clearly show that the tensile properties of welds vary systematically with the number of passes used during welding. The dependence of tensile properties can be attributed to the differences in 1) chemical composition, particularly nitrogen content; 2) ferrite number (FN); 3) sub-grain size; 4) columnar grain size; and 5) deformation caused by thermal cycling.

Berggren, et al. (Ref. 23), have observed variation in tensile properties with FN, but this variation was found to be of the same magnitude as the experimental scatter. However, delta-ferrite is reported to behave as a strengthening agent in the weld metal at room temperature (Ref. 24). The yield strength and ultimate tensile strength increases substantially with accompanying reductions in percentage of elongation. On the basis of investigations of Types 308 and 309 weld metals, Delong (Ref. 13) has shown that the changes in tensile properties are related to the ferrite content and not to the variation in composition per se. It has also been suggested that the strengthening effect of ferrite at test temperatures below ~600°C is only a consequence of the reduction in austenite grain size with an increase in the ferrite content (Ref. 25). Ward (Ref. 26) has also shown that the tensile ductility depends on, and is possibly controlled by, the content, morphology, distribution and, hence, the stability of the delta-ferrite phase. Furthermore, Gusev (Ref. 27) has shown that as ferrite dendrites progressively subdivide the austenite grains, shear deformation is retarded. This leads to high strength and low ductility, and a tendency for phosphorus and calcium boundary failure. The effect of delta-ferrite on tensile properties is shown to be dependent on strain rate and test temperature. For test temperatures below ~650°C and for strain rates of the order of \( 10^{-4} \) s\(^{-1} \), delta-ferrite acts as an effective strengthening. In contrast, at strain rates and temperatures typical of creep deformation, delta-ferrite has been found to affect the creep/rupture properties quite adversely (Refs. 28, 29).

Low angle sub-boundaries form during the rapid solidification of the weld pool, due to mismatch between adjacent growing dendrites, and are reported (Ref. 30) to be persistent on reannealing. These sub-grains are comparable in size with secondary arm spacing (Ref. 31) and act as material strengtheners. Sub-grain contribution to \( \sigma_y \) can be described by a modified Hall-Petch expression (Ref. 32):

\[
\sigma_y = \sigma_y + K_D \cdot \frac{1}{D} + \frac{C(c)}{m} \cdot \frac{1}{\lambda}
\]

where \( \sigma_y \) is the Hall-Petch friction stress, \( \lambda \) is the substructure size, D is the grain size, and K, C and m are constants.

The columnar grain size (D) is found to decrease considerably with an increasing number of passes, thus contributing to the higher values of \( \sigma_y \). Furthermore, the secondary dendrite arm spacing was found to increase from 8 to 12 \( \mu \)m when the number of passes was changed from 13 to 5. This increase in sub-grain size could also have contributed to the increase in \( \sigma_y \). Finally, plastic strains generated by thermal cycling increase \( \sigma_y \). Since Weld A had experienced a higher number of thermal cycles than Weld C, it is logical to expect that the contribution to the yield and tensile strength from this factor could be larger for Weld A. Thus, it can be seen that all the three terms in

\[
\sigma_y = \sigma_y + K_D \cdot \frac{1}{D} + \frac{C(c)}{m} \cdot \frac{1}{\lambda}
\]

where \( \sigma_y \) is the Hall-Petch friction stress, \( \lambda \) is the substructure size, D is the grain size, and K, C and m are constants.
Equation 5 contribute to the high strength of Weld A.

The work hardening rate-true stress plots (Figs. 12A and 12B) show that $a_\phi$, the stress at the point of inflection, increases with an increase in the number of passes. This indicates an increase in the number of passes during welding produces a “harder” microstructure both at room temperature and at 500°C. This is in agreement with the observed variation in yield strength with the increase in the number of passes.

**Impact Toughness**

The factors which could affect the dependence of impact properties on the number of passes include: refinement of columnar grains and subgrains, plastic deformation and thermal aging, and amount and morphology of delta-ferrite.

The decrease in the impact toughness with an increase in the number of passes is consistent with the observed increase in $\phi$, values and the decrease in the percentage of elongation values. If embrittlement intermetallic phases such as $\sigma$ and $\chi$ are present in the weld microstructure, then they could result in reduced impact energies, although their influence on the tensile properties will be negligible (Ref. 33, 34). However, kinetics of their precipitation is rather slow (Ref. 35). It is unlikely that these phases could be precipitated during the short period of solidification and cooling of the weld metal. Even the thermal cycles associated with multipass welds would not lead to significant precipitation of these phases. On the other hand, $\mathrm{M}_2\mathrm{C}$ carbides can form rapidly along the delta-ferrite/austenite interphase boundaries. Carbide formation has a detrimental effect on the impact energies only when carbide films reach a critical length and thickness.

Above this critical size, the carbide films provide continuous and easy paths for crack propagation. Such carbide films could form only if the welds were aged for long durations, e.g., at 600°C (1112°F) for 3000 h (Ref. 36). This suggests that thick carbide films are unlikely to form in Type 316L stainless steel weldments during the relatively short periods of heating and cooling cycles involved in multipass welding. Thus, the contribution of carbide films to the reduction in impact energies can be ignored. Hence, we can conclude that the variation in the amount of plastic deformation induced by repeated thermal cycles is the main factor responsible for the observed reduction in the impact energies with an increase in the number of passes.

**Conclusions**

An increase in the number of passes during welding results in an increase in the minimum delta-ferrite content in the root region of the weld, and a decrease in the difference between the delta-ferrite contents at the surface and root of the weld. Furthermore, a systematic increase in the hardness and tensile strength properties and a decrease in the ductility and impact properties of the weld metal have been observed with an increase in the number of passes during welding.

**Acknowledgments**

The authors wish to thank Dr. P. Rodriguez, Head, Metallurgy Program, for his constant encouragement and support for this investigation. They also acknowledge the cooperation and help received from Messrs P. R. Sreenivasan and S. Venugopal in the course of this experimental program.
Becher, contained two errors on page 27-s. Corrections are as follows:

1) The last sentence in the second paragraph of the second column should read, "In addition, the critical adherence fracture energies \( \gamma_{\text{c}} \) or the critical adherence fracture toughness values \( \Gamma_{\text{c}}\) obtained can be related to the values of the bulk materials comprising a joint, as well as the stress state at the interface."

2) Equation 2, in the third column, should read: \[
G_{\text{ic}} = M^2 \frac{1}{\text{Elt}} = M^2 \left[ \frac{1}{(\text{Elt})_1} + \frac{1}{(\text{Elt})_2} \right] \quad (2)
\]