Study of cracking susceptibility in similar and dissimilar welds between carbon steel and austenitic stainless steel through finger test and FE numerical model

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Abstract
Hot cracking susceptibility and the formation of brittle martensite phase are the main factors that limit the weldability of a dissimilar joint between carbon steel (CS) and austenitic stainless steel (SS). In this study, the self-constraint finger test was used to correlate the welding thermo-mechanical field with the crack susceptibility of a dissimilar weld between the CS ASTM A36 and SS AISI 304L. The finger test allowed to intercalate fingers (portions) of tested materials in the weld samples to produce dissimilar welds. The heat dissipation and the distortion behavior were related to the crack susceptibility, critical weld regions extension, and chemical species diffusion. Four samples were welded (two similar welds and two dissimilar welds) using the filler metals ER70S-6 and EC410NiMo. Welds were analyzed through light optical microscopy (LOM) and scanning electron microscopy (SEM) to characterize phases, detect cracks, microstructural changes, and element diffusion. A finite element (FE) numerical model was applied to simulate the welding thermo-mechanical field. FE estimations of distortion and residual stress helped to predict induced crack propagation (the initial gap between fingers) towards the fusion zone. Additionally, electrochemical tests were carried out to assess the corrosion susceptibility of the dissimilar welds. The observed cracks were produced due to different factors such as residual stress distribution, the formation of brittle and untempered martensitic phase in the fusion zone (FZ), and hot cracking associated with the weld sample distortion behavior. According to the FE estimations, the high thermal expansion of the SS was responsible for the bending curvature change in welds 2 and 4, which produced a gap between fingers and increased the crack extension in the FZ of weld 4. The dilution contributed to the formation of δ-ferrite in the FZ, which limited the growth of cold and hot cracks. The decarburization and sensitization were not observed in dissimilar welds due to the low element diffusion.

Keywords Dissimilar metal welding, GMAW, FEM, Cracking susceptibility, Finger test, Corrosion

1 Introduction
The dissimilar weld joint between structural carbon steel (CS) and austenitic stainless steel (SS) is of great interest due to its application in the construction of nuclear reactors [1], petrochemical industry [2] and piping systems [3]. It is known that CS and austenitic SS have excellent weldability with any welding process, which is associated with the percentage carbon equivalent and nickel/chromium equivalent relation [4].

However, the dissimilar welded joint between CS and austenitic SS has several drawbacks including martensitic transformation [2, 5], decarburization of CS [5], and sensitization of austenitic SS generated by the formation of chromium carbides [1, 3].

Additionally, the cold and hot cracking affect the weldability of CS and austenitic SS, respectively. Some researchers have pointed out that some causes of hot cracking in the dissimilar weld joint between CS and austenitic SS are the fully austenitic solidification mode [2, 5] and the brittle martensite formation [2, 6]. Further metallurgical changes in the dissimilar weld joint between CS and austenitic SS are associated with the inevitable element diffusion phenomenon (C, Fe, Ni y Cr) thermally activated [3, 5].
For instance, Lee [3] detected cracking along the grain boundary at the FZ in a dissimilar weld joint between CS (0.3%wt C) and 316L austenitic SS. The cracking occurred due to the segregation band near the grain boundaries at the FZ. Furthermore, some cavities and micro-cracks were formed during the solidification process caused by the variation of tensile stresses (cooling stage) [3]. Ul-Hamid et al. [6] observed cracks in the dissimilar weld joint between ASTM A53 steel and SS 304 obtained through the gas tungsten arc welding (GTAW) process. Cracks were found in a carbon-rich segregated layer of the ASTM A53 steel, where the martensite was formed.

Therefore, several investigations have focused on studying alternatives that improve the weldability of the dissimilar weld joint between CS and austenitic SS. A widely accepted solution for controlling hot cracking and metallurgical changes in dissimilar weld joints is the proper selection of welding parameters and filler metals [2, 7]. Khalifeh et al. [7] used four SS filler metals to weld S37 and AISI 304L steels through the GTAW process. The welded joints obtained with the different filler metals were sound. However, some filler metals increased the amount of δ-ferrite in the FZ, which has a detrimental effect on mechanical strength and corrosion resistance. Arivazhagan et al. [1] produced weld joints between AISI 4140 and AISI 304 steels using three different welding processes such as GTAW, electron beam welding (EBW) and friction resistance welding (FRW). The three welding processes allowed to obtain sound welds. The instantaneous heat input improves the element diffusion (Cr, Ni, Fe, C). Thus, the LBW process exhibited the most significant diffusion effect [1].

Additionally, selecting a suitable filler metal for a dissimilar weld joint (CS-austenitic SS) is crucial to affect the corrosion resistance. Previous researches have indicated that the FZ corrosion resistance is higher than the base metal (BM) when the filler metal has a high content of Cr and Ni. This fact has been confirmed with different welding processes such as gas metal arc welding (GMAW) [8], shield metal arc welding (SMAW), and GTAW processes [9].

However, the effect of the welding thermo-mechanical field on the cracking susceptibility of the dissimilar weld joint between CS and austenitic SS has not been studied. The thermo-mechanical field represents the temperature and stress distributions produced by the welding process. These distributions strongly depend on the thermal conductivity and thermal expansion coefficient of parent materials. Further studies have reported that CS works as a heat sink, which does not allow austenitic SS to reach phase transformation temperatures (δ-ferrite) [2, 5], affecting the solidification mode [3].

The correlation between the thermal conductivity and thermal expansion coefficient with the cracking susceptibility is important to find causes of cracking in the dissimilar weld joint, which can be i) martensitic brittle phase [5] and solidification mode [3, 10] ii) the combined effect of distortion and hard precipitates (carbides) or iii) the propagation of micro-pores / micro-cracks of the base metal.

In practice, hot cracking susceptibility can be measured through self-restraint and externally loaded hot cracking tests [11]. The self-restraint tests are the Y-test, Houldcroft cracking test, and keyhole slotted-plate-test [11, 12]. Whereas the programmed deformation rate test, hot-deformation-rate (HDR), varestraint (VT), and transvarestraint (TVT) are classified as hot cracking externally loaded tests [13–15]. VT and TVT are widely used [16, 17] because they offer a way for weldability quantification by determining the accumulated total hot crack length or the maximum hot crack length to identify the solidification cracking temperature range [11, 17]. It is important to note that, regardless of the selected test type, self or externally loaded, the basic principle to evaluate the hot cracking susceptibility is the mechanical or thermomechanical deformation of the region near the weld pool [11].

A self-constraint test that can help to correlate the thermo-mechanical field and cracking susceptibility is the finger test. The finger test evaluates hot cracking susceptibility based on the weld bead width percentage containing cracks [18]. An advantage of the finger test is that it does not require special preparation to add a filler metal like the VT [19]. Since the finger test specimen is formed with independent portions (fingers), it is possible to intercalate the two materials (austenitic CS and SS) to measure their response to the welding thermal cycle. Hence, it is possible to associate the heat accumulation and the welding distortion distribution with the thermal conductivity and thermal expansion coefficient of each material.

In the present research work, the finger test was used to measure the effect of the welding thermo-mechanical field on the cracking susceptibility of dissimilar weld joints between ASTM A36 and AISI 304L steels. The fingers were welded using the GMAW process with EC410NiMo and ER70S-6 filler metals.

In addition, the study applied a thermo-mechanical finite element (FE) model to simulate the thermal cycle and weld distortion distributions. The correlation between the metallographic analysis and the FE numerical model allowed to determine the causes that produced the cracking in each weld. The element diffusion was measured through scanning electron microscopy/energy dispersive X-ray spectroscopy (SEM-EDS). Finally, the corrosion resistance in dissimilar weld joints was measured through electrochemical tests. Subsequently, the corrosion results were correlated with the element diffusion and metallurgical changes observed in the dissimilar joints.

2 Experimental procedure

The finger test evaluated the cracking susceptibility of similar and dissimilar welds between CS A36 and austenitic SS AISI
The chemical composition of those steels is given in Table 1. The chemical composition of the filler metals was according to AWS A5.9 and A5.18 standards. In addition, the chemical composition of SS304L and CS-A36 steel was determined through the spark atomic emission spectrometer (AES) SpectroMax®. The spectrometry analysis has a duration time of 5 s using a voltage of 400 V and a frequency of 400 HZ.

Each finger test specimen was formed with six portions (fingers) 20 mm long, 50 mm wide and 5 mm thick [20]. Four specimens were welded, two of these were similar welds of CS A36 and SS AISI 304L. In the last two specimens were carried out dissimilar welds between CS A36 and SS AISI 304L. Dissimilar weld samples used an alternate arrangement between CS A36 and SS 304L fingers, as shown in Fig. 1. The filler metals Infra® ER70S-6 and Böhler® EC410NiMo were used in similar and dissimilar welds as summarized in the experiment matrix of Table 2. The chemical composition of the filler metals is also indicated in Table 1.

Welding experiments were conducted using the semi-automatic GMAW process. The Binzel Alpha-3® torch was mounted on a tractor to keep the stick-out (3mm) and welding speed constants (Fig. 1). Previously, Cuiuri [21] informed that a voltage range between 18V and 25V stabilizes short-circuit transfer mode in the GMAW process. Furthermore, Norrish et al. [22] claimed that applying the short-circuit transfer mode in the GMAW process depends on the equipment and wire feed rate to control spatters and heat input. Deo [23] determined that the maximum wire speed was 148 mm/s in the GMAW process. In this study, all experiments used the short-circuit transfer mode with 21V and, feed and wire speeds of 3.18 mm/s and 84.6 mm/s, respectively.

Additionally, argon gas was used as a shielding media at a 20 ft³/hr flow rate. As shown in Fig. 1, the vise provided a constant torque of 108.5 kN·mm. This torque was optimal to keep the fingers of the weld samples perfectly aligned.

The temperature history was measured through two K-type thermocouples mounted on the weld sample rear face. Thermocouples were located at 3 mm (thermocouple 1) and 6 mm (thermocouple 2) from the weld seam center (transverse direction) and a 3 mm depth. In welds 1 and 2, thermocouples were mounted only on one side of the third finger. Meanwhile, in welds 3 and 4, thermocouples were settled in the third and fourth fingers. The thermal history was recorded in real-time during the welding process using a data acquisition module (DAQ) TC-08 model and a computer.

Before the welding process was performed, the longitudinal distortion was measured through a dial indicator mounted on the tractor to check the perfect alignment of the specimen fingers after the torque application (Fig. 1). Once the GMAW process was completed, the longitudinal distortion was measured again in each specimen. The distortion measurements were done once the weld sample reached the thermal equilibrium.

The microstructural analysis of the base metals (CS A36 and SS 304L) and weld samples was carried out using light optical microscopy (LOM). The metallographic preparation consisted of two stages, mechanical ground and polishing with 0.3 μm alumina. Subsequently, CS A36 was etched with a 3% Nital solution for 15s (base metal). An electro etching with oxalic acid at 10V DC for 30 s was applied to reveal the SS 304L microstructure (base metal). Meanwhile, prior to the metallographic preparation of weld samples, the reinforcement generated by the GMAW process was removed. Then, a selective etching technique was employed to reveal the microstructure of weld samples. Times of etching and solutions used in the selective etching technique are indicated in Table 3. Finally, the SEM-energy-dispersive spectrometry (EDS) technique was used to characterize the elemental composition of the critical weld regions FZ and HAZ. Images were acquired in a resolution of 1 eV per step, 100 ms per step and 150 eV pass energy through the SEM JEOL JSM-IT300 operated at 20 kV at high vacuum conditions.

The microhardness was measured in BM and weld samples at the central intersection of all samples (Fig. 1b). Microhardness measurements were performed using a load of 200 gf with a dwell time of 10s.

| CS A36 | C | Mn | P | S | Si |
|--------|---|----|---|---|----|
| 0.28   | 0.75 | 0.04 | 0.05 | 0.4 |

| SS 304L | C | Mn | Cr | Ni | Mo | Ti |
|---------|---|----|----|----|----|----|
| 0.0345  | 1.096 | 18.29 | 8.32 | 0.043 | 0.0304 |

| EC410NiMo | C | Mn | Cr | Ni | Mo |
|-----------|---|----|----|----|----|
| ≤0.025   | 0.7 | 0.9 | 12 | 4.6 | 0.6 |

| ER70S-6 | C | Mn | Cr | Ni | Mo |
|---------|---|----|----|----|----|
| 0.10    | 0.97 | 1.62 | 0.15 | 0.15 | 0.15 |
2.1 Electrochemical measurement of corrosion resistance

Corrosion tests were performed by means of an electrochemical cell with a standard configuration of three electrodes: i) a platinum auxiliary electrode (AE), ii) saturated calomel electrode (SCE) as a reference electrode (RE), and iii) dissimilar weld sample as work electrode. During the electrochemical tests, a potential range of −1500 to 1500 mV was applied to open circuit potential (OCP). Corrosion tests used a scanning-sweep rate of 0.5 mV/s. Electrochemical corrosion parameters, such as current density (I_{corr}), anodic (b_a) and

| Zone                        | Reagent                      | Time | Zone                        | Reagent                      | Time |
|-----------------------------|------------------------------|------|-----------------------------|------------------------------|------|
| BM and HAZ (CS A36)         | Nital 3%                     | 15s  | BM and HAZ (SS 304L)       | Aqua Regia                   | 30s  |
|                             | 98 ml ethanol                |      |                             | 60ml HNO_3                   |      |
|                             | 3ml HNO_3                    |      |                             | 40ml water                   |      |
| FZ                          | Nital 3%                     | 15s  | FZ                          | Lichtenegger-Bloech          | 60s  |
|                             | 98ml ethanol                 |      |                             | 100ml distilled water        |      |
|                             | 3ml HNO_3                    |      |                             | 20g NH_4F·HF                 |      |
|                             |                              |      |                             | 0.5g K_2S_2O_5               |      |
|                             | Ferric chloride              | 60s  |                             | Ferric chloride              | 60s  |
|                             | 100 ml ethanol               |      |                             | 100 ml ethanol               |      |
|                             | 5g FeCl_3                    |      |                             | 5g FeCl_3                    |      |
|                             | 3ml HCl                      |      |                             | 3ml HCl                      |      |
| HAZ (SS 304L)               | Carpenter                    | 12s  | HAZ (SS 304L)               | Carpenter                    | 10s  |
|                             | 8.5g FeCl_3                  |      |                             | 8.5g FeCl_3                  |      |
|                             | 2.4g CuCl_2                  |      |                             | 2.4g CuCl_2                  |      |
|                             | 122ml HCl                    |      |                             | 122ml HCl                    |      |
|                             | 6ml HNO_3                    |      |                             | 6ml HNO_3                    |      |
|                             | 122ml ethanol                |      |                             | 122ml ethanol                |      |
| BM (SS 304L)                | Aqua Regia                   | 30s  | BM and HAZ (CS A36)         | Nital 3%                     | 20s  |
|                             | 60ml HNO_3                   |      |                             | 98 ml ethanol                |      |
|                             | 40ml water                   |      |                             | 3ml HNO_3                    |      |
Table 4  Temperature-dependent thermophysical and mechanical properties of base metals (CS A36 and SS 304L) and filler metal EC410NiMo used in the FE thermo-mechanical model

| Properties                      | Material       |
|--------------------------------|----------------|
| Thermal conductivity           |                |
| Density                        |                |
| Specific heat                  |                |
| Young modulus                  |                |
| Poisson ratio                  |                |
| Thermal expansion coefficient  |                |

A list of the simplifying assumptions for the mathematical model of the welding thermal field is provided below:

1. The FE simulation of the thermal field did not consider the convective flow of the melted metal in the weld pool.
2. Temperature-dependent thermophysical properties such as density (\(\rho\)), thermal conductivity (\(k\)), and specific heat (\(C_p\)) were considered (Table 4) to improve the numerical model accuracy [26].
3. The double ellipsoid model proposed by Goldak et al. [27] was used to simulate the volumetric heat distribution of the electric arc (\(Q_T\)) on the weld pool (Eq. 3). Additionally, the term \(Q_w\) was added to the double ellipsoidal heat source (Eq. 3). The term \(Q_w\) includes the additional heat provided by the molten drops of filler metal as suggested by Lee and Chang et al. [28].

\[
Q_T(x, y, s) = \frac{6\sqrt{3}f_sQ}{abcf_s\pi}\exp\left(-3\left(\frac{x^2}{a^2} + \frac{y^2}{b^2} + \frac{s^2}{c_f^2}\right)\right)
+ Q_w
\]  

where \(Q\) is the energy input (\(Q = \eta VI\)) calculated through the current intensity (\(I\)), voltage (\(V\)) and process efficiency (\(\eta = 80\%\)) [18, 29]. The parameters \(a\) (width), \(b\) (depth), \(c_f\) (front length) and \(c_r\) (rear length) correspond to the dimensions of the front and rear ellipsoids. The weight functions \(f_f\) and \(f_r\) were 0.6 and 1.4, respectively [30]. Also, the non-inertial coordinate \(s\) was applied to determine the heat source position during its travel.

3 FE thermo-mechanical model

A FE thermo-mechanical model was used to estimate the thermal cycle and the distortion of similar and dissimilar welds. The decoupled numerical solution was obtained computationally through the commercial software ANSYS Mechanical® (19.0, 2019).

The temperature distribution in similar and dissimilar welds was estimated by means of the numerical solution of the heat diffusion equation:

\[
\rho C_p \frac{\partial T}{\partial t} = \nabla (k \nabla T) + Q_T
\]
The calculation domain (base metal fingers) was subdivided through the FE mesh shown in Fig. 2. Element sizes transverse to the weld seam were 1 mm (weld metal) and 2 mm (HAZ), as proposed by Heinze et al. [31]. The mesh conformed by 21120 elements (Hex20 and Tet15) and 18365 nodes exhibited the orthogonal quality and the skewness displayed in Fig. 2b–c, respectively.

Additionally, FE models with different mesh densities (refinements) were used to assess the adequacy of element sizes and the effect of mesh refinement in the welding thermal cycle. Table 5 summarizes peak temperatures estimated at 3 mm from the FZ (upper face) for similar welds between CS A36 and SS 304L (Fig. 2a). The temperature variation between the finer mesh (68376) and mesh in Fig. 2a was lower (3.3%), but the computing time increases more than twice (Table 5). Therefore, the FE model (Fig. 2a) without mesh refinement can be worked accurately for this study providing computing cost savings.

Boundary conditions applied to the welding thermal model were the convection \((q_c = h(T - T_\infty))\) and radiation \((q_r = \varepsilon \sigma (T^4 - T_\infty^4))\) environmental heat losses. The properties considered by the boundary conditions are given in Table 6. Meanwhile, the thermal equilibrium with the environment \((T = T_\infty)\) was taken as the initial condition \((t = 0)\).

The estimation of temperature distribution in the weld samples was taken as initial data of the mechanical model to calculate the thermal load. The welding mechanical field considered the numerical solution of equilibrium (Eq. 4) and total deformation (Eq. 5) equations.

\[
\varepsilon^{\text{total}} = \varepsilon^T + \varepsilon^e + \varepsilon^P + \varepsilon^C
\]

where \(\sigma_{ij}\) corresponds to the thermal stresses in the weldment, the body force is the product \(\rho b_i\). In Eq. 5, the total strain \(\varepsilon^{\text{total}}\) is calculated as the sum of strains: i) thermal \((\varepsilon^T)\), ii) elastic \((\varepsilon^e)\), and iii) plastic [34].

The assumptions made were the following:

1. The creep strain was not considered due to the higher temperatures (ranging between 700 °C and 1100 °C) only was experienced by the weld sample in a short-period [35].
2. Temperature-dependent mechanical properties such as Young modulus \((E)\), Poisson ratio \((\nu)\), and thermal expansion coefficient \((\alpha)\) were used in the FE mechanical model (Table 4).
3. The bilinear hardening model was applied to estimate the welding plastic strain.
4. It was considered that base materials follow the Von Mises yield criterion as reported in [36, 37].

The FE mesh shown in Fig. 2a was also used to solve the mechanical field numerically. The same FE mesh used in thermal and mechanical models allowed the transfer of the thermal model data (temperature distribution) quickly. Boundary conditions were the constraint caused by the vise jaws and the compressive force \((F)\) provided by the torque \((T = F \times L)\) as indicated in Fig. 1a. The computational model was solved in a workstation with Intel Core i7-6500U 3.1 GHz 16GB RAM processor.

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Fig. 2  a) FE mesh used by the thermo-mechanical simulation, b) orthogonal quality (average =0.97), c) Skewness (average =0.013)
4 Results and discussion

4.1 Metallographic characterization

4.1.1 BM

Figure 3a–b shows the microstructure of CS A36, which is formed by a ferritic matrix (equiaxed grains) with pearlite islands. Perlite shows lightly banded regions of different sizes due to the hot forming process. A similar microstructure was observed by Khalifeh et al. [7] on the St37 steel with a ferritic matrix and pearlite islands. The percentage of pearlitic phase measured in CS A36 was 20.7%.

The BM microstructure of the SS 304L is shown in Fig. 3c. The presence of δ-ferrite as bands in the rolling direction on the austenitic matrix was observed. (Fig. 3c). The percentage of δ-ferrite measured in the SS 304L was 3.2% (BM condition).

4.2 Critical weld regions

The Schaeffler diagram was used as a reference to predict the phases and possible weld defects in weld samples 2, 3 and 4. (Table 2). Figure 4 shows the compositions of base materials (CS A36 and SS 304L) and filler metals (ER70S-6 and EC410NiMo) plotted on the Schaeffler diagram (Cr eq and Ni eq) considering the dilution rates. The dilution was calculated from Eq. 6. The weld penetration depth achieved in each weld sample was measured through LOM (Table 7).

\[
\%\text{Dilution} = \frac{B}{A+B} \times 100
\]  

where \( A \) and \( B \) correspond to the cross-section area of the reinforcement and the weld penetration depth, respectively [38].

As expected, the weld joint of SS 304L with EC410NiMo filler metal (weld 2, Table 2) is in a region free of hot cracking (HC line, Fig. 4) due to the formation of δ-ferrite (approximately 10%, Fig. 4). However, the average percentage of δ-ferrite was 22.5%, which was experimentally measured. δ-ferrite breaks the interdendritic movement of the low melting point liquid film by retarding the wetting of the grain boundaries and thus the formation of hot cracks [4]. Similarly, there were no cracks in the weld joint of CS A36 with filler metal ER70S-6 (weld 1). The CS A36 does not exhibit hot cracking susceptibility. Instead, cold cracking affects the weldability of CS. Cold cracking susceptibility decreases at low carbon equivalent (CE) [39, 40]. In this case, the CE of CS A36 was 0.37 and its weldability was good. (CE<0.4 [40]). The CE was calculated from the standard formula (Eq. 7) proposed by the International Institute of Welding (IIW). Figure 5a–b shows longitudinal section of the similar weld joints of CS A36 and SS 304L, respectively. Both welds were free of cracks.

\[
CE = \frac{\%C + \frac{\%Mn}{6} + \left(\frac{\%Cr + \%V + \%Mo}{5}\right) + \left(\frac{\%Ni + \%Cu}{15}\right)}{}
\]  

Table 5 Mesh-independence analysis performed on the FE thermal model

| Mesh element size (mm) | Number of elements | Number of nodes | Peak temperature estimated in point A (Fig. 2a) (°C) | Computing time (s) |
|------------------------|--------------------|----------------|------------------------------------------------------|-------------------|
| Weld metal HAZ         |                    |                |                                                      |                   |
| 0.25                   | 0.5                | 68376          | 60509                                               | 795               | 6500 |
| 0.5                    | 1                  | 39072          | 34035                                               | 789               | 3700 |
| 1                      | 2                  | 21120          | 18365                                               | 768               | 2800 |
| 2                      | 3                  | 15840          | 13538                                               | 718               | 2100 |
| 2.5                    | 4                  | 11616          | 9928                                                | 688               | 1500 |

Table 6 Properties considered in the boundary conditions applied to the FE thermal model

| Parameter                          | Magnitude       |
|------------------------------------|-----------------|
| Convection coefficient (W/m² K) [32]| \( h = -9 \times 10^{-6}T^2 + 0.39T + 8.1 \) |
| Thermal emissivity (-)             |                 |
| CS A36 [30]                        | 0.9             |
| SS 304L [33]                       | 0.2             |
| Stefan-Boltzmann constant (W/m² K⁴) [33]| \( 5.67 \times 10^{-8} \) |
| Environment temperature (°C)       | 25              |
Conversely, both dissimilar joints exhibited cracks and micro-cracks, as shown in Fig. 5c–d. In dissimilar welds, four regions corresponding to the FZ and HAZ formed in each BM were distinguished (Fig. 5c–d). According to the Schaeffler diagram prediction, the dissimilar weld joints were out of the hot cracking region (HC line, Fig. 4). Moreover, the FZ of weld 3 (Table 2) was predicted fully martensitic (point B, Fig. 4). In contrast, a dual phase transformation (martensite + δ-ferrite) was estimated in the FZ of weld 4 (point C, Fig. 4).

On the other hand, dissimilar welds are within a region cracking susceptible (line AM, Fig. 4). This cracking susceptibility is due to the formation of the tough and brittle martensitic phase during the welding cooling stage. Previously, Ul-Hamid et al. [6] found trans-granular cracks in a dissimilar weld joint of CS and SS. The cracks originated at the interface of the materials and propagated towards the CS. It is believed that these cracks could have produced in the martensitic phase formed in the FZ, which resulted in a high hardness region C-enriched.

Figure 6 displays the microstructure observed in welds 1 and 2. In the FZ of weld 1 was observed the presence of allotriomorphic ferrite (AF) with adjacent growth of Widmanstatten ferrite (WF) (Fig. 6a). The orientation of the grains was in the heat source direction. While in the HAZ was detected the presence of perlite surrounded by AF and WF, these phases were formed at the grain boundary in the coarse (CGRZ) and fine (FGRZ) grain recrystallization zones, respectively (Fig. 6b). Meanwhile, the formation of vermicular δ-ferrite and thick lathy ferrite regions on the martensitic matrix were detected in the FZ of weld 2 (Fig. 6c). Finally, vermicular δ-ferrite was observed in the HAZ of weld 2. Note that the austenitic grain growth in the HAZ was negligible (Fig. 6d).

Figure 7a shows the FZ and HAZ of weld 3 (Table 2). The martensitic phase was observed in the FZ of the SS 304L side (Fig. 7b), as predicted by the Schaeffler diagram (Fig. 4). The block-shape martensite (Fig. 7b) was formed due to the enrichment of Cr and C elements and the high welding cooling.

### Table 7

| Weld joint | Penetration depth (mm) | Weld bead width (mm) | Reinforcement (mm) | % dilution |
|------------|------------------------|----------------------|-------------------|-----------|
| Weld 1     | 2.09                   | 9.14                 | 2.59              | 33.80     |
| Weld 2     | 2.25                   | 10.5                 | 2.62              | 44.43     |
| Weld 3     | 2.01                   | 8.54                 | 2.57              | 50.26     |
| Weld 4     | 1.64                   | 8.21                 | 2.47              | 32.09     |
rate. Traces of vermicular δ-ferrite were found in the FZ-HAZ interface (Fig. 7c). δ-ferrite is formed at high temperature (>1150°C) and is mainly retained at the grain boundaries [7]. Also, a δ-ferrite free zone was distinguished in the HAZ (Fig. 7a). It is important to mention that grain growth was not observed in the HAZ of the 304L SS side, which indicates that this region had a negligible extension.

In the FZ of the CS A36 side (weld 3), WF was formed. This phase grew through the formation of the AF, as shown in Fig. 7e. In low carbon steels, AF is the primary phase formed at austenite grain boundaries during cooling below A3 [41], which was corroborated in the FZ of weld 3 (Fig. 7e). WF grew through the AF in the HAZ of the CS A36 due to relatively low undercooling. (Fig. 7e). Previously, Jafarzadegan et al. [42] pointed out that the WF was not the result of a displacing transformation but is formed by a para-equilibrium mechanism, which involves the rapid diffusion of interstitial carbon atoms across the boundary movement towards the remaining austenite during the shear transformation [42]. The growth of AF and WF promoted the enrichment
of the remaining austenite, which led to the formation of ferrite and pearlite phases (Fig. 7e) [42].

Figure 8a shows the EDS line scanning performed between the HAZ and the FZ in weld 3 (SS 304L side). The Cr diffusion occurs from the SS 304L to the FZ (ER70S-6 filler metal), as indicated in Fig. 8b. Figure 8c–d showed that Ni and C contents remain unchanged, i.e., there was no diffusion of these elements. Although C has a higher diffusion coefficient than Cr [5], the low C content of both SS 304L and the ER70S-6 filler metal (Table 1) did not produce any considerable concentration gradient. The Cr diffusion was facilitated due to the high temperature reached by the SS 304L fingers during the welding thermal cycle. It is worth mention that the SEM-EDS technique is not accurate enough to measure C
contents lower than 1%. Therefore, C contents are mainly qualitative results to indicate its variation between the HAZ and FZ.

Also, an EDS line scanning was performed between the HAZ (CS A36 side) and the FZ (ER70S-6 filler metal), as shown in Fig. 9a. The Cr diffusion occurs in the FZ of the CS A36 side (Fig. 9b). The Cr found in the FZ was diffused from the HAZ of the SS 304L side. There was no Cr diffusion from the FZ towards CS A36 (Fig. 9b–c) nor C diffusion from the HAZ (CS A36 side) to FZ (decarburization). The lower the heat flux in CS A36 fingers, the lower the driving force for diffusion between the FZ and the HAZ.

A martensitic matrix was observed in the FZ (SS 304L side, Fig. 10a). Meanwhile, an austenitic matrix with negligible grain growth and vermicular δ-ferrite bands was observed in the HAZ of the SS 304L side (Fig. 10b). Figure 10c shows the polygonal and vermicular δ-ferrite found in the FZ of the SS 304L side. The selective etching technique allowed to observe δ-ferrite distribution in the FZ (Fig. 10c). In the HAZ of the CS A36 side, a coarse grain recrystallization zone (CGRZ) was observed (Fig. 10d). The presence of AF at the
grain boundaries, WF growing from the AF and acicular ferrite (A_{cF}) growing within the grain were observed in the CGRZ (Fig. 10d). Lathy and skeleton δ-ferrite were detected in the FZ of the CS A36 side (Fig. 10e). These morphologies occurred due to the ferrite remains at the subgrain boundaries during the solidification process. Thus, the final weld microstructure consists of a mixture of martensite and eutectic ferrite. The eutectic ferrite forms via a eutectic reaction at the end of solidification [43]. In Fig. 10f, grain refinement was observed in the inter-critical zone (ICZ) as well as the presence of ferrite and pearlite phases. The sub-critical zone (SCZ) also showed refined grain zones with dispersed pearlite islands on the ferrite matrix (Fig. 10f).

Figures 11, 12 show EDS line scanning performed on both FZ-HAZ interfaces of weld 4 (Fig. 10a and d). At the FZ boundary (Fig. 11a), a Cr and Ni depleted region of approximately 30 μm was generated (Fig. 11b–c). Subsequently, the Cr and Ni contents were recovered inside the FZ (Fig. 11b–c). There was no Cr and Ni diffusion towards CS A36 (Fig. 11b–c). However, a short distance of C diffusion was produced from the HAZ (CS A36) towards the FZ (EC410NiMo), as indicated in Fig. 11d.

In contrast, an enriched Cr and Ni region was generated between the HAZ of the SS 304L side and the FZ (Fig. 12a–c). This region was produced due to the Cr and Ni diffusion from the HAZ (304L SS side) towards the FZ. The element diffusion was driven by the heat flux between the SS 304L fingers and the weld bead. Note that the low C content remained constant in both regions (Fig. 12d).

4.3 FE thermo-mechanical analysis of the weld samples

The numerical solution of the heat diffusion equation (Eq. 2) allowed to estimate the welding thermal cycle. Figure 13
shows the temperature distribution estimated in different instants for the weld samples.

The heat accumulation in the weld metal region and the adjacent zone was higher in welds 2 and 4, as shown in Fig. 13a and d. In weld 4, the higher temperatures (Fig. 13d) promoted the element diffusion (Figs. 11d and 12b–c). Also, the lower thermal conductivity of SS 304L contributed to the heat accumulation in both FZ and HAZ. The higher thermal conductivity of the CS A36 at lower temperatures (Table 4) promoted the heat dissipation from the FZ and HAZ towards the BM. This phenomenon occurred in welds 1 and 3 (Fig. 13a–c).

The heat accumulation was corroborated through the transient peak temperature distribution. Figure 14a–b shows the peak temperatures estimated by the FE model in similar welds (welds 1 and 2). It can be observed that the temperature curve did not exhibit significant variations, the typical temperature increment with the heat source travel highlights. However, the peak temperature curves for dissimilar welds (welds 3 and 4) presented fluctuations (Fig. 14c–d). These fluctuations coincided in both welds with the SS 304L fingers, indicating that the heat accumulates in this material during the cooling stage (Fig. 14c–d).

Figure 15 shows the comparison between experimental thermal history and FE numerical estimations for the same positions in the weld samples. It was obtained a good agreement between experimental results and numerical predictions (Fig. 15). Note that the higher variation in peak temperatures for all weldments was produced nearest to the FZ.

Fig. 11  a) EDS analysis performed on the HAZ (CS A36) and FZ (EC410NiMo) in weld 4 to quantify the distributions of: b) Cr, c) Ni, d) C

Fig. 12  a) EDS analysis performed on the HAZ (304L SS) and FZ (EC410NiMo) in weld 4 to quantify the distributions of: b) Cr, c) Ni, d) C
However, the average variation was low (5.3%), confirming that FE mesh and thermophysical properties temperature-dependent were suitable to improve numerical accuracy.

The heat dissipation in the FZ strongly depends on the filler metal thermal conductivity due to the shallow penetration depth achieved by the GMAW process (Table 6). Hence, the heat accumulation in weld 4 was higher. Besides, the gap (thermocouple 1). However, the average variation was low (5.3%), confirming that FE mesh and thermophysical properties temperature-dependent were suitable to improve numerical accuracy.

Fig. 13 FE estimations of transient temperature distribution for: a) Weld 1, b) Weld 2, c) Weld 3, d) Weld 4

Fig. 14 Peak temperatures estimated by the FE numerical model during the welding thermal cycle for: a) Weld 1, b) Weld 2, c) Weld 3, d) Weld 4
between fingers of the weld samples was negligible due to the force exerted by the vise; this contributed to increasing the heat flow between metals, specifically during the cooling stage (Fig. 13).

As mentioned before, the FE estimations of the temperature distribution were used as a thermal load in the mechanical model to simulate the welding distortion. The comparison between the post-welding longitudinal distortion measured...
experimentally and the FE estimations, once the weld samples reached the thermal equilibrium, are shown in Fig. 16.

The post-welding longitudinal distortion estimations obtained by means of the FE numerical model showed the same behavior as the distortion measured experimentally (Fig. 16). The variation between experimental and FE estimations of distortion was higher in welds 2 and 4 (Fig. 16b and d). This variation was associated with the used filler metal (EC410NiMo in both welds). The expansion/shrinkage underwent by the filler metal (EC410NiMo) changed with the heat dissipation rate, which was affected by the fingers material.

Figures 17, 18 show the FE estimations of residual stress and longitudinal distortion (y-deformation) distributions generated by the GMAW process in welds 1 and 2 during the thermal cycle. In both welds, the heat source travel through the second finger of the weld sample generated a gap between fingers 1 and 3 (Figs. 17b and 18b). This gap was produced due to the thermal expansion experienced by the filler metal and BM. When the heat source arrived at the intersection between fingers 1 and 2, the peak temperature significantly increasing (Fig. 14a–b), which generates a high thermal deformation.

Once the peak temperature was stabilized (Fig. 14a–b), fingers 1-3 of weld 1 presented a convex bending, as displayed in Fig. 17f. Meanwhile, the bending was concave in fingers 4-6 (Fig. 17g). On the other hand, weld 2 exhibited a concave bending in the whole sample (Fig. 18g). The difference in bending between welds 1 and 2 was related to the heat dissipation, the mechanical strength and the thermal expansion behavior (BM and filler metal).

![Fig. 17 FE estimations of residual stress (a–d) and longitudinal distortion (e–g) during the thermal cycle in weld 1](image1)

![Fig. 18 FE estimations of residual stress (a–d) and longitudinal distortion (e–g) during the thermal cycle in weld 2](image2)
The lower temperatures of the thermal cycle estimated in weld 1 (Fig. 13a) are related to the higher thermal conductivity. When the heat source travels through a finger, this undergoes its maximum thermal expansion exerting a compression force on the adjacent fingers. The finger starts to dissipate the heat from the FZ towards the HAZ and BM when the heat source moves away. At the same time, the thermal expansion reduces and the mechanical strength increases. Thus, the constraint imposed by the vise jaws forced the sample to expand towards the opposite side. Similarly, the compression force exerted by the torque (Fig. 1a) also restricts the fingers expansion, which gives rise the convex bending (Fig. 17g).

When the heat source arrives at the last fingers (5-6), the peak temperature and the thermal expansion are high (Fig. 14a), but the mechanical strength is low (Table 4). The longitudinal displacement of the last fingers (5-6) is restricted by the compression force and the adjacent fingers, which have recovered their mechanical strength. Therefore, the last fingers (5-6) undergo a concave bending (Fig. 17g), this was corroborated experimentally, as shown in Fig. 16a.

Conversely, the estimated distortion in weld 2 was higher in the intersection between fingers 4 and 5 (Fig. 18g). In this case, the thermal expansion was high due to the peak temperatures (Fig. 14b). The heat dissipation was slow because the SS 304L thermal conductivity decreases at lower temperatures (Table 4). Thereby, the fingers 1-4 and 5-6 exerted forces in opposite directions generated by the thermal expansion and the jaws constraint. These forces produced the convex bending of the weld sample (Fig. 18g). A little gap was produced between fingers 5 and 6 in weld 2, as shown in Fig. 18a. This gap coincided with the change of the bending curvature (Fig. 18g).

In both similar welds, residual stresses were higher in the region near the weld bead. Ficquet et al. [44] obtained similar results for the residual stress distribution estimated by a FE numerical model in a bead on plate weld. They reported that
tensile stresses were higher in the weld bead. On the contrary, Alorai et al. [45] informed that high residual stresses were measured in the adjacent region to the FZ of a bead on plate weld, but not in the weld bead toe.

The higher tensile residual stresses estimated in the weld bead of welds 1 and 2 (Figs. 17, 18) were related to the finger distortion. This distortion produces bending due to the fingers only were joined in the weld bead region (upper part). Meanwhile, the lower part of the finger remains free to bend. Depending on the peak temperature distribution, the bending can be concave (Fig. 17g) or convex (Fig. 18g). Thus, the residual stress concentrates in the upper part (weld bead).

Figures 19, 20 show the residual stresses and longitudinal distortion distributions estimated during the welding thermal cycle in dissimilar weld joints (welds 3 and 4). In weld 3, residual stresses were higher in the weld bead (Fig. 19b–d) like welds 1 and 2. However, residual stresses were negligible in weld 4 (Fig. 20b–d) because gaps between fingers produced during the thermal cycle (Fig. 20a) contributed to releasing residuals stresses [46].

Weld 3 had a similar distortion behavior to weld 1 (Figs. 17g and 19g) since the CS was predominant (fingers and filler metal). SS 304L fingers (2, 4 and 6) exhibited a higher thermal expansion due to the peak temperatures (Fig. 14e). The deformation of SS 304L fingers generated gaps in the intersections during the thermal cycle, as indicated in Figs. 19b–d. Then, these gaps disappear due to the thermal expansion of SS 304L fingers during the cooling stage (Fig. 19a). The thermal expansion brought about the higher residual stresses were produced in the last finger of the weld sample (Fig. 19d).

In weld 4, the position change of SS 304L fingers (positions 1, 3 and 5) and the filler metal slightly modified the distortion behavior (Fig. 20e–g). Fingers 1-4 exhibited a convex bending, while concave bending occurred in fingers 5 and 6.

The higher heat accumulation and thermal deformation of SS 304L fingers produced a concave bending (Fig. 20b–c). The concave bending generated gaps in all intersections as the heat source traveled through the weld sample (Fig. 20b–c). Residual stresses accumulated in the last finger due to the higher thermal expansion of both SS 304L fingers and the weld bead similar to weld 3 (Fig. 20d). The same deformation behavior was estimated in welds 1, 3 and 4 (Figs. 17, 19, 20). Weld 4 exhibited higher convex and concave bending (Fig. 20e–g).

The FE estimations of the bending for the weld samples were in good agreement with the literature. It is known that bending of the bead on plate welding requires that the inherent shrinkage force exceeds the bending force [47]. Also, disturbance is needed [47]. The transverse deformation of the weld sample induces the disturbance. Figure 21 shows the estimations of the transverse deformation for all weld samples.

### 4.4 Cracking analysis of weld samples

The dissimilar weld between CS and austenitic SS has drawbacks as the martensitic transformation of the CS and the hot

| Weld joint                               | TCL         | MCL         |
|------------------------------------------|-------------|-------------|
| CS A36-SS 304L (filler metal ER70-S), weld 3 | 5713.11 μm  | 4179.38 μm  |
| CS A36-SS 304L (filler metal EC400NiMo), weld 4 | 604.23 μm   | 604.23 μm   |

Table 8 Total and maximum cracking length measured experimentally in welds 3 and 4 by means of LOM
cracking susceptibility of the SS due to the full austenitic (A) solidification mode \cite{2,48}. In this study, some cracks were produced in weld beads of the dissimilar welds (Fig. 5c–d). These cracks were detected in all intersections of both dissimilar welds. Table 8 indicates the total cracking length (TCL) and the maximum cracking length (MCL) measured in each dissimilar weld.

According to the two assessment criteria for hot cracking susceptibility \cite{19}, cracking was more severe in weld 3 (Table 8). Figure 22 shows some cracks observed in welds 3 and 4 by means of LOM and SEM. As can be seen in Fig. 22a–b, the crack with maximum length measured in weld 3 arises from the induced-crack (fingers gap), then this crack follows the FZ-HAZ boundary in the CS A36 side and propagates towards the FZ. Some micro-cracks also were observed in weld 3 (Fig. 23a). Meanwhile, only short-length cracks were found in weld 4 (Fig. 22c–d). The induced-crack in weld 4 propagates towards the FZ (Fig. 22d). At the same time, more cracks were produced in the FZ, as shown in Fig. 22e–d.

![Fig. 22 Cracks and micro-cracks found in weld 3 (a–b) (FZ ferric chloride etched and HAZ CS-A36 nital 3% etched) and weld 4 (c–d) (FZ ferric chloride etched and HAZ CS-A36 nital 3% etched, HAZ SS 304L carpenter etched) through LOM and SEM](image1)

![Fig. 23 a) Elemental mapping of chemical elements performed on cracks and micro-cracks found in the FZ (weld 3) to detect concentration of: b) C, c) Al, d) Mn, e) Cr, f) Ni](image2)
In order to determine the causes that produce cracks and micro-cracks in the dissimilar welds, it was performed an electron probe micro-analysis (EPMA), which was correlated with the FE thermo-mechanical model estimations. Figures 23, 24 show the distribution of chemical elements on cracks and micro-cracks with the largest length found in the dissimilar welds. Chemical elements such as C, Cr, Al, Si, and Mn were mapping in the crack region. As is observed in Figs. 23, 24, there were no elements segregated in the cracks of both dissimilar welds. Figure 23b–e shows punctual concentrations of C and Si in the largest crack found in weld 3. Nevertheless, the C and Si concentrations did not form interesting precipitate particles such as carbides. Instead, some oxides (Al$_2$O$_3$ and SiO$_2$) were found in the crack boundaries of both welds (Figs. 23a, 24a), these oxides were associated to the affinity of certain chemical elements (Si, Al, Mn) with the oxygen.

In weld 3, the FZ did not exhibit the solidification mode A (Fig. 7a). However, the largest cracks (Fig. 22a–b) were detected in the brittle martensitic phase formed in the FZ (Figs. 7 and 22a–b). These cracks were produced in the intersections of the last two fingers, CS A36 (finger 5) and SS 304L (finger 6). Cracks are produced due to residual stress accumulation (Fig. 19d). Also, the untempered martensite brittleness formed in the FZ (Fig. 7b) and the change of convex to concave bending estimated by the FE model in the intersection of fingers 5 and 6 (Fig. 19g) contributed to the crack formation. The concave bending generated a tensile stress state on the lower part of the weld sample. Thus, the tensile residuals stresses were accumulated on the weld bead root due to the finger gap.

![Fig. 24](image)

Fig. 24  a) Elemental mapping of chemical elements performed on cracks and micro-cracks found in the FZ (weld 4) to detect concentration of: b) C, c) Mn, d) Si, e) Cr, f) Ni

![Fig. 25](image)

Fig. 25  Microhardness profiles measured experimentally for a) similar welds (welds 1 and 2, Table 2), b) dissimilar welds (welds 3 and 4, Table 2)
On the other hand, the filler metal EC410NiMo (weld 4) also produced martensite in the FZ. It is known that the martensitic phase promotes the formation of micro-cracks in the FZ of martensitic SS [49]. Therefore, cracks found in weld 4 (Fig. 22c–d) are associated with the cold crack susceptibility of the martensitic phase [50, 51]. Meanwhile, cracks found in the finger intersections of weld 4 (induced cracks) were coincident with the gaps produced during the welding thermal cycle (Fig. 20a). These cracks were produced during the solidification process of the weld metal, i.e., hot cracking. It is worth mention that FE numerical model helped to predict the gaps between fingers. Otherwise, these gaps cannot be observed in the naked eye (Fig. 20a).

In both dissimilar welds, the martensitic phase formed in the FZ was cracking-susceptible. Cracking was induced mainly by residual stress accumulation and distortion behavior. Hot and cold cracks length found near the FZ boundary (weld 4) were shorter than cracks detected in weld 3 (Fig. 5c–d, Table 8). Hot cracks were the shortest (Figs. 7d and 22d–e). δ-ferrite found in the FZ boundary (Fig. 10a–b) can help avoid propagating of both cold and induced cracks due to its ductility. On the contrary, the full brittle martensitic FZ in weld 3 increased hardness and strength. When a crack was produced due to the residual stress accumulation, the tough martensitic phase was fractured easily (Fig. 22a–b). Previously, Kose et al. [49] informed that a certain percentage of retained δ-ferrite in the martensitic matrix of the FZ promoted the formation and propagation of cold cracks in martensitic SS welds. However, this was not the case of the CS-austenitic SS dissimilar welds due to δ-ferrite formed in the weld bead was related to the dilution.

### 4.5 Mechanical characterization of welds

Figure 25a shows the microhardness profiles measured in similar welds (welds 1 and 2, Table 2). In weld 1, the average microhardness measured in the HAZ was 154.77 HV. In the FZ (weld 1), the average microhardness was 161.33 HV. The microhardness profile of weld 1 did not exhibit significant changes between the critical regions (BM, HAZ and FZ) since the same phases were observed in these regions.

In weld 2, the average microhardness measured in the HAZ was 309 HV. In the FZ was measured an average microhardness of 353.28HV associated with the combination of martensite and ferrite phases. The BM was affected thermally to a lesser extent in both similar welds (Fig. 13a–b). The average microhardness was 153 HV for the CS A36 (BM). While the average microhardness of the SS 304L (BM) was 211 HV.

The microhardness profiles of dissimilar welds (welds 3 and 4, Table 2) are shown in Fig. 25b. In weld 3, the average microhardness of the BM-1 (CS A36) was 179 HV. WF needles and perlite increased the average microhardness of the HAZ-1 up to 204 HV (Fig. 20b). Meanwhile, the FZ

### Table 9

| Weld joint | E_{corr} (V) | I_{corr} (A/cm²) | b_a (mV) | b_c (mV) | V_{corr} (mmpy) |
|------------|--------------|-----------------|----------|----------|-----------------|
| Weld 3     | -0.44        | 7.9E-4          | 103.72   | 184.69   | 9.58            |
| Weld 4     | -0.432       | 9.9E-4          | 106.03   | 170.05   | 11.55           |

### Table 10

| Weld joint | R_x 1Ωcm² | CPE 1Fcm⁻²Ω⁻¹ | n | R_p 1Ωcm² | L Henri cm² | R_L 1Ωcm² |
|------------|-----------|----------------|---|-----------|-------------|-----------|
| Weld 3     | 4.90      | 1.85e-4        | 0.93 | 34.86     | 4           | 4.8       |
| Weld 4     | 3.54      | 4.26e-4        | 0.89 | 25.97     | 2.5         | 1.3       |

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**Fig. 26**

a) Potentiodynamic curves of dissimilar weld joints obtained in 0.5M H₂SO₄ solution. b) Nyquist diagrams of dissimilar welds in H₂SO₄ media.
average microhardness was higher (400 HV) due to the martensite formed during the welding cooling stage. Previously, the microstructural analysis demonstrated that the HAZ of the SS 304L side was negligible. Thereby, the microhardness measured in the HAZ-2 and BM-2 (SS 304L) was the same (235 HV), as shown in Fig. 25b.

On the contrary, the average microhardness of the BM-1 (CS A36) in weld 4 was lower (147 HV) than in weld 3 due to the higher ferrite concentration. The HAZ-1 exhibited a slight increase in microhardness (167HV) compared with the BM-1. This increment of hardness was associated with the grain refinement of ferrite and perlite phases (Fig. 7e). The high average microhardness (367 HV) measured in the FZ is related to the lathy martensite microstructure (Fig. 7b). Again, the HAZ-2 and BM-2 (SS 304L) had quite similar average microhardness magnitudes, 222 HV and 218 HV, respectively.

The microhardness measured in dissimilar welds was in good agreement with the results reported in the literature. For instance, Khalifeh et al. [7] and Ma et al. [5] measured the higher microhardness in the FZ of dissimilar welds between CS and austenitic SS. Khalifeh et al. [7] reported a hardness increment in the HAZ of both materials, this increment was higher in the austenitic SS due to the carbide precipitation and recrystallization. Conversely, Ma et al. [5] reported a softening in the HAZ associated with the phase change. In dissimilar welds 3 and 4, the formation of carbides in the FZ of the SS 304L side was not observed. Rather, grain growth was limited by increasing the heat dissipation provided by the CS A36 adjacent fingers (Fig. 13c–d). On the other hand, the heat accumulation in the FZ-HAZ interface estimated by the FE model (Fig. 13d) produced a slight grain growth, higher in the HAZ-2 than in the HAZ-1. Grain growth and the ferritic phase decreased by 6% the HAZ-2 microhardness (weld 4) compared to the HAZ-2 of weld 3 (Fig. 25b).

It is worth mention that the microhardness increases in the FZ of dissimilar welds due to the formed phases. Previously, Torkamany et al. [52] and Tassaloti et al. [2] correlated the %-dilution with the microhardness measured in the FZ of dissimilar welds between CS and SS obtained through the resistance spot welding (RSW) and GMAW processes, respectively. Both studies concluded that the formation of a tough phase (martensite) increased the microhardness regardless of the %-dilution. Tassaloti et al. [2] pointed out that the decarburization and precipitate formation modified the microhardness magnitude in specific FZ regions. In this case, the FZ microhardness of weld 4 decreased by 8% compared to the FZ of weld 3 (Fig. 25b). This hardness reduction was in agreement with phases found in the FZ, martensite in weld 3 (Fig. 7b) and martensite/ferrite in weld 4 (Fig. 10a).

### 4.6 Corrosion analysis in dissimilar welds

#### 4.6.1 H2SO4 medium

Polarization curves to evaluate dissimilar weld joints corrosion resistance in an H2SO4 0.5M solution are shown in Fig. 26a. Electrochemical corrosion parameters are summarized in Table 9. The polarization curve of weld 3 shows a lower corrosion current density (I_{corr}) (7.9 x 10^{-4} A/cm²) compared to weld 4 but, similar potentials are present in both cases (Table 9). The corrosion rate measured for welds 3 and 4 were 9.58 mmpy and 11.55 mmpy, respectively.

Electrochemical impedance curves of dissimilar weld joints are displayed in Fig. 26b. The two semicircles at high frequencies and a “pseudo-inductive” loop at low frequencies were related to the obtained equivalent circuit. The capacitive loop is related to the charge transfer resistance process, while the appearance of the pseudo-inductive loop can be explained by the relaxation of adsorption species such as (SO4^{2-})_{ads} and (H^+)_{ads} on the surface of the electrode. It can also be related to the re-dissolution of the passivated surface [53, 54].

In Fig. 26b, L and R_L are the inductance (Henri cm^2) and inductive resistance (Ω cm^2), respectively. The inductance value in weld 3 was 4 Henri cm^2, which was higher than the weld 4 inductance value (2.5 Henri cm^2). These inductance values can be related to the active area and the corrosion rate. Higher values of L and R_L (Table 10) were obtained with the ER70S-6 filler metal as well as a n-value close to 1. The inductance effect must be considered with n-values different from 1 [55]. The presence of the inductive loop can be related to i) pitting corrosion linked to the adsorption/desorption of intermediates on the electrode surface [56–58] or ii) the accelerated anodic dissolution [59]. The inductive loop could disappear when a corrosion protective layer is formed on the

### Table 11 Electrochemical corrosion parameters of dissimilar weld joints evaluated in a 3.5 wt.% NaCl solution

| Weld joint | E_{corr} (V) | I_{corr} (A/cm²) | b_x (mV) | b_y (mV) | V_{corr} (mmpy) |
|------------|-------------|------------------|----------|----------|----------------|
| Weld 3     | -1.00       | 1.63E-5          | 505.34   | 119.96   | 0.190          |
| Weld 4     | -0.93       | 8.16E-6          | 266.78   | 127.19   | 0.09           |

### Table 12 CEE parameters obtained for dissimilar weld joints in a 3.5 wt. % NaCl solution

| Weld joint | R_a (Ωcm²) | CPE Fcm⁻²S⁻ⁿ | n | R_p (Ωcm²) | C_d (Fcm⁻²) |
|------------|------------|----------------|---|------------|-------------|
| Weld 3     | 10.61      | 4.73e-4        | 0.74 | 1457       | 6.29e-4     |
| Weld 4     | 7.98       | 4.18e-4        | 0.73 | 1018.1     | 4.95e-4     |
4.6.2 NaCl medium

Polarization curves to evaluate the corrosion resistance of dissimilar weld joints in a 3.5 wt. % NaCl solution and Nyquist diagrams are shown in Fig. 27. Electrochemical corrosion parameters are summarized in Table 11. The polarization curve of weld 3 exhibited a corrosion current density (1.63 ×10−5 A/cm²) slightly higher than weld 4 (8.16 ×10−6 A/cm²). The polarization resistance measured in weld 3 was 1457 Ω cm² (Table 12), associated with a higher corrosion rate (0.19 mmpy). This corrosion rate can be correlated to the two phases (ferrite and cementite) detected in the HAZ-1 (Fig. 7e), which formed galvanic cells accelerating electrochemical corrosion reactions. In these reactions, the cementite acts as the cathode and the ferrite as the anode [64]. In addition, the corrosion rate of CS A36 increases with the volume fraction of ferrite [65]. The corrosion rate (0.09 mmpy) of weld 4 decreased in the 3.5 wt. % NaCl solution. However, the corrosion resistance in weld 4 was higher due to the composition of the base and filler metals (high chromium content). A similar result was reported by N. Razak and S. Ng [66] in 1010 steel weld joints. They found that corrosion resistance was improved with the 308L filler metal due to its chemical composition, preventing intergranular corrosion.

The circuit modeled in Fig. 27b is known as the Randles circuit. This is the typical electrical model used to describe the corrosion phenomenon under corrosion attack by electron charge-transfer at the interface metal/electrolyte [67]. Also, the Randles circuit is used to simulate the uniform corrosion on a homogeneous surface, which has been the most used during decades for several studies [67]. The results of fitting the experimental data of impedance using the equivalent circuit (Fig. 27b) are given in Table 12. The constant phase element (CPE) and n-value allowed to calculate the double layer capacitance (Cdl) magnitude through Eq. 8.

\[ C_{dl} = Y_0 (\omega_{max})^{n-1} \]  

where \( Y_0 \) represents the magnitude of CPE, \( \omega_{max} \) is the maximum angular frequency and \( n \) is a physical parameter that gives the interphase properties of the working electrode. In weld 3, the \( C_{dl} \) magnitude was lower (6.29×10−4 Fcm²) than in weld 4 (4.95×10−4 Fcm²). High \( C_{dl} \) magnitude indicates a high electrochemically active surface area. Consequently, the material has a dynamic behavior and calculated \( R_p \) values are inversely proportional to \( C_{dl} \) [68]. The corrosion rate measured in dissimilar weld joints indicated that weld 4 has higher corrosion resistance in a 3.5 wt. % NaCl solution (Table 11).

Additionally, when \( n = 0 \), the equivalent circuit shows characteristics of a pure resistor, and when \( n = 1 \), the equivalent circuit behaves like a pure capacitor. Dissimilar welds behave as capacitors based on the calculated \( n \) values, about 0.74 and 0.73 for welds 3 and 4, respectively. The ideal capacitor is compensated by the CPE when \( n \) values are less than 1. This compensation is due to the deviations caused by the steel surface roughness, the porous layer formation, or the electrochemical heterogeneity, which can be observed in the proposed equivalent circuits [69, 70].

5 Conclusions

The herein research work studied the cracking-susceptibility of dissimilar weld joints between CS A36 and SS 304L through the finger test. In order to find the cracking causes, the metallographic results were correlated with a FE numerical model. Additionally, the corrosion susceptibility of dissimilar weld joints was assessed employing electrochemical tests. Conclusions drawn were the following:
1. Several cracks were found in the dissimilar weld joint with filler metal ER70S-6 (weld 3). These cracks were induced by the residual stresses and the distortion mode (bending). The formation of a unique brittle martensite phase in the FZ contributed to propagating the largest cracks.

2. The dissimilar weld joint with filler metal EC410NiMo (weld 4) exhibited hot cracking caused by the gap between fingers. These gaps were produced during the welding thermal cycle. Also, some cold cracks were found in weld 4, which were associated with the martensitic phase and the distortion (convex and concave bending). δ-ferrite formed in the FZ due to the dilution helped to inhibit the cold and hot cracks growth.

3. In weld 4, the FE simulation predicted a residual stress concentration and concave bending in the last fingers, producing a gap between them. The aforementioned contributed to crack formation in the FZ. Despite the FE model predicting a small gap between fingers 4 and 5 due to the convex bending in weld 2, this gap was not large enough to induce hot cracking.

4. The FE estimations indicated that CS A36 increased the heat dissipation of the SS 304L fingers in dissimilar welds. As a result, the HAZ in the SS 304L side was negligible due to the limited grain growth. However, the heat accumulation in the FZ (weld 4) produced by the low thermal conductivity of the filler metal (EC410NiMo) and the SS 304L fingers generated a slight increase in the grain size of the HAZ compared to the HAZ of weld 3.

5. The element diffusion (C, Cr, and Ni) followed a defined direction, from the base metal to the weld metal. Cr and C were the chemical elements diffused in welds 3 and 4. In weld 3, the Cr diffused to the FZ contributed to forming the brittle martensite phase.

6. The Cr diffusion contributed to obtaining a better corrosion resistance of weld 4 in the saline medium, avoiding pitting corrosion. The corrosion resistance of weld 3 in the acid medium was higher than weld 4 due to the Cr diffusion from the SS 304L (BM) to the FZ.

**Declarations**

**Conflict of interest** The authors declare no competing interests.

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