Effect of Filler Metal on Mechanical Properties of HSLA Welds

The microstructure and mechanical properties of API 5L-X65 tandem submerged arc welded pipeline steel welds were improved using different combinations of filler metals

BY B. BEIDOKHTI AND R. POURIAMANESH

ABSTRACT

The effect of filler metal composition on the microstructure and mechanical properties of four-wire tandem submerged arc welded API 5L-X65 steel was investigated. Microhardness, Charpy impact, and all-weld longitudinal round tensile tests were applied to evaluate the mechanical properties of the weld metals. The microstructure of specimens was studied by optical and scanning electron microscopes. The microstructural observations showed that the more acicular ferrite was formed in the microstructure, the better mechanical properties were achieved. The impact energy and strength values were higher for the specimen containing more amounts of titanium and boron. The ductile-to-brittle transition temperature could be decreased by increasing the amount of acicular ferrite. For the microstructure with 80% acicular ferrite, the impact energy was about 70 J at –60°C. The addition of titanium dispersed fine oxide inclusions and facilitated the nucleation of acicular ferrite in the microstructure. The oxide particles with a size of 2 µm acted as the nucleation sites of acicular ferrite.

KEYWORDS

• Mechanical Properties • Filler Metal • Tandem Submerged Arc Welding (TSAW) • Microstructure • API 5L-X65 Pipe

Introduction

High-strength low-alloy (HSLA) steels, which have a good combination of strength and toughness, have been widely used in the construction of long-distance oil and gas pipelines (Refs. 1–7). Although formability, fracture toughness, resistance to hydrogen-induced cracking (HIC), stress corrosion cracking (SCC) behavior, sulfide stress cracking (SSC) resistance, and fatigue strength are the workability requirements, weldability is one of the most important necessities for pipeline steels (Refs. 7–11).

The mechanical properties of welds are influenced by chemical composition, cooling rate (related to heat input), and microstructure. Microstructure is particularly a function of the other variables. Chemical composition is the most important factor determining the weld metal (WM) properties (Ref. 12). The electrode composition, type of flux, base metal dilution, and pyrometallurgical chemical reactions in the weld arc could affect the weld metal composition (Refs. 9, 12, 13). In order to improve the WM toughness, two major accesses, including the use of different fluxes (Ref. 9) and the change of filler metals, have been followed (Refs. 7, 8).

It has been shown that a predominant acicular ferrite (AF) microstructure with martensite/austenite (M/A) islands as a second phase showed the highest resistance to cleavage fracture due to the existence of fine interlocking laths of AF phase (Refs. 14–16). Conversely, cracks could propagate easily through the microstructure containing blocky proeutectoid ferrite (Ref. 14).

It is well known that the alloying elements such as Ni, Mo, Cr, Nb, and Ti have a considerable effect on the microstructure (Refs. 7, 13, 17). According to Bhole et al. (Ref. 7), the combined presence of Ni (2.03–2.91 wt-%) and Mo (0.7–0.995 wt-%) resulted in an increase in AF. On the other hand, when Ni was added alone in the range of 2.03–3.75 wt-%, the weld metal showed a lower toughness value due to the reduced amount of AF and the increased volume fraction of ferrite with second phase (FS). In the case of silicon, acicular ferrite could be formed in the microstructure when the Si content was relatively low, i.e., 0.03–0.26 wt-% (Ref. 6). At higher Si contents between 0.42 and 0.95 wt-%, AF was not observed and the microstructure was close to bainite.

It has been reported that boron segregates to the grain boundaries and reduces the boundary energy; as a result, heterogeneous nucleation is prohibited and the amount of allotriomorphic ferrite and bainite is reduced. This phe-
The phenomenon allows AF to form at inclusions, and the hardenability is promoted (Refs. 18–20). Boron has a high tendency to form compounds with O2 and N2; therefore, titanium is usually added to the weld wire to protect the boron. Furthermore, titanium makes boron more available for the grain boundary segregation by entrapping C and N (Ref. 20). This fact is the reason for the development of the Mo-Ti-B wires in the submerged arc welding (SAW) process (Ref. 14). Also, titanium improves the toughness of steel weldments and refines the microstructure by forming stable Ti-rich particles (Refs. 21, 22). Increasing Ti content of the weld metal leads to an increase in the number of inclusions, which act as nuclei for acicular ferrite formation (Ref. 6).

Lee et al. (Ref. 21) investigated the increase in sulfur content from 0.0005 to 0.0124% in Ti-killed steels by the simulation of the heat-affected zone (HAZ) thermal cycle. They reported that by increasing S content to 0.0102%, the amount of AF was increased, and then acicular ferrite was replaced by Widmanstätten ferrite (WF) and polygonal ferrite (PF) (Ref. 21). They explained that the dominant inclusions were changed from Ti-oxides to MnS and Ti-oxysulfides, and this could be the reason for the increased nucle-

![Fig. 1](image1.png)  
**Fig. 1** — A — The schematic view of the welding pass sequence; B — TAW configuration of electrodes; C — welding on the inside.

![Fig. 2](image2.png)  
**Fig. 2** — Optical micrographs of OD region. A — Specimen A; B — specimen B; C — specimen C; D — specimen D.

<table>
<thead>
<tr>
<th>Alloying Elements (wt-%)</th>
<th>Phase Distribution</th>
<th>Mechanical Properties (+ or −)</th>
<th>Material (WM)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni</td>
<td>Mo</td>
<td>Mn</td>
<td>Si</td>
</tr>
<tr>
<td>−</td>
<td>0.81–0.88</td>
<td>−</td>
<td>−</td>
</tr>
<tr>
<td>2.03–2.91</td>
<td>0.7–0.99</td>
<td>−</td>
<td>−</td>
</tr>
<tr>
<td>2.03–3.75</td>
<td>−</td>
<td>1.55–1.59</td>
<td>−</td>
</tr>
<tr>
<td>More than 1.02</td>
<td>−</td>
<td>0.25</td>
<td>1.50</td>
</tr>
<tr>
<td>More than 2.25</td>
<td>−</td>
<td>Up to 0.20</td>
<td>1.40</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.84</td>
<td>0.42</td>
<td>−</td>
<td>−</td>
</tr>
<tr>
<td>−</td>
<td>−</td>
<td>0.03–0.26</td>
<td>−</td>
</tr>
<tr>
<td>−</td>
<td>−</td>
<td>0.42–0.95</td>
<td>−</td>
</tr>
</tbody>
</table>

![Table 1](image3.png)  
**Table 1** — Effect of Alloying Elements on Phase Distribution and Mechanical Properties of Steel Weldments
Table 2 — Different Combination of Wires During Welding

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Electrode No. * (Inner Passes)</th>
<th>Electrode No. (Outer Passes)</th>
<th>Flux**</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>S2Mo</td>
<td>S2Mo</td>
<td>S2Mo</td>
</tr>
<tr>
<td>B</td>
<td>S2Mo</td>
<td>S2Mo</td>
<td>S2Si</td>
</tr>
<tr>
<td>C</td>
<td>S2Mo</td>
<td>S2Mo</td>
<td>S2Mo</td>
</tr>
<tr>
<td>D</td>
<td>S2Mo</td>
<td>S2Mo</td>
<td>S2Mo TB</td>
</tr>
</tbody>
</table>

*EN 756, **EN 760
as described in Ref. 29 was used to determine the characteristics of M/A constituents and carbides. In the latter, M/A constituents are embossed and carbides are depressed in a matrix of ferrite. Therefore, M/A constituents can easily be identified by SEM. A Clemex Imagine Analysis System was used for the quantitative phase evaluation. The mode of fracture, and type of microconstituents and inclusions were determined by Tescan Vega II XMU SEM linked to a Rontec EDS system.

**Results**

The optical and SEM micrographs, as well as the volume fraction of different phases, are shown in Figs. 2 and 3, and Table 4. It can be seen from the corresponding micrographs that a mixed microstructure of polygonal ferrite (PF), acicular ferrite (AF), and ferrite with aligned second phase (FSP) was obtained in each specimen.

Table 4 shows the percentage of distributed phases in the weldments. It was observed that specimen D had the maximum amount of acicular ferrite (80%) and the minimum percentage of polygonal ferrite among the others. This fact could be the reason for the best mechanical properties of this specimen as will be discussed following.

During the two-pass TSW process, the HAZ produced by the first SAW pass was partly reheated by the second pass. As a result, the intercritically reheated coarse-grained HAZ (ICCG-HAZ) was produced in these joints. In this study, the ICCGHAZ of all specimens showed similar microstructural characteristics. The microstructure

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**Table 3 — Chemical Composition of Wires, Base Metal, and Weld Metals (wt-%)**

<table>
<thead>
<tr>
<th>Material</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P**</th>
<th>S**</th>
<th>Mo</th>
<th>Cu</th>
<th>Al**</th>
<th>V**</th>
<th>Ti**</th>
<th>Nb**</th>
<th>Pcm</th>
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</thead>
<tbody>
<tr>
<td>Base Metal</td>
<td>0.10</td>
<td>0.40</td>
<td>1.45</td>
<td>150</td>
<td>30</td>
<td>0.20</td>
<td>0.35</td>
<td>500</td>
<td>800</td>
<td>120</td>
<td>600</td>
<td>0.170</td>
</tr>
<tr>
<td>SG3*</td>
<td>0.10</td>
<td>0.94</td>
<td>1.50</td>
<td>40</td>
<td>90</td>
<td>0.01</td>
<td>0.06</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>S2Mo</td>
<td>0.10</td>
<td>0.13</td>
<td>1.03</td>
<td>120</td>
<td>60</td>
<td>0.46</td>
<td>0.04</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>S2Si</td>
<td>0.07</td>
<td>0.20</td>
<td>0.92</td>
<td>100</td>
<td>80</td>
<td>0.02</td>
<td>0.11</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>S2MoTB</td>
<td>0.09</td>
<td>0.28</td>
<td>1.23</td>
<td>100</td>
<td>50</td>
<td>0.45</td>
<td>0.09</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
<tr>
<td>Weld Metal A</td>
<td>0.06</td>
<td>0.22</td>
<td>1.36</td>
<td>153</td>
<td>86</td>
<td>0.20</td>
<td>0.02</td>
<td>133</td>
<td>250</td>
<td>57</td>
<td>226</td>
<td>0.163</td>
</tr>
<tr>
<td>Weld Metal B</td>
<td>0.06</td>
<td>0.29</td>
<td>1.38</td>
<td>173</td>
<td>91</td>
<td>0.13</td>
<td>0.04</td>
<td>136</td>
<td>237</td>
<td>50</td>
<td>220</td>
<td>0.156</td>
</tr>
<tr>
<td>Weld Metal C</td>
<td>0.05</td>
<td>0.26</td>
<td>1.30</td>
<td>153</td>
<td>73</td>
<td>0.14</td>
<td>0.02</td>
<td>210</td>
<td>283</td>
<td>83</td>
<td>283</td>
<td>0.147</td>
</tr>
<tr>
<td>Weld Metal D</td>
<td>0.07</td>
<td>0.20</td>
<td>1.44</td>
<td>183</td>
<td>93</td>
<td>0.13</td>
<td>0.04</td>
<td>120</td>
<td>247</td>
<td>206</td>
<td>227</td>
<td>0.165</td>
</tr>
</tbody>
</table>

* The SG3 filler metal was used for tack welds by GMAW.
** The values are in ppm.
was a combination of ferrite and small amounts of M/A, as shown in Fig. 4.

Table 4 shows the mechanical properties of WMs. The longitudinal yield strength (YS) and ultimate tensile strength (UTS) were in the range of 534–595 MPa and 606–649 MPa, respectively. Among all specimens, the highest values of yield strength, ultimate strength, and hardness were obtained in the case of weldment D. The hardness in different regions of the base metal, HAZ, inside diameter (ID), and outside diameter (OD) were in the range of 180–206, 170–208, 206–245, and 190–214 HV, respectively. The Vickers microhardness does not vary significantly in all specimens.

Figure 5 shows the macroscale fracture surface of Charpy impact test samples at ~30°C, which consisted of four regions: the fracture initiation region close to the notch (marked as A), the main fracture region in the center of the sample (B), the shear lip regions toward the center (C), and the final fracture region at the end of the fracture surface (D). It could be seen that samples A, B, and C were split into two separate pieces after the Charpy test, whereas sample D was not. This fact could be the sign of the high impact energy of sample D.

Discussion

According to Figs. 2 and 3, it is clear the microstructures were dominated by different morphologies of ferrite. The AF matrix is described by small effective grain size with high angle boundaries, high density of dislocations, and chaotic structure (Refs. 12, 30, 31). This unique microstructural characteristic has prevented crack propagation (Refs. 32, 33). Weld metals A and D presented more than 60% acicular ferrite, while specimens B and C showed large amounts of PF in their microstructures (Table 4). The addition of titanium in specimen D promoted the formation of acicular ferrite, and this phase was the major constituent of the microstructure. Although molybdenum could form nucleation sites for AF (Ref. 17), by comparing specimens A and D it seems that the effect of titanium is more important.

Weld metal D was fabricated using filler metal S2Mo TB. In this filler material, boron is used to suppress the formation of grain boundary ferrite and ferrite side plates (Refs. 19, 22); also, it is liable to be oxidized and nitrified. Therefore, titanium is added to the weld to protect boron. Previous studies have shown the addition of Ti increased the nucleation of AF, and beyond the optimum range of Ti (0.02–0.05%), the recovery of Mn was increased (Ref. 8). For
both actions of titanium, i.e., the protection of boron and the nucleation of acicular ferrite, there is a need for more titanium addition to a weld metal. However, it has been reported that the level of Ti should be well controlled because the unsuitable addition will produce coarse TiN particles that induce cleavage fracture (Ref. 20).

Polygonal ferrite was typically polygonal or plate-like in shape, which was nucleated at either austenite grain boundaries or intragranular regions. It has been reported that PF phase was deleterious to toughness because of its little resistance to cleavage crack propagation (Ref.s 12, 30). Ferrite with aligned second phase was formed on the grain boundaries and had a lath-like morphology. The microstructures like FSP or bainite with the parallel plates and low angle boundaries showed low impact toughness (Ref. 12). Referring to Table 3, carbon was almost constant in all specimens. The Mn contents remained in the range of 1.30–0.44% for all WMs. Manganese is one of the most common alloying elements in steels and acts as a carbide former and austenite stabilizer. Also, it has been found that silicon could slightly retard the bainite formation (Ref. 34). In this investigation, Si content of WMs reached as high as 0.29% for specimen B. According to the Ellingham diagram, the tendencies of Si and Ti to react with oxygen are nearly alike. Therefore, the competition between formation of SiO$_2$ and TiO$_2$ could reduce the number of TiO$_2$ particles. Titanium-based inclusions were favorable sites for the nucleation of acicular ferrite (Refs. 2, 6, 8), and this reduction in the amount of nucleation sites affected the microstructure of specimens B and C. Therefore, the amounts of PF and FSP were increased in the microstructure of welds B and C.

Different reactions could take place in the weld pool and form various inclusions in the weld metal. Some inclusions could facilitate the intragranular nucleation of AF and pin austenite grain boundaries, resulting in a reduced grain size and increased strength and toughness (Refs. 12, 13). It has been reported that 0.002 wt-% Ti in WM did not yield significant amount of Ti in the inclusions (Ref. 32). The chemical composition of inclusions in each specimen were examined using EDS. The results showed that the major constituent elements were Fe, O, C, Ti, Si, Al, Mn, and Mo. Among these elements, titanium was more useful for the formation of AF. Weld metal D showed approximately 0.0206% Ti and the highest percentage of acicular ferrite in the microstructure. The nature of inclusions was complex and multiphase, and it was difficult to provide their complete characterizations through EDS analysis without more supporting analytical studies.

Figure 6 shows the SEM-EDS analysis of one inclusion with a size of 2 μm. It has been reported (Refs. 33, 35) that the appropriate size of nonmetallic Ti-containing inclusions for AF nucleation is between 0.2 and 3 μm. When the size of nonmetallic inclusion was less than 0.2 μm, the decrease in free energy barrier for the nucleation of AF and development of solute-depleted zones within the austenite grain was insufficient (Ref. 33). EDS analyses revealed these inclusions were complex oxides, which mainly consisted of Ti, Fe, Mn, Si, and O. It was also found the inclusions absorbed manganese. It seems the precipitation of MnS at titanium oxides might result in the depletion of manganese, which in turn promotes the nucleation of AF.
Figure 7 shows the SEM micrographs of WMWs that are etched through electroetching. The ferrite matrix, M/A islands, and carbides (if they exist) could be observed in light-black-, white-, and dark-black-colored parts, respectively. The size of M/A islands in weld metal D was the smallest among all specimens and showed the average size of about 2 μm. This value for specimens A, B, and C was approximately 3, 4, and 3 μm, respectively. The partitioning of carbon during growth of acicular ferrite could enrich the carbon content of austenite. This austenite transforms to martensite on cooling. It has been declared that the size and distribution of M/A in the matrix could affect the fracture process at low temperatures; the larger the size of M/A, the smaller the load to initiate new crack nuclei (Refs. 4, 36).

It has been reported that the microstructure of the HAZ is affected by both weld thermal cycle and chemical composition of the steel (Ref. 37). It has been observed that the addition of Ti could control the grain size of austenite in the coarse-grained HAZ and prevent the decrease in toughness (Ref. 38). In this work, different filler metal combinations did not change the microstructural characterization of the HAZ and the predominant microstructure was a combination of ferrite and M/A constituents in all specimens — Fig. 4. The Vickers hardness of any regions of the HAZ was in the range of 185–205 HV10. This indicated the HAZ was not significantly influenced by the addition of Ti. The fraction of M/A phase in the HAZ was smaller than the critical value (i.e., 6%) proposed by Zhu et al. (Ref. 38) for impairing the HAZ toughness.

According to Table 5, although the hardness range of ID was the highest, the hardness values were lower than the acceptance criterion (248 HV) for these pipelines in accordance with NACE MR0175. Therefore, the risk of cracking in sour service applications could be reduced. The highest hardness was related to specimen D and no hard spot was found in any specimens. The fracture surface of the Charpy impact samples at –30°C revealed that the fracture mode of samples A, B, and C were quasi-cleavage, cleavage, and quasi-cleavage, respectively; while in the case of sample D, the fracture surface showed a dimpled appearance — Fig. 8. Also, decreasing the temperature extended the brittle area in all fractured surfaces. Sample D was more ductile than samples A, B, and C in all applied temperatures, i.e., –30°C, –45°C, –60°C, and –75°C. The increased amount of AF in the microstructure of specimen D could be responsible for this behavior. It has been reported (Ref. 31) that high angle grain boundaries and fine packet size as well as the interweave AF microstructure showed good impact toughness and finer cleavage facet compared to the coarse grain boundary allotriomorph ferrite and Widmanstatten ferrite.

The absorbed Charpy impact energies have been plotted against testing temperatures — Fig. 9. At –30°C, the lowest and highest impact energies were 34 J (specimen B) and 164 J (specimen D). At –45°C, the impact energy of specimens B and C was dropped from 34 and 59 J to 24 and 26 J, respectively. Since the brittle-ductile transition (BDT) temperature is calculated based on the impact energy of 28 J (Ref. 39), the horizontal line in the diagram in Fig. 9 represents this criterion. Therefore, the temperature of about –45°C could be considered as the BDT temperature of weldments B and C. By decreasing the temperature to –60°C, the impact energy of specimen A was dropped to 17 J; whereas, for specimen D, the impact energy had the remarkable value of 70 J. It was observed that the best impact properties were obtained by specimen D.

According to the mechanical properties results (Table 5), the yield strength, tensile strength, hardness, and impact energy were well improved in specimen D. This behavior is related to the increase of AF in the microstructure as well as the grain refinement phenomenon due to the more dispersed Ti-containing inclusions. The microstructure containing acicular ferrite has a good Charpy impact toughness, which is mainly related to small effective packet size and fine interlocking pattern of AF. The experimental results revealed that the more acicular ferrite was found in the microstructure, the higher mechanical properties were achieved. By using Ti-B wires, the amount of PF was reduced partially and replaced by AF. This fact was attributed to the increased number of AF nucleation sites (Ti-containing inclusions) and the pinning effect of B on the grain boundaries. From the pipeline fabricator point of view, the good mechanical properties of the welds compensate for the higher cost of Ti-B filler metals.

Conclusions

The effect of filler material on the microstructure and mechanical properties of weld metal was studied in the practical TIG welding process. The main results of the present study could be summarized as follows:

1. Although a mixed microstructure of PF, AF, FSP, and M/A islands was obtained in each WM, acicular ferrite was the dominant microstructure. The microstructure of the HAZ in all specimens was a combination of ferrite, bainite, and martensite/austenite (M/A) constituents.

2. The major constituent elements of inclusions were Fe, O, C, Ti, Si, Al, Mn, and Mo. The increased amount of titanium promoted the formation of AF and improved the mechanical properties.

3. The formation of oxide particles with a size of 2 μm increased the nucleation of AF. Also, the microstructure refinement was encouraged.

4. The increased amount of AF promoted strength, toughness, and impact properties of the weldments. For the microstructure with 80% acicular ferrite, the impact energy was about 70 J at –60°C.

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