Low-Transformation-Temperature (LTT) Welding Consumables for Residual Stress Management: Consumables Development and Testing Qualification

Low-transformation-temperature welding wires were designed to minimize distortion due to residual stress developed during cooling of the welded assembly

BY T. ALGHAMDI AND S. LIU

ABSTRACT
Residual stresses induced during joining are major contributors to the overall stresses in the vicinity of structural weld joints. Undesirable tensile residual stress is observed in low-carbon structural steel weld joints upon completion of solidification and after the weld is cooled to room temperature. This harmful stress makes the welded joints more prone to fatigue failure. The use of low-transformation-temperature welding (LTTW) wires, producing martensitic transformation close to room temperature, was validated as a technique to manage the residual stresses and improve the fatigue strength of the welded joints. Compressive residual stresses are produced around the weld metal formed by LTTW consumables by taking advantage of the expansion associated with martensitic transformation at low temperature. When the phase-induced volumetric expansion counteracts and negates the thermal contraction, a compressive residual stress in the fusion zone is developed. This work introduces the guidelines for identifying and qualifying newly developed LTTW filler metals. Microstructures, microhardness, and distortion behavior were analyzed for welds made with experimental welding wires and compared with that in a weld produced using a commercial welding wire.

KEYWORDS
- Low-Transformation-Temperature Welding Consumables
- Martensitic Transformation
- Phase-Induced Volumetric Expansion
- Residual Stress Management
- Compressive Residual Stress
- Fatigue Strength
- Microstructures
- Microhardness

INTRODUCTION
Tensile residual stress observed in low-carbon structural steel weld joints after the weld is cooled to room temperature originates from the thermal strains developed between the newly solidified weld metal and the surrounding material. The resulting tensile residual stresses, particularly in the weld toe region, are often on the order of the yield stress of the material that lower the static and cyclic service loads that can be safely tolerated by the component, deteriorate the fatigue life of the welded structure, and promote cold cracking associated with hydrogen in high-strength structural steels. In addition, they can induce stress corrosion cracking, brittle failure, and distortion. The thermal strains give rise to distortion, which may lead to geometrical nonconformance in the welded parts (Ref. 1). Thus, the presence of compressive residual stress, instead of the typical tensile residual stress, will improve the fatigue properties of the structural welded joint by making crack nucleation and propagation more difficult (Refs. 2–5).

Several thermal and mechanical methods can be applied to relieve any harmful tensile residual stresses and increase the fatigue strength of the welded component. Postweld heat treatment and peening are processes known to reduce the tensile residual stress and possibly induce a compressive stress. However, both approaches are considered impractical since they require more time and money spent in the production of a component. Thermal stress relief treatment is also undesirable because it accelerates hydrogen migration in high-strength structural steel and may lead to hydrogen-related failures. Postweld heat treatment is one form of thermal treatment normally used for

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reducing and redistributing the residual stresses in the welded component that have been introduced by welding. The word “thermal” was used to distinguish the thermal processes from the vibration, acoustic processes. Modifying the welding process can affect the magnitude of tensile residual stress generated in the weld. For example, low heat input and a small weld pool are known to reduce residual stress (Refs. 6–8).

More recently, welding consumables were successfully designed to induce compressive residual stresses near the weld toe regions via martensitic transformation. This compressive residual stress can be significantly larger than the one generated by mechanical or thermal rework of the weldment. Martensite transformation start (\(M_s\)) and finish (\(M_f\)) temperatures are essential parameters in residual stress development in the weld. In normal processing of a typical structural steel weld, contraction of the weld metal occurs as the weld joint cools down, leading to the induction of tensile residual stresses. The expansion that accompanies martensite transformation can counteract and mitigate the effects of thermal contraction. The transformation temperature, \(M_s\), can be adjusted by means of varying the chemical composition, in particular, nickel and chromium. Alloy composition is critical to the generation of compressive residual stresses and also the magnitude of these stresses. If the martensitic transformation occurs at a high temperature, the thermally induced contraction that takes place between this temperature and room temperature may likely exceed the martensite expansion, precluding the establishment of a final compressive residual stress state. Thus, it is preferred that the martensite transformation initiates at a temperature as low as possible and finishes just above the final temperature to which the weldment is expected to cool, as shown in Fig. 1. With compressive residual stress established in the weld joint, the fatigue life of the deposited metal will greatly increase (Ref. 1).

The austenite-to-martensite transformation is accompanied by an approximately 5% increase in volume and when martensite pushes against a rigid structure (steel outside the weld zone), compressive residual stress will result. The desirable compressive residual stress is based on the relative unit cell size differences between body-centered tetragonal (BCT) crystal structure of martensite and the face-centered cubic (FCC) structure of austenite (Refs. 7, 9–11, 25–27). Presence of phases other than martensite can reduce the magnitude of expansion and compression achieved during martensitic transformation. The actual magnitude of the expansion will also depend on the content of the different alloying elements. As a result of the different physical, chemical, and electronic properties of these alloying elements, the austenite-to-martensite transformation in different steels may occur at different temperatures, thus affecting the nature and magnitude of the resulting residual stresses. Even though the chemical composition of the alloy can be modified to adjust the transformation temperature to the appropriate value, the exact nature and extent of the effects of chemical composition on the metallurgically induced compressive residual stresses has not yet been clearly evaluated.

A decade ago, Ohta et al. developed a low-transformation-temperature (LTT) welding solid wire with a 10Cr-10Ni composition that
would improve the fatigue strength of steel welded joints without requiring post-heating treatment. Residual compressive stress was induced by martensitic transformation (and expansion) of the weld metal near room temperature (Refs. 12–15). Similar results were observed by C. Miki and K. Anami (Ref. 16) and N. Suzuki et al. (Ref. 17).

This paper summarizes the work in designing different metal-cored LTT welding wires for the purpose of developing martensitic microstructure in structural steel weld metals and generating compressive residual stresses surrounding the weldment. Findings of this work greatly assist in the development of leaner and more economical LTT welding consumables. Even with leaner chromium, nickel, and manganese additions, maximum compressive residual stresses can still be achieved.

**Experimental Procedures**

Four filler metals with low martensitic transformation temperatures were designed and tested in this work. These compositions were selected from a sizable matrix that spanned across the entire martensite field on the Schaeffler constitution diagram. The Self and Olson (Ref. 21) equation as shown below was used to calculate the Ms temperatures for a large number of alloys.

\[
Ms(°C) = 521 - 14.3Cr - 17.5Ni - 28.9Mn - 37.6Si - 350C - 9.5Mo + 23.1(Cr+Mo)C - 1.19CrNi \quad (1)
\]

Self and Olson included the individual effect of alloying elements such as nickel, chromium, carbon, silicon, and manganese as well as some interactive effects as demonstrated by the product terms. The effect of carbides on the Ms temperature as represented by the (Cr+Mo)C term was considered. Derived statistically based on a large number of alloys, the Self and Olson equation is expected to provide accurate results for alloys whose compositions fall within the range of their database of austenitic stainless steels (Ref. 21). The calculated results for this work were plotted in Fig. 2. In this figure, the Iso Ms contours were mapped. Despite different compositions, each alloy that lies on the same Iso Ms contours will have the same Ms temperature (Refs. 11, 25–27). This finding seems to suggest that the Ms temperature may not be a unique control indicator to represent the resulting compressive residual stresses. The first set of the designed filler metals was selected along the 200°C isotherm and the other set was selected along the isotherm of 350°C in Fig. 2.

Additional Ms temperatures for the newly developed alloyed wires were calculated using other predictive methods reported in the literature and shown in Table 1. The two additional methodologies were devised by Ghosh and Olson (Refs. 18–20), and Eichelmann and Hull (Ref. 22). Among the many more expressions found in the literature, these three

<table>
<thead>
<tr>
<th>Wire</th>
<th>Voltage (V)</th>
<th>Current (A)</th>
<th>Wire Feed Rate (in./min)</th>
<th>Actual Traveling Speed (in./min)</th>
<th>Heat Input (kJ/in.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>SO-200A</td>
<td>29</td>
<td>273</td>
<td>205</td>
<td>19</td>
<td>25.0</td>
</tr>
<tr>
<td>SO-200B</td>
<td>29</td>
<td>273</td>
<td>180</td>
<td>19</td>
<td>25.0</td>
</tr>
<tr>
<td>SO-350A</td>
<td>29</td>
<td>262</td>
<td>180</td>
<td>19</td>
<td>24.0</td>
</tr>
<tr>
<td>SO-350B</td>
<td>29</td>
<td>270</td>
<td>190</td>
<td>19</td>
<td>25.0</td>
</tr>
<tr>
<td>ER70S-3</td>
<td>29</td>
<td>290</td>
<td>390</td>
<td>19</td>
<td>31.0</td>
</tr>
</tbody>
</table>
Methodologies were developed for stainless steel alloys, which make them more appropriate to be used for the chromium-nickel alloy compositions described in this work. Ghosh and Olson proposed a model that describes the critical driving force for nucleation as a function of composition. Their model included the effects of both interstitial and substitutional solutes (see Equation 2). In this equation, $K_1$ is a constant that includes strain and interfacial energies and the defect size, $K_{\mu}$ is the athermal strength of the solute, and $X$ is the content ratio of the alloying element. The subscript $i$ refers to $C$ and $N$; $j$ refers to $Cr$, $Mn$, $Mo$, $Nb$, $Si$, $Ti$, $V$; and $k$ refers to elements $Al$, $Ni$, $Cu$, and $W$. They showed that enough chemical driving force, $\Delta G$, is required for the martensitic transformation to overcome the resistance barrier, i.e., elastic (transformation), strain energy, frictional work required by the glissile interface, additional interfacial energy, etc. Using the Kaufman (Ref. 23) equation (Equation 3), the martensite start temperature can be calculated. Ghosh and Olson also provided the parameters for a number of elements ($C$, $N$, $Mn$, $Si$, $Cr$, $Nb$, $V$, $Ti$, $Mo$, $Cu$, $Al$, $W$). These elements were grouped using data on binary alloys (Refs. 18–20).

The effect of chemical composition on the austenite-to-martensite reaction was treated in detail by Eichelmann and Hull (Equation 4). They proposed an equation for the $M_s$ temperature of steel, which included the effects of nickel, chromium, carbon, nitrogen, silicon, and manganese. In the Eichelmann and Hull equation, molybdenum ($Mo$) was not included (Ref. 22).

$$G^* = 526 - 42T_{Ms}$$ (3)

$$M_s(C) = 1305 - 41.7Cr - 61.1Ni - 33.3Mn - 27.8Si - 1667(C+N)$$ (4)

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**Table 3 — Chemical Compositions (in wt-%) of Base Metal, Welding Wires, and Weld Joints**

<table>
<thead>
<tr>
<th>Description</th>
<th>C</th>
<th>Mn</th>
<th>Cr</th>
<th>Ni</th>
<th>Si</th>
<th>Mo</th>
<th>N</th>
<th>Nb</th>
<th>Ti</th>
<th>Cu</th>
<th>P</th>
<th>S</th>
<th>Al</th>
<th>V</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Base Metal</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
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<td></td>
<td></td>
</tr>
<tr>
<td>A-36 Structural Steel</td>
<td>0.106</td>
<td>0.80</td>
<td>0.15</td>
<td>0.11</td>
<td>0.06</td>
<td>0.02</td>
<td>0</td>
<td>0.002</td>
<td>&lt;0.01</td>
<td>0.31</td>
<td>0.012</td>
<td>0.003</td>
<td>0.047</td>
<td>0.0003</td>
</tr>
<tr>
<td>SO-200A Wire</td>
<td>0.1</td>
<td>1.0</td>
<td>1.0</td>
<td>10.0</td>
<td>0.1</td>
<td>2.0</td>
<td>0.002</td>
<td>0</td>
<td>0.01</td>
<td>0.1</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
</tr>
<tr>
<td>As welded</td>
<td>0.11</td>
<td>1.07</td>
<td>0.97</td>
<td>5.83</td>
<td>0.26</td>
<td>2.11</td>
<td>0</td>
<td>0.02</td>
<td>&lt;0.01</td>
<td>0.15</td>
<td>0.008</td>
<td>0.006</td>
<td>0</td>
<td>0.01</td>
</tr>
<tr>
<td>SO-200B Wire</td>
<td>0.04</td>
<td>1.25</td>
<td>10.0</td>
<td>4.0</td>
<td>0.15</td>
<td>0.5</td>
<td>0.002</td>
<td>0</td>
<td>0.01</td>
<td>0.1</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
</tr>
<tr>
<td>As welded</td>
<td>0.11</td>
<td>1.09</td>
<td>9.04</td>
<td>2.98</td>
<td>0.29</td>
<td>0.50</td>
<td>0</td>
<td>0.04</td>
<td>0.01</td>
<td>0.16</td>
<td>0.01</td>
<td>0.005</td>
<td>0.04</td>
<td>0.01</td>
</tr>
<tr>
<td>SO-350A Wire</td>
<td>0.1</td>
<td>1.0</td>
<td>1.0</td>
<td>2.0</td>
<td>0.1</td>
<td>2.0</td>
<td>0.002</td>
<td>0</td>
<td>0.01</td>
<td>0.1</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
</tr>
<tr>
<td>As welded</td>
<td>0.08</td>
<td>1.28</td>
<td>1.12</td>
<td>1.92</td>
<td>0.25</td>
<td>1.74</td>
<td>0</td>
<td>0.04</td>
<td>&lt;0.01</td>
<td>0.24</td>
<td>0.02</td>
<td>0.009</td>
<td>0.06</td>
<td>0.01</td>
</tr>
<tr>
<td>SO-350B Wire</td>
<td>0.05</td>
<td>1.0</td>
<td>4.0</td>
<td>0.5</td>
<td>0.1</td>
<td>2.0</td>
<td>0.002</td>
<td>0</td>
<td>0.01</td>
<td>0.1</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
</tr>
<tr>
<td>As welded</td>
<td>0.06</td>
<td>1.19</td>
<td>3.57</td>
<td>0.45</td>
<td>0.24</td>
<td>1.79</td>
<td>0</td>
<td>0.03</td>
<td>&lt;0.01</td>
<td>0.22</td>
<td>0.017</td>
<td>0.009</td>
<td>0.04</td>
<td>0.01</td>
</tr>
<tr>
<td>ER70S-3</td>
<td>0.09</td>
<td>1.02</td>
<td>0.05</td>
<td>0.03</td>
<td>0.41</td>
<td>&lt;0.01</td>
<td>—</td>
<td>—</td>
<td>&lt;0.01</td>
<td>—</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>—</td>
<td>—</td>
</tr>
</tbody>
</table>
Beads on plates were deposited on ASTM A36 grade structural carbon-steel plates using the semiautomatic gas metal arc welding (GMAW) process (Fig. 3) using the parameters in Table 2. The welding gun was mounted on a track and the travel speed was controlled automatically. All welding parameters were suggested by the welding wire’s manufacturer, Devasco International, Inc. The electrodes had a diameter of 0.0625 in. (1.59 mm) and a 75%Ar-25%CO₂ gas was used for shielding. Welding was conducted using direct current electrode positive (DCEP). The welding parameters were collected using an Arc Agent 3000P data-acquisition system. Metal transfer modes during welding were recorded and analyzed. The photos in Fig. 3 were taken for the as-welded distorted plates to show the actual appearance of the weld plates. The presence of spatter aside from aesthetic reasons did not affect the outcome of distortion or compressive stresses. The chemical compositions of the welding wires, as-deposited welds, and base metal are shown in Table 3. X-ray diffraction (XRD) was carried out to measure the amount of retained austenite existing in the weld microstructures. This measurement was intended to determine whether the martensitic transformation was complete or not for each of the welding consumables. For comparison of distortion and microstructure differences, a commercial ER70S-3 electrode was also used in this research work. The morphology and heat input of the welds deposited using the experimental wires were approximately identical. The heat input of the conventional ER70S-3 cannot be identical to the heat input of the experimental wires due to the change in the recommended welding parameters. This is considered advantageous for the welds with the experimental wires utilizing lower heat inputs.

Mirror-finish metallographic specimens were prepared and etched for the bead-on-plate welds with Kallings No. 1 reagent (1.5 g CuCl₂, 33 mL H₂O, 33 mL HCL, and 33 mL ethanol) to analyze the microstructure and quantify the martensite and ferrite (α) fractions as well as to measure the microhardness. Microhardness testing on Vickers scale with a load of 100 g and
A dwell time of 10 s was conducted on transverse cross sections taken from all the experimental and reference welds. The measurements were made on the top (close to weld cap), middle, and bottom (close to the weld root) of the weld cross section as illustrated in Fig. 4.

Following the bead-on-plate welds, V-grooved multipass welds were also prepared on 0.5-in.-thick, 7.0-in.-wide, and 10.0-in.-long structural steel plates using the experimental consumables as well as the commercial welding wire. The V-groove had a 70-deg included angle and 0.3-in. depth as shown in Fig. 5. The weld was made in the flat position using the semiautomatic GMAW process. Different transverse, out-of-plane deflections (or angular distortions) were observed for the different LTT welding wires. Figure 6 shows photographs of a multipass weld illustrating the out-of-plane distortion measurement. The cumulative deflection or angular distortion for each weld was measured and compiled with an uncertainty of ±0.008 in. (±0.2 mm). All welds were performed using a semiautomatic welding setup. The cumulative deflection as defined in Fig. 6 for each weld can be calculated as follows:

$$\text{Deflection angle}(\theta) = \sin^{-1}\left(\frac{O-H}{W}\right)$$  \hspace{1cm} (5)

Tensile tests were also conducted to gather more information about the newly developed consumables and to compare how the base metal and weld metals behave differently under uniaxial loading. Groove welds were performed according to the joint preparation illustrated in Fig. 7. Tensile specimens were then extracted and tested in the desired orientation and according to Standard ASTM E8. Tensile specimens were extracted from the SO-200B deposited weld metal and the base metal, ASTM A36 plate, and tested in this work.

From the stress-strain plot in Fig. 8, it is clear that the specimen extracted from the SO-200B weld longitudinally exhibited higher ultimate tensile strength (UTS) and lower ductility when compared to the specimen taken from the base metal. The yield strength and the ultimate tensile strength of the SO-200B deposit were 70 and 95 ksi as compared with 45 and
60 ksi of the base metal. The higher strength of the weld joint was attained from the higher alloy content and the martensitic microstructure of the weld. The strength properties of the newly designed welding consumables will be helpful in subsequent Sysweld analysis. The drop in the ductility of the weld made by the SO-200B experimental wire compared to the base metal was expected due to the formation of martensitic microstructure. The weld specimen with nearly fully martensitic microstructure will enhance the compressive residual stresses around the weld joint by weld expansion. The reduction in ductility will not be critical in the carbon-steel structural steel weld joints, which are not typically in tension or cryogenic conditions.

Results and Discussion

Figures 9A, 10A, 11A, and 12A show the macrostructures of the welds made using the experimental welding wires, SO-200A, SO-200B, SO-350A, and SO-350B, with depth-to-width ratios of 0.40, 0.64, 0.54, and 0.49, respectively. The contact angles, respectively, were 34.0, 48.0, 65.0, and 39.0 deg. Figures 9B, 10B, 11B, and 12B show all the microstructures as fully martensitic at the light microscopic level. The amount of martensite is directly related to the volume expansion of the weld joints and consequently will affect the final compressive residual stress around the weld toes.
The Vickers microhardness readings in the range of 213–262 HV are relatively high as a result of the martensitic transformation. However, the contribution of base metal dilution to the weld hardness, i.e., carbon pickup for the four experimental welding wires, cannot be minimized. The hardness readings are listed in Table 4.

X-ray diffraction analysis was also performed on the four weld specimens, using line scan for bulk analysis and point focus (1 mm²) for spot analysis. As shown in Table 5, XRD results confirmed that all four designated samples were indeed martensitic. Since the carbon content is very low for all LTTW wires, the difference in lattice parameters between ferrite (BCC) and martensite (BCT) is expected to be small. The indexes for the ferrite (BCC) are expected then to be matching the indexes of martensite (BCT). For the BCC structure, diffraction occurs only on planes whose Miller indexes when added together (h+k+l) total to an even number. However, the principal diffracting planes in the FCC crystal structure are those whose Miller indexes are either all even or all odd (zero is considered even). The XRD patterns of the newly developed welded samples are shown in Fig. 13A–D, indicating a BCC structure, which is equivalent to a BCT crystal structure, without any clear indication of retained austenite (FCC) (i.e., the index (3 1 1) was not observed, which is the greatest indication of FCC crystal structure). However, it is possible that the amount of retained austenite is small in these welds and below the resolution of standard X-ray technique.

To further evaluate the presence of austenite in the final weld microstructure, i.e., the degree of completion of austenite decomposition upon cooling of weld joints, additional analysis was performed using electron backscattered diffraction (EBSD). The samples were prepared on a vibratory polisher using 0.05-µm colloidal silica. The polished samples were mounted on a specially designed stage that held the samples at a 70-degree angle relative to the incident electron beam. The EBSD was performed in areas of 50 µm × 50 µm, and two scans were taken for each weld near the center of the weld bead. The scans were performed at a working distance of 20 mm with a step size of 75 nm. The microscope was operated at an accelerating voltage of 20 kV. Again, the difference in lattice parameters between ferrite (BCC) and martensite (BCT) for low carbon concentrations is expected to be small. Thus, the scans were set to index only austenite and martensite based on the microstructure, X-ray diffraction, and Gleeble test results.

Yet again, no retained austenite was observed in the final microstructures of the welds deposited (Table 6). Only a very small amount of retained austenite (0.4%) was found in the SO-200A weld. It is possible that the higher amount of nickel equivalent compared to others was the reason for the presence of retained austenite.

Despite the fact the welds had similar bead morphology and heat input, the welded joints using the different LTT welding consumables exhibited different transverse, out-of-plane distortions with respect to the conventional wire. As shown in Fig. 14A, the expansion due to martensitic transformation becomes larger, meaning the distortion is smaller, as the Ms temperature drops. Indeed the ER70S-3 weld showed greater distortion than the other four LTT welding deposits. In fact, the welds with 200°C Ms temperature showed smaller distortion than those of 350°C. The actual magnitude of the expansion will also depend on the content of the different alloying elements as clearly demonstrated by the different behaviors exhibited by SO-200A and SO-200B, and between SO-350A and SO-350B.

Volumetric expansion in martensite transformation is responsible for the compressive residual stress generated in the weld joint. Similarly, martensite transformation will also affect weld distortion. However, martensite start temperature alone does not fully describe the residual stress and distortion generation. The actual chemical composition of the alloy also plays a significant role in distortion control. As such, even though the SO-200A welding wire exhibits lower martensitic transformation temperature, the SO-350B welding wire has higher Cr content and larger Creq. Taking into account the larger atomic radius of chromium, alloys with higher chromium equivalent promote greater compressive residual stresses around the weld toe. This explains why the welds made with the experimental wires SO-200A and SO-350B are nearly identical in terms of distortion.

Figure 14B shows the amount of reduction in transverse distortions of the welds made by the experimental welding wires compared to the actual distortion observed in the weld made with the commercial wire. It can be readily recognized that the SO-200B weld joint exhibited the least distortion and correspondingly the highest expected compressive residual stress. On the other hand, the experimental alloy with the lowest Creq and highest martensitic transformation temperature, SO-350A, exhibits the maximum distortion among the other welds made by the developed welding wires and correspondingly the least reduction in distortion compared to the actual distortion made by the conventional welding wire ER70S-3.

To assist in future implementation of the experimental electrodes, all welding parameters were collected using an ARC Agent™ 3000P data-acquisition system, and the metal transfer modes were analyzed. Figure 15A, C, E, and G show the voltage signals for the designated wires indicating a number of events with moderate amplitude and frequency. The four figures displaying real-time data show essentially a mixed globular and spray mode for all four wires. Further discrimination between globular and spray mode must rely on other types of analysis. Fast Fourier Transform (FFT) spectra shown in Fig. 15B, D, F, and H further analyze the transfer behavior. Using Fig. 15D as example and assuming that the lower frequency peaks (<30 Hz) are related to weld pool oscillation, the peaks at 70, 140, and 220 Hz stood out more obviously from the background. Thus, it is expected then that the molten droplets transferred across the arc at these frequencies. To further characterize the transfer modes of all welding wires, the
droplet size was calculated and compared to the wire diameter of 0.0625 in. (1.59 mm). Droplet diameters were calculated using the following expression (Ref. 24):

\[ d_d = \left( \frac{\text{WFS} \cdot d_w^2}{40 \cdot f_d} \right)^{\frac{3}{2}} \]  

(6)

Where, \( d_d \) is the droplet size in inches, WFS is wire feed speed in in./min, \( d_w \) is the electrode diameter in inches, and \( f_d \) is the droplet detachment frequency in Hertz. For the 0.0625-in. SO-200A welding wire, the calculated diameters of the SO-200B droplets ranged between 0.0795 and 0.0434 in. (2.02 to 1.10 mm). Therefore, the droplets are both larger and smaller than the wire diameter, indicating mainly globular and spray transfer. This same behavior was observed for the remaining three welding wires, SO-200A, SO-350A, and SO-350B. The SO-200A electrode showed droplets between 0.0830 and 0.0510 in. For the SO350 wires, the droplet diameters ranged between 0.0706 and 0.0402 in. (1.79 and 1.02 mm), and 0.0787–0.0478 in. (2.0 to 1.21 mm). It can also be concluded that some short-circuiting transfer likely occurred during the 200A and 200B welds because of the relatively large voltage fluctuations (at the order of 20 V) observed in the voltage-time plots. These findings also suggest that additional ingredients to the metal cored electrodes may improve the metal transfer behavior and electrode weldability.

Conclusions

Low-transformation-temperature welding (LTTW) wires were designed with the goal of minimizing the distortion due to residual stress developed during cooling of the welded assembly. A comprehensive review of predictive equations for martensite transformation start temperatures, \( M_s \), was conducted. Using the Self and Olson equation, Eichelman and Hull equation, and the methodology proposed by Ghosh and Olson, \( M_s \) temperatures were calculated for a number of alloy compositions that spanned the martensitic region on the Schaeffler constitution diagram. It was observed that the chemical composition-\( M_s \) relationship is not unique, i.e., various chemical compositions were found to produce the same \( M_s \) temperature. This finding also hints that \( M_s \) temperature may not necessarily be the single indicator that characterizes the magnitude of the transformation-induced compressive residual stress around the weld area.

In the welds deposited with the welding wires developed by proper adjustment of the Mn, Cr, and Ni contents, a martensitic microstructure within the fusion zone was achieved. Despite the fact the heat input of the welds using the newly developed welding wires were the same, the welded joints using LTT welding consumables showed smaller transversal, out-of-plane distortions with respect to a conventional wire, ER70S-3, which does not transform to martensite as measured by Payares-Asprino using dilatometry analysis (Ref. 6). This distortion is affected by the \( M_s \) temperature and the content of alloying elements. It is confirmed that use of welding wires containing Cr and Ni in particular contents is effective for reducing the \( M_s \) temperature, increasing the fraction of martensite, and reducing the welding distortion (better controlling residual stress). Further experimental works and modeling analysis will be presented in a separate paper, to discuss in detail the distribution of residual stresses in welded plates and determine the nature and extent of the effect of chemical composition on \( M_s \) and the final residual stresses that develop in the toe region of a weld.

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