Effect of welding speed on friction stir welds of GL E36 shipbuilding steel

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A B S T R A C T
The aim of this study was to characterize mechanical and microstructural characteristics of friction-stir-welded GL E36 shipbuilding steel. The tool rotational speed was kept constant at 500 rpm and different welding speeds were used (1, 2 and 3 mm/s) to achieve different heat inputs. Thermal cycles were monitored by thermocouples placed near the weld face. The welded joints showed a very good esthetics and homogeneous surface quality that indicate a stability of the considered process parameters. The welded joints properties were analyzed by metallographical and mechanical tests such as microhardness, tensile and bending. Macrostructural observations were done at the beginning, middle and ending of the welded length. In addition, radiographic inspection was carried out. The pcBN tool exhibited good wear behavior even after welding around 8.5 m where no apparent loss in dimensions and geometrical features of the probe and shoulder were found. The macrographs displayed different microstructural features and material flow pattern among the heat inputs achieved. A large microstructure gradient was observed, especially within the stirred zone. All the tensile samples broke at the base material showing that the joints achieved higher strength. Microhardness peaks of about 400 HV were also encountered in all the joints. Finally, for welding speeds of 2–3 mm/s the thermocouples presented the most uniform thermal profiles.

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1. Introduction
Friction stir welding (FSW) has expanded rapidly since its development in 1991 and has found applications in a wide variety of industries, including aerospace, automotive, railway, and maritime [1]. The process had a great success when...
applied to aluminum alloys, and this fact was a driving force to stimulated exploration of its applicability to other materials such as steel. Steels represent by far the greatest opportunity for any new process given their undisputed prominence in structural applications due to the fact that the material is strong, versatile, cost effective and reliable [2]. Review publications for FSW are available, Gibson et al. [1], Mishra et al. [3], Nandan et al. [4], Çam [5], Rai et al. [6] and Padhy et al. [7] made comprehensive surveys for FSW technique. Moreover, Pradeep [8], Venkatesh Kannan et al. [9] and Liu et al. [10] published specific reviews concerning FSW applied on steels.

Cui et al. welded the following carbon steels: IF steel, S12C, S20C, S35C, SS0C under various welding conditions, and the mechanical properties and the microstructure were evaluated [11]. Ghosh et al. reported results for plain carbon steel (0.44% carbon content) under variable rotational and welding speeds [12]. Matsushita et al. performed FSW with single-sided one-pass butt welding using 12-mm thick structural steel grade plates with 400N mm⁻² of tensile strength. The microstructural and mechanical properties such as hardness, tensile strength, and Charpy V-notch toughness were investigated [13]. In addition, a special issue of Science and Technology of Welding and Joining (Vol. 14, No. 3) is available only with papers regarding FSW applied on carbon and stainless steels.

The potential for friction stir welding of steel is high, but up to now only for niche applications since for most applications, traditional arc welding processes are well established and cost effective. The FSW process is being actively developed for various shipbuilding activities (most of these are military) and for other critical applications such as welding of high strength pipelines, where conventional arc welding consumables are not adequately developed. Furthermore, the potential for lower distortion has also attracted interest from shipyards was reported by Konkol et al. [14]. In addition to these advantages, the problem of hydrogen cracking is not expected to be encountered in FSW due to the elimination of the normal source of hydrogen in fusion welding [3] which translates into very low diffusible hydrogen levels as those measured even for FSW welds performed underwater [15].

Research focused on the application of FSW on shipbuilding steel is scarce; nevertheless, studies concerning HSLA grades are more often available [16–21]. Reynolds et al. produced joints of approximately 470 mm in length on DH36 grade with 6.4 thickness. They reported single pass, full penetration at welding speeds as high as 7.6 mm/s substantially outmatched. Bainite and martensite were found in the nugget region [22]. Cater et al. made a comparison between friction stir and submerged arc welding to join DH36 and E36 grades. An assessment of the distortion for the two welding techniques was reported and an initial comparative data on weld tensile strength, toughness and fatigue was provided. Based on their results, friction stir welding was shown to outperform submerged arc welding [23]. Azevedo et al. published fatigue results for GL-A36 steel; 4 mm plates thick were joined using different process parameters and tools. Their experimental results showed that the joints exhibited a fatigue behavior similar to the base material [24]. Tompkin et al. proposed a process envelope for the grade DH36 varying rotational and welding speeds for 6 mm plates. They evaluated the joints by
tensile, charpy and microhardness tests as well as microstructure characterization [25].

According to this scenario, the present study intends to fulfill the lack of technical and scientific knowledge concerning friction stir welding application to GL E36 shipbuilding steel and provide, therefore, mechanical and microstructure features of the welded joints as a function of welding speed.

### Table 1 - Chemical composition of the shipbuilding steel used.

<table>
<thead>
<tr>
<th>Grade</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>Cu</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>Al</th>
<th>N</th>
<th>Nb</th>
</tr>
</thead>
<tbody>
<tr>
<td>GL E36</td>
<td>0.170</td>
<td>1.400</td>
<td>0.390</td>
<td>0.013</td>
<td>0.030</td>
<td>0.020</td>
<td>0.060</td>
<td>0.006</td>
<td>0.027</td>
<td>0.007</td>
<td>0.025</td>
</tr>
</tbody>
</table>

### 2. Methods

The base material used for this study was the shipbuilding steel grade GL E36 supplied with a normalized heat treatment condition. Plates of 1200 mm³ × 200 mm² × 6 mm were processed according to the rolling direction using butt joint configuration. The chemical composition given by the supplier is presented in Table 1.

Fig. 1 shows the microstructure of this material; it consists of very refined ferrite and pearlite with banded structure in the rolling direction with 180 vickers hardness. The mechanical properties informed by the supplier inspection certificate are: 436 MPa for yield strength, 564 MPa for tensile strength, elongation of 24% and toughness of 87J at −40°C. The welds were processed at Helmholtz-Zentrum Geesthacht using a friction stir welding portal equipment using a force control process. A threaded polycrystalline cubic boron nitride (pcBN) tool from MegaStir™ with W-Re binder was used, commercially supplied with the specification Q70. The tool angle was tilted to 1.5°.

Before the process the plates were cleaned with sandpaper to remove the oxide layer from the surface and additionally cleaned with ethanol. Argon was used as shielding gas to protect both the tool and the weld area from surface oxidation. The dimensions of the tool consisted of shoulder diameter of

![Fig. 1 - Base material microstructure. Etched with Nital 3%, scale bar 100 μm.](image-url)
36.8 mm and probe length of 6 mm. The welding parameters employed were a constant rotational speed of 500 rpm with 1 mm/s, 2 mm/s and 3 mm/s welding speeds to achieve three different heat input values. The process parameters have been selected based on the available literature [1,3–7] and on the know-how from Helmholtz Zentrum Geesthacht.

The heat input (HI) index was calculated according to Eq. (1):

$$HI = \frac{\omega \cdot r}{v}$$  \hspace{1cm} (1)

where $\omega$ represents the rotational speed (radians per second), $r$ represents the torque supplied from the machine (Nm) and $v$ the welding speed (m/s). During the process, the tool temperature was measured with an infrared pyrometer pointed at the face of the locking collar (Fig. 2). For the thermal cycles experimented for the plates during the welding, 18 thermocouples (K type) were placed on the plates as shown in Fig. 3.

The location of the samples removed for bending (BD1, BD2 and BD3), tensile (TT1, TT2, TT3, TT4 and TT5) and microstructure and microhardness (M1, M2 and M3) investigations are shown in Fig. 4. In case of the bending, microstructure and microhardness samples, the numbers 1, 2 and 3 stand for the beginning, the middle and the ending of the joint, respectively. X-ray inspection was done for the whole length of the joint.

Fig. 2 – Thermal acquisition by the pyrometer pointed at the locking collar of the tool.

Fig. 3 – Thermocouple positions used for thermal data acquisition.

Fig. 4 – Location of samples used for characterization of the joints, values in mm.
The tensile tests were performed using a Zwick Roell universal testing machine with a 100 kN load cell performed at 1 mm/min at room temperature according to the standard DIN EN 895:1995-08 for the joints specimens and for the base material DIN 50125:2009-07. The bending tests have been carried out according to the standard DIN EN ISO 5173:2010-08, using HIDROALFA press model 20. The microhardness profiles were measured by a Zwick Roell indentec ZHV 2 system using HVO.5 and 0.4 mm distance between indentations. Additionally, microhardness mapping measurements were performed using a hardness scanner model UT 100 supplied by BAQ. The macrograph and micrograph samples were prepared using different grades of sandpaper with the final polishing done using 1 μm diamond suspension. The specimens were etched for micrograph evaluation with the reagents listed in Table 2. Nevertheless, for the macrograph inspection only Nital was used.

### 3. Results and discussion

#### 3.1. Process parameters relationship and joint quality assessment

Table 3 shows the welding parameters used to process the joints as well as the torque, the axial force and the heat input calculated for each process condition. The torque generated among the three different conditions increased only slightly with the increase in welding speed and this was expected to happen according to the literature [4]. The weld face of the joints displayed very good esthetic and homogeneous surface quality without virtually any flash after the welding procedure, as illustrated in Fig. 5. Flash is produced by material displacement from the face of friction stir welded components. It is often used as a visual sign that the proper tool depth has been used. For example, if the insertion depth is too deep, excessive flash is created. Excessive flash may also result from improper tooling or parameter settings. Shoulder scrolling and reduced rotation rate are examples of flash mitigation techniques [1]. Axial force has an important role on the quality of the weld as well. Very high pressures lead to overheating and thinning of the joint while very low pressures lead to insufficient heating and voids [4].

The weld face quality along the whole length is a very good indicative that the forces involved during the process were stable and the absence of flash after the process proved that the parameters used were suitable to maintain the stirred material within the joint, with an exception of the joint processed with 3 mm/s of welding speed because in this case the bending test showed a small defect at the root of the weld.

The results from the X-ray inspection did not show any macro defects along the whole length of the three different weld conditions processed. In this sense, Fig. 6 shows the images obtained at the middle of the joints that exemplify these tests. Usually, radiographic test can detect broken tool pieces and defects such as voids and wormholes. The technique easily identifies volumetric defects larger than 0.5 mm in diameter; however, this test has some difficulty with non-

![Fig. 5](image)

**Surface appearance of the joint processed with 1 mm/s, region close to the exit hole.**

![Fig. 6](image)

**X-ray images of the joints processed with (a) 1 mm/s, (b) 2 mm/s and (c) 3 mm/s as welding speed.**

### Table 2 – Etchants used to reveal the microstructure.

<table>
<thead>
<tr>
<th>Etchant</th>
<th>Composition</th>
</tr>
</thead>
<tbody>
<tr>
<td>Nital</td>
<td>Ethanol, nitric acid</td>
</tr>
<tr>
<td>Beraha I</td>
<td>1 g potassic sulfite, 100 ml parent dilution</td>
</tr>
<tr>
<td></td>
<td>Beraha I (1000 ml distilled water, 200 ml</td>
</tr>
<tr>
<td></td>
<td>hydrochloric acid 32%, 24 g ammonium hydrogen</td>
</tr>
<tr>
<td></td>
<td>difluoride</td>
</tr>
<tr>
<td>Nital + Na₂S₂O₅</td>
<td>Pre-etching with Nital, wet etching with</td>
</tr>
<tr>
<td></td>
<td>Na₂S₂O₅ 10% in aqueous dilution</td>
</tr>
<tr>
<td>LePera</td>
<td>50 ml Na₂S₂O₅ 1% in aqueous dilution,</td>
</tr>
<tr>
<td></td>
<td>50 ml picric acid 4% in ethanol</td>
</tr>
</tbody>
</table>

### Table 3 – Process parameters utilized to produce the joints.

<table>
<thead>
<tr>
<th>Condition</th>
<th>Rotational speed (rpm)</th>
<th>Welding speed (mm/s)</th>
<th>Torque (Nm)</th>
<th>Axial force (kN)</th>
<th>Heat input (kJ/mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>500</td>
<td>1</td>
<td>59</td>
<td>40</td>
<td>3.089</td>
</tr>
<tr>
<td>2</td>
<td>500</td>
<td>2</td>
<td>66</td>
<td>40</td>
<td>1.728</td>
</tr>
<tr>
<td>3</td>
<td>500</td>
<td>3</td>
<td>75</td>
<td>40</td>
<td>1.309</td>
</tr>
</tbody>
</table>
volumetric defects (particularly those oriented in the plane parallel to the surface) [1].

### 3.2. Thermal profiles and heat input approach

Wei and Nelson have studied four different process heat indexes and reported that the heat input provides the best correlation with post-friction stir weld microstructures [17]. Nevertheless, it is important to keep in mind that this equation only gives an estimate of the heat input into the joint, since it does not take into account tool heat losses through radiation or by conduction [18]. Furthermore, it was reported that the maximum temperature reached for FSW on steels is less than 1200 °C and the time Δt, taken to cool over the range 800–500 °C is approximately 11 s. This is comparable to, for example, Δt for a manual metal arc weld with a heat input of about 1.3 kJ/mm. Therefore, the metallurgical transformations expected on the basis of cooling rates alone are not expected to be remarkably different from ordinary welds [4,26]. The insufficient heat input accompanied by the lack in material flow may be caused in a stagnant zone formation as proved by modeling studies [27]. The material in this region will be vulnerable to crack formation under the normal plunge force [28]. And lower heat inputs are associated with higher welding speeds if kept the rotational speed.

The temperatures measured with the pyrometer pointed to the side face of the locking collar showed peak temperatures of 850 °C, 800 °C and 760 °C during the process respectively for the condition processed with 1, 2 and 3 mm/s of welding speed. These values confirmed that keeping the same rotational speed with the increase of the welding speed the tool is going to reach a lower peak temperature because it stays less time in contact at the same heated position on the plate during the process, as illustrated in Fig. 7. Additionally, the temperature measurements acquired on the plates with the thermocouples, whose were located near to the face of the joints, proved the similar phenomenon, i.e., the increase of the welding speed decreases the peak temperature undergone by the workpiece as it is shown in Fig. 8.

Cho et al. have reported based on the experimental and modeling results at 304L stainless steel that temperatures were higher on the advancing side than on the retreating side around 100 K (173 °C) [29]. Arbegast and Hartley have reported that there was a slightly higher temperature on the advancing side of the joint because the tangential velocity vector direction was the same as the forward velocity vector [30]. This observation has been confirmed for almost all the thermocouples measurements at the present study by the values from Fig. 8. The large temperature difference for the 304L alloy is higher than the values typically observed in the FSW of other materials under most welding conditions. However, the asymmetry in temperature can be explained due to the very low thermal conductivity of stainless steel compared to most other alloys [4]. The heat input index approach should consider the thickness of the plates being welded because for different plate thickness the peak temperatures and the cooling rates are expected to be different because there will be different bulk material to dissipate the thermal cycle supplied by the friction heat of the process. Therefore, the same heat input value adopted for different plate thickness will result in different macrograph features and different joint microstructures characteristics resulting in distinct joint performance.

### 3.3. Mechanical and microstructural features

Considering that X-ray inspection has a limitation to detect discontinuities smaller than 0.5 mm in diameter, simple bending tests are a useful and fast way to evaluate more precisely the existence of smaller defects not detected by the radiograph technique. Fig. 9 presents the bent samples and their respective angles. The sample conditions processed with 1 and 2 mm/s have been bent completely proving absence of defects within the joint. However, during the bend test, the condition processed with 3 mm/s of welding speed, the sample BD2 (middle of the joint) showed a lack of penetration (LOP); Fig. 10 shows this defect at the joint root at the bent specimen and the defect detail at the metallographic specimen.

A root bend test, for instance, is a good predictor of fatigue test performance related to root flaws. The failure in root bending is accompanied by a measurable and generally significant reduction of fatigue life [1]. Poor bonding of the joint can often be attributed to inadequate pressure on the workpiece material due to a lack of axial force placed onto the surface of the workpieces owning inadequate axial load [14]. Stevenson et al. [31] also reported that as weld root defects resulted from the lack in plunge depth of the FSW tool. Therefore, a recommendation to solve this issue and to be able to produce sound welds using higher welding speeds should be to increase the axial load to make sure that the material stirring will be bond properly, i.e., without defects that could compromise the performance of the joint. Rameshi et al. reported the tendency to form macroscopic defects with the increase in welding speed and that root flaw and groove defect were observed at higher welding speeds [32]. Moussawi et al. reported results from DH36 and EH46 steel grades FSW joints correlates defects with high tool welding speeds such voids, weld root defect and kissing bonds. It was found that the lack in material flow as a result of stagnant zone formation was the main reason of these defects [28].

Fig. 11 exhibited tensile results for the joints and the base material mechanical properties were taken from data given by the supplier. In general, friction stir welds of steels displayed higher mechanical properties than the respective base material [5]. This is explained because of the thermo mechanical effect imposed by the welding process on the base material.

![Fig. 7 – Pyrometer tool measurements.](image-url)
Fig. 8 – Thermal profiles measured by the thermocouples, retreating side (RS) and advancing side (AS).

Fig. 9 – Bent samples, heat input and the measured angle for each welding speed condition.
The increase of the welding speed, i.e., decrease of the heat input would result in a smaller heat affected zone and narrow stirred zone. The macrographs results are useful for quality inspection and an initial guidance for microstructure evaluation, as well as for better understanding of the mechanical performance of the joints.

Microhardness profiles were made at the beginning, the middle and ending of the joints, but since these results were quite similar, and the macrographs, the profile at the beginning of the joint are representative (Fig. 13). These results exhibit that the microhardness profiles increase from the base material until the stirred zone due the gradual microstructure transformation of the original microstructure toward the direction of the stirred zone. The perlite and ferrite from the base material were gradually modified by the FSW process first to a degenerated perlite, then to a spheroidized microstructure (heat affected zone) and finally to a complex mixed microstructure consisting mainly of ferrite, martensite and bainite within the stirred zone. The values measured within the stirred zone ranged by 300–400 HV at the middle thickness of the joint, with slightly higher values at the retreating side.

The color maps exhibited in Fig. 14 offers a better way to evaluate the microhardness than the single lines profiles because they demonstrate the hardness overview behavior. However, these results should be interpreted carefully, since they are qualitative and should be used as a reference because the machine takes a calibration measure from the base material of the sample. According to this calibration, the machine

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**Fig. 10** – (a) Top view from the root of the specimen BD2 joined with 3 mm/s showing the LOP (b) cross section image taken at the bottom of the specimen near to the bending sample; the morphology is similar to a forging lap; scale bar is 50 µm.

**Fig. 11** – Tensile testing results.

**Fig. 12** – Macrographs of the welded joints.
makes all the individual microhardness measurements and builds the overview map. For a precise measurement a standard test machine should be used. These results revealed lower values at the bottom and higher values at the retreating side. These facts were explained because close to the root of the joint a higher amount of ferrite was found and at the retreating side a higher amount of martensite and bainite was found. The first observation leads to the assumption that cooling rate within this region was lower than at the top of the joint. The second remark could be explained since the retreating side achieved a lower peak temperature which leads to a higher cooling rate.

Cui et al. reported that as the peak temperature decreases, the cooling rate increases (with the increasing welding speed) and these values are independent of the types of steels [11]. Besides, the argon used as a shielding gas for the tool might play a role as a forced cooling source leading to the formation of harder microstructures at the top of the joint. The maps also demonstrated that with the increase of the welding speed the microhardness behavior showed less gradient between the advancing and the retreating sides and more homogenous hardness distribution, especially the condition processed with 2 mm/s of welding speed. The microhardness color maps demonstrated a suitable correlation with the macrographs specimens, therefore being appropriate to realize the location of the hardness gradients.

The etching trials made, using the solutions listed in Table 2 for the microstructure characterization, proved that Nital is the suitable choice since it revealed the microstructure more clearly and easily. Beraha I and Nital + Na₂S₂O₅ did not show results better than Nital, and LePera presented heterogeneous, unclear and not reliable results. The mixed, highly deformed and refined microstructure presented at the stirred zone of the samples probably contributed for the unsuccessful other etching trials, since these reagents should reveal such microstructures properly [34].

Fig. 15 indicates the microstructures of the center at the middle thickness of the stirred zones of the joints. They consisted of a complex mixed refined microstructure of ferrite (highlighted as ‘F’), martensite (highlighted as ‘M’) and bainite (highlighted as ‘B’). As higher peak temperatures are achieved from lower welding speeds, they will result in larger austenite grains. This grain size then influences the hardenability of the material and thus the phase type and morphology that will form during the cooling [18]. Since a way to evaluate the weldability for steels is the carbon equivalent, this value calculated according to the International Institute for Welding (IIW) for the base material joined in the current work was 0.48, by itself this index predicts that the grade has high tendency to form martensite. A strong correlation for hardness as a function of the IIW carbon-equivalent, according to the Eq. (2) has been reported, and the highest hardness values reported correspond to microstructures which contain substantial quantities of martensite and bainite in carbon-containing steels [4].

\[
CE = C + \frac{\text{Mn} + \text{Si}}{6} + \frac{\text{Ni} + \text{Cu}}{15} + \frac{\text{Cr} + \text{Mo} + \text{V}}{5} 
\]

(2)

The condition welded with the lower heat input of this study, i.e., processed with the higher welding speed of 3 mm/s reached the lowest peak temperatures and hence had a higher cooling rate leading to a higher amount of displacive transformation such as bainite, Widmanstätten ferrite and acicular
ferrite instead of diffusional transformations such as allotriomorphic ferrite and pearlite. The microstructures identified at the stirred zone of the welds produced on the present study showed consistency with the evaluation of Reynolds et al. [22] when they applied the FSW process on the grade DH36 with 6.4 mm in thickness, as well as the results reported by Troupis et al. [25] also for the DH36 alloy with 6 mm thick and with the work published by Barnes et al. for the grade HSLA-65 with 6.35 thickness [18]. Rameshi et al. found the microstructure of the stirred zone consisted of upper bainite and fine ferrite phases and that the amount of upper bainite reduced with welding speed due to decrease in peak temperature and faster cooling [32].

Imam et al. reported that the volume fraction of martensite and average hardness in the stirred zone decrease with the increasing welding speed and that the peak temperatures decreased and cooling rates increase with the increasing welding speed [35]. Troupis and his coworkers reported that slow welding speed welds, 100 mm/min and 200 rpm, presented a ferrite predominant, homogeneous microstructure with significant grain refinement; a heterogeneous microstructure was exhibited by the weld at the intermediate welding speeds, 250 mm/min and 300 rpm, consisted of acicular shaped bainitic ferrite rich regions and ferrite predominant regions of either acicular shape or of random geometry and the fast welding speed welds, 500 mm/min and 700 rpm, features to be a heterogeneous but predominantly acicular shaped bainitic ferrite microstructure with small regions of acicular ferrite [36].

Notwithstanding, it is very important to keep in mind that the prior base material microstructure features, i.e., grain size and heat treatment supply condition, are going to play an important role concerning all the metallurgical transformations induced by the thermo and mechanical process.

3.4. Tool performance

Fig. 16 demonstrates the condition of the tool before any welds and after approximately 8.5 m of welds. The tool visual evaluation showed good wear behavior without apparent loss in dimensions and geometrical features of the probe and shoulder. This is an indicative that the process parameters selected were sufficiently appropriate to reach the necessary heat and plasticize the material right in front of the tool during its linear motion. Consequently, ensuring that the forces and the torque experienced by the tool were suitable to minimize its wear.

The literature claims that tool wear was found to increase with rotational speed and decrease at lower traverse speed, which suggests that process parameters can be adjusted to increase tool life [6]. Furthermore, it was reported that due to the low fracture toughness of the pcBN tool, this material is susceptible to thermal cycling cracking as a result of differences of heating and cooling rates between the pin and shoulder [18], Despite this, pcBN tool is still preferred FSW of steels and other high melting temperature alloys due to its high strength and hardness at elevated temperatures along
with high temperature stability. Furthermore, the low coefficient of friction for pcBN results in smooth weld surface [6].

4. Conclusions

The effect of welding speed on friction stir welds of GL E36 shipbuilding steel was evaluated in the present study. The findings of the current study can be summarized as follows:

- The process parameters were appropriately chosen since all the joints showed very good esthetics and homogeneous surface quality. Nevertheless, when used 3 mm/s as welding speed it was found lack of penetration, but this could be bypass increasing the axial force. Concerning the tool, it did not exhibit considerable wear even after 8.5 m of weld length.
- The macrographs at the beginning, middle and ending of the welded joints had similar microstructure features proving the process stability, especially between the joints produced using 2 and 3 mm/s as welding speed.
- The X-ray inspection has limitation to detect smaller defects, so it was mandatory to additionally perform bending tests to evaluate the joint integrity.
- For the welds processed with welding speeds of 2–3 mm/s, the temperature measures showed most uniform thermal profiles. Hence, the microstructure and mechanical properties obtained with these two conditions are more uniform along the whole joint length.
- A high microstructure gradient was found within the stirred zone. The microstructure found consisted mainly of ferrite, martensite and bainite with different levels of refinement and morphologies showing that joints were obtained with a balance between strength and toughness.

Conflicts of interest

The authors declare no conflicts of interest.

Acknowledgments

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