Friction stir welding of duplex stainless steels*

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ABSTRACT

Duplex stainless steels are successful in a variety of applications such as the food industry, petrochemicals and plants for desalination of seawater, where high corrosion resistance and high mechanical strength are required. However, the beneficial microstructure may change during fusion welding steps, and it can compromise the performance of these materials. Friction stir welding is a solid-state process avoiding typical problems concerning solidification such as solidification cracks, liquation and segregation of alloying elements. Superduplex stainless steels can avoid unbalanced proportions of ferrite and austenite, formation of secondary deleterious phases and grain growth of ferrite in the heat-affected zone. Consolidated friction stir welded joints with full penetration 6 mm thick were obtained for UNS S32101 and S32205 duplex and S32750 and S32760 superduplex stainless steels. The friction stir welds were submitted to tensile tests indicating an improvement of strength in welded joints, showing increased yield and tensile strength for all studied cases. Regarding the microstructural characterization, an outstanding grain refinement was observed in the welded joint, achieving grain sizes as small as 1 μm. This refinement was associated with the combination of microstructural restoration mechanisms in the dual-phase microstructure promoted by severe deformation associated with a high temperature during the welding process.

1. Introduction

Duplex stainless steels (DSSs) occupy an exceptional position among the stainless steels. They are widely used in industries such as the chemical industry, the food industry and the oil and gas industry, among others, owing to the good combination of mechanical performance and corrosion resistance, a combination that is the product of the balanced two-phase microstructure and the absence of harmful secondary phases. The two-phase structure of ferrite (α) and austenite (γ) combines the beneficial effects of the two phases and allows DSS to combine high mechanical strength (ferrite) and good toughness (austenite), even at low temperatures [1].

The practical application of any steel on a large scale is critically dependent on the use of welding. In the case of DSSs, good performance of a welded joint inevitably depends on the acceptable fraction of ferrite and on the absence of harmful phases, such as nitrides and intermetallics in the fusion zone (FZ) and in the heat-affected zone (HAZ).

The friction welding process with a non-consumable pin, known as Friction Stir Welding (FSW), is a technique that uses the rotation of the tool to generate heat by friction at the same time as it imposes pronounced deformation of the material, which is necessary for plasticizing the surrounding metal. After introducing the tool on the sheet, it is moved along the joint, promoting mechanical mixing of the plasticized material and forming a joint behind it (Figure 1). The advantages of the FSW process compared to the fusion welding processes are: (1) excellent reproducibility; (2) elimination of defects produced during fusion and solidification of the metal; (3) lower heat input and maximum temperature associated with the process; (4) elimination of the difficulties connected with hydrogen embrittlement in steels; (5) reduction or elimination of welding fumes; (6) high productivity; (7) lower production cost for some applications; and (8) joining of alloys of limited weldability or dissimilar alloys [2,3].

In general, high cooling rates may result in large amounts of ferrite and formation of nitrides in all DSSs, although formation of austenite is quicker for DSSs with higher contents of N. On the other hand, lower cooling rates and reheating in the temperature range from 600 °C to 1000 °C promote the occurrence of intermetallic phases, mainly in high-alloy duplex or superduplex stainless steels (SDSSs). The desirable microstructure for...
DSSs is achieved by controlling the cooling rate (pre-heating, interpass temperature and heat input) and the chemical composition of the material (choice of steel, filler metal, purge and dilution gas) [4]. In general, all the welding processes used for stainless steels are suitable for DSSs. However, low- and high-energy welding processes must be used with caution, and welding processes without the use of filler metal are normally avoided according to Karlsson [4]. The author mentions FSW only as one possibility for the SDSS UNS S32750. Among all the DSSs, some studies have already been conducted on FSW with reference to steels UNS S32205 by [5] and [6] and for steel S32750 by [7], but these works mainly focused on the metallurgy of the process.

As already mentioned, DSSs generally have good weld ability, but when joined by fusion welding they may undergo changes in their favourable microstructure. These microstructural changes – unbalanced proportions of ferrite and austenite, formation of harmful secondary phases and ferrite grain growth – may cause a decrease in toughness and corrosion resistance in the welded joint [8–11]. FSW technology offers various advantages for the welding of DSSs, among which we may mention in particular joining in the solid state, making it possible to avoid the consequences of the change in chemical composition in the molten pool, which alters the microstructural balance, besides precluding problems such as segregation of alloying elements and formation of solidification cracks [12].

A joint welded by the FSW process has regions with different thermo-mechanical histories, which can be identified in the cross section of the joint, as shown in Figure 1. The stir zone (SZ) is a region of complete recrystallization with a high degree of grain refinement owing to the high levels of strain at high temperature; the thermo-mechanically affected zone (TMAZ) is a region that experiences contributions of heat and strain of lower intensity, exhibiting partial recrystallization or non-recrystallization; the HAZ has heat effects; and the base metal (BM) remains unchanged [12]. The FSW process is not symmetric with respect to the centre line of the joint owing to the different relative speeds reached at the two lateral interfaces of the tool and the welded material, as well as by the different levels of strain on the two sides. The side where the movement of the tool and the flow of material remain in the same direction is called the advancing side (AS), and the side where the movement of the tool is contrary to the flow of material is called the retreating side (RS); both sides are shown in Figure 1.

### 2. Materials and methods

The study was carried out on commercial grades of the DSSs UNS S32101 and S32205 and of the SDSSs UNS S32750 and S32760, which have the chemical compositions shown in Table 1. The chemical composition was supplied by the manufacturers of the respective steels.

The welded joints were produced using special equipment for FSW, model RM-1a from Transformation Technologies Inc., which allows tool control by two mechanisms: positional control or by force control. In the first, how far the pin of the tool penetrates into the joint is defined by means of its position relative to the surface of the joint (tool position in z). In the second, the axial force with which the tool is inserted into the joint is defined. The equipment employed has a capacity for axial force of 67 kN, speed of advance of 1000 mm min\(^{-1}\) and rotational speed of 3000 rpm. The variables of the FSW process were defined and assessed at the design stage. The most important parameters of the process are presented in Table 2.

Most of the initial parameters were defined and maintained throughout the project, such as the materials to be welded, type of joint (butt joint), thickness of the sheets (6 mm) and tool of polycrystalline cubic boron nitride composite in a tungsten–rhenium matrix

### Table 1. Chemical composition (wt%) of the DSSs UNS S32101 and S32205 and of the SDSSs S32750 and S32760.

| UNS     | C   | Si  | Mn  | Cr  | Ni  | Mo | W  | Cu | N  | P   | S   |
|---------|-----|-----|-----|-----|-----|----|----|----|----|-----|-----|
| S32101  | 0.02| 0.70| 5.13| 21.4| 1.62| 0.21| –  | 0.28| 0.024| 0.001|
| S32205  | 0.02| 0.30| 1.80| 22.5| 5.40| 2.80| –  | 0.16| 0.030| 0.001|
| S32750  | 0.02| 0.25| 0.78| 24.9| 6.88| 3.79| –  | 0.34| 0.023| 0.001|
| S32760  | 0.02| 0.35| 0.64| 25.2| 7.00| 3.7 | 0.62| 0.23| 0.024| 0.002|
PCBN-40%WRe). However, the speed of advance, speed of rotation and axial force were variable, and their effects on the welding process were assessed.

Plates with length of 500 mm, width of 90 mm and thickness of 6.0 mm were welded by FSW. The welded joints were produced in the direction normal to the direction of rolling. It should be pointed out that the DSSs underwent crosswise rolling during the production process; therefore, the last rolling pass is taken as the direction of rolling. It should be emphasized, however, that the typical rolled structure is present in the direction of rolling and in the transverse direction, without large differences between them. Development of the welded joints took place in two steps: preliminary joints and final joints. The preliminary welded joints were produced with variation of the rotational speed (200–600 rpm), speed of advance (50–150 mm min⁻¹) and tool penetration by positional control. PN-EN ISO 25239-5 was adopted as the criterion for assessment of the preliminary joints, as presented by Pietras and Węglowski [13], besides taking the appearance of the welded joint into account. The final joints were produced at 200 and 450 rpm, 60 and 100 mm min⁻¹, with force-controlled tool penetration, with axial load of 22 and 37 kN.

Metallographic preparation of the specimens consisted of abrasion with wet-grade emery paper of granulometry from 180 to 1500 mesh, followed by diamond polishing with granulometry of 3 μm and 1 μm. Mechanical–chemical final polishing in silica suspension with granulometry of 0.05 μm was carried out using Vibromet® equipment. Electrolytic etching with 60 vol% of nitric acid (H₂NO₃) in distilled water was used for revealing the microstructure. The parameters used were voltage of 1.50 V and time of 75 s; post-etching was carried out at a voltage of 0.7–0.8 V, for 5 min, to improve the contrast of the ferrite grain boundaries. Test specimens were prepared from the final joints for carrying out mechanical tensile testing, according to standard AWS B 4.0.92 [14]. A test specimen was taken from each welded joint parallel to the line of the joint, for the purpose of determining the strength of the metal processed by the tool and three test specimens being assessed for each steel. The fracture surfaces were examined by SEM. Finally, maps of Vickers hardness (HV₀.₂/₁₅) were prepared by microindentation, to compare the changes in each region of the welded joint.

3. Results and discussion

3.1. Welding process

Figure 2(a) shows a preliminary joint welded with positional control, where the initial penetration of the tool into the material caused fin formation. Higher rotational speeds and axial force are normally used during this step, to facilitate tool penetration into the material and reduce the risk of tool breakage. The highest loads reached in the tool occur in this period, because the material is not sufficiently plasticized. However, once tool penetration is achieved, the axial force is reduced in order to minimize fin formation, since with plasticized material the tool is subjected to far less risk. While carrying out the preliminary tests, the speeds of rotation and advance were determined that are suitable for obtaining joints with a satisfactory surface finish, until a width of about 18 mm was obtained by varying the tool penetration. After welding the joint, lack of penetration at the surface is verified, as shown in Figure 2(a), or at the root of the joint; Figure 2(b) shows the welding of two plates of DSS, in which lack of penetration of the surface was not detected.

Figure 3(a) illustrates the behaviour of the axial load during welding with positional control; comparing this figure with the positions in the welded joint where full penetration was obtained (Figure 2), taking as reference...
the position $x = 0$ (tool penetration), values of axial load were obtained that were suitable for making the defect-free final joints. Figure 3(a) shows in particular the extremely high load, of the order of 43 kN, which is reduced considerably, to about 15 kN, generating the appearance of lack of penetration as indicated in Figure 2(a). Figure 3(a) clearly shows the change in penetration in the positional control mode, until more stable values were reached, which were then used for the final joints in force control mode.

No literature could be found that defines the effect of the tool positioning mode – positional control or force control – on the quality of the welded joint. It is clear that in the positional control mode the tool force changes to maintain constant penetration and the opposite occurs in the force control mode. However, during the preliminary tests and for the final joints, the use of force control was identified as the best option for obtaining sound joints with full penetration without root defects, as can be seen in Figure 3(b). During welding using positional control, continual change in tool penetration was observed, which was attributed to the variation of the axial force with the aim of achieving satisfactory conditions, as described above. In the axial force control mode, tool penetration remained constant, giving joints without root defects. López and Ramirez [15] used positional control exclusively, both for the preliminary joints and for the final joints for production of welded joints in the steel–aluminium system in sheets with thickness of 2.0 mm. For the system studied by the authors, steel–aluminium, the different capacity for plasticization of the two materials had a better response in positional control, since corrections of penetration could be effected as welding progressed.

Figure 3(a) also shows, as response variables, the change in the force in the direction of welding (force in $x$) and the tool temperature. The tool temperature is recorded by a thermocouple inserted near the tool. Although the thermocouple inserted in the tool collar does not record the temperature at the centre of the welded joint, or exactly in the tool, it is possible to obtain comparative values for different welding parameters owing to the high thermal conductivity of PCBN [3,16]. The temperature rise during welding is clear (Figure 3), as is the influence of the axial force on the heating rate and on the maximum values attained. On the other hand, the force in the direction of welding is a criterion of useful life of the tool, according to the manufacturers, and it must remain below 10 kN. Studies carried out by Steel and Sterling [17] showed tool fracture when the value exceeds 15 kN. Therefore, a first welding procedure with positional control makes it possible to determine the conditions for best performance for the tool and, potentially, for the welded joint.

After determining the conditions giving good surface appearance, absence of macroscopic defects, good width/weld penetration ratio (~3:1) and full penetration, a value of axial force was selected for reproducing the welded joints in force control mode. For this, butt joints were made in the alloys studied, as presented in Figure 3(b).

### 3.2. Microstructural characterization

Figure 4 shows the microstructure of the cross section of the welded joint in DSS UNS S32101. The macrograph in Figure 4(a) shows the various regions identified in the joint, which will be referred to throughout the text: base metal (BM), stir zone (SZ), stir zone retreating side (SZ-RS), stir zone advancing side (SZ-AS), root of the stir zone (SZ-root), thermo-mechanically affected zone retreating side (TMAZ-RS) and thermo-mechanically affected zone advancing side (TMAZ-AS). According to Mishra and Mahoney [12], these regions arise on account of geometric factors of the tool and the complex movement of the material causing gradients of strain, temperature and strain rate. The flow lines of material can also be seen in the SZ-AS, which appear on account

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**Figure 3.** Comparative curves during welding for the conditions: (a) positional control in steel S32760 and (b) force control for all the DSSs studied in this work.

Notes: Força = Force; Força axial = Axial force; Força em $x$ = Force in $x$; Deslocamento (mm) = Displacement (mm); Temperatura da ferramenta (°C) = Tool temperature (°C); Força axial: todos = Axial force: all; Força em $x$: todos = Force in $x$: all; Deslocamento da ferramenta (mm) = Tool displacement (mm).
behaviour of the phases that make up the DSSs investigated. Ferrite and austenite have different mechanical performance, as well as mechanism of microstructural restoration at high temperatures. Ferrite is characterized in that it possesses high stacking-fault energy (SFE) and typically exhibits the mechanism of dynamic recovery at high temperatures. On the other hand, austenite possesses low SFE and typically exhibits the mechanism of dynamic recrystallization at high temperatures. Due to its tendency to recover, formation of new ferrite grains may take place by various mechanisms: discontinuous dynamic recrystallization (DDRX), which corresponds to the classical mechanism of recrystallization, by nucleation and grain growth – present in austenite; continuous dynamic recrystallization (CDRX), involving the formation of nuclei by dynamic recovery, with increase in disorientation by rotation due to significant deformation; geometric dynamic recrystallization (GDRX), characterized by marked grain elongation, forming accumulation of dislocations in the interior where they intersect with the high-angle grain boundaries of the original grains, generating a serrated appearance of the grain, so that the undulations get closer until they make

Figure 4. Light microscopy characterization of the welded joint in DSS UNS S32101. (a) Macrograph, (b) BM, (c) TMAZ-RS, (d) SZ and (e) SZ-AS/TMAZ-AS interface. Stereoscopy and OM. Electrolytic etching: 60 vol% HNO₃ in distilled water.

Notes: LR = RS; ZM-LR = SZ-RS; ZM = SZ; ZM-LA = SZ-AS; LA = AS; MB = BM; ZTMA-LA = TMAZ-AS; ZTMA-LR = TMAZ-RS; ZM-raiz = SZ-root; linhas de fluxo = flow lines.

of the asymmetric movement of the material. Mishra and Mahoney [12] also present various studies using markers, which have indicated that the plasticized material is rotated from the AS and advances simultaneously with the pin towards the RS. The region of the AS is subjected to higher levels of strain than the RS, forming flow lines, which are in fact regions with different levels of refinement of grain size and volume fraction of the phases present.

Figure 4(b) shows the typical microstructure of the BM of DSS UNS S32101 corresponding to the islands of austenite (γ) in the ferrite matrix (α). Figure 4(c) shows slightly deformed grains in the TMAZ-RS, but still having similarities with the BM. On the other hand, the SZ, in Figure 4(d), shows pronounced grain refinement of the austenite and a less marked decrease in size of the ferrite grains, with a similar proportion of both phases. Finally, Figure 4(e) shows the micrograph of the SZ-AS/TMAZ-AS interface, clearly showing the deformation of the grains on the TMAZ-AS side.

The difference in size of the austenite and ferrite grains in the SZ and the deformation observed in the TMAZ-AS can be explained on the basis of the mechanical behaviour of the phases that make up the DSSs investigated. Ferrite and austenite have different mechanical performance, as well as mechanism of microstructural restoration at high temperatures. Ferrite is characterized in that it possesses high stacking-fault energy (SFE) and typically exhibits the mechanism of dynamic recovery at high temperatures. On the other hand, austenite possesses low SFE and typically exhibits the mechanism of dynamic recrystallization at high temperatures. Due to its tendency to recover, formation of new ferrite grains may take place by various mechanisms: discontinuous dynamic recrystallization (DDRX), which corresponds to the classical mechanism of recrystallization, by nucleation and grain growth – present in austenite; continuous dynamic recrystallization (CDRX), involving the formation of nuclei by dynamic recovery, with increase in disorientation by rotation due to significant deformation; geometric dynamic recrystallization (GDRX), characterized by marked grain elongation, forming accumulation of dislocations in the interior where they intersect with the high-angle grain boundaries of the original grains, generating a serrated appearance of the grain, so that the undulations get closer until they make
contact, dividing the initial grains and thus creating new micrograins [18]. Furthermore, the mechanisms of recrystallization involved when the microstructure is the same – in the case of austenitic or ferritic stainless steels – are different from those observed in the case of duplex structures, such as in the DSSs and SDSSs [19]. For the systems with duplex structures, in the initial stages of deformation at high temperature, the ferrite in the austenite is deformed more severely, owing to the greater resistance of the austenite, which acts as a matrix for the more ductile phase; with the increase in strength through work hardening in the ferrite, the load is transferred to the austenite and the strain gradient decreases as a result of the adaptation arising from the mechanisms of restoration [20,21].

### 3.3. Mechanical tests

The tensile tests of the base metal and welded joints in the steels investigated were carried out employing test specimens as presented in Figure 5(a), emphasizing that the axial load is exerted parallel to the line of the joint. A force-displacement curve, obtained during tensile testing for steel UNS S32760, is presented in Figure 5(b), which clearly shows a higher yield stress and ultimate tensile strength; however, in contrast to steels S32101 and S32750, it clearly showed a decrease in toughness.

In general, the results reveal an increase in yield point and tensile strength together with an increase in ductility, except for steel UNS S32760. The tests were not carried out for steel UNS S32205 owing to the unavailability of specimens of suitable size. Table 3 presents the data obtained in tensile tests on specimens of the base metal and on longitudinal specimens from the welded joints [22].

From examination of the fracture surfaces after the tensile tests, Figure 6 presents the fractographs of the welded joints for steels UNS (a) S32101, (b) S32750 and (c) S32760. A cellular fracture morphology can be seen, which is a micromechanism typically found in materials with ductile fracture, which is especially the case with DSS UNS S32101 and SDSS UNS S32750, since they displayed high ductility before fracture and a cellular fracture morphology. The appearance of the cells depends on the state of stress. In conditions of axial loading, the cells tend to form in association with particles of second phase and/or interfaces and generally have a spherical appearance, and grow in the plane normal to the stress axis. However, parabolic and elliptical shapes may be present under the action of shearing stresses or a combination of uniaxial stressing with shear [24]. The fracture energy is related to the size and depth of the cells; the larger these are, the greater the associated fracture energy is. Therefore, the fracture in steels S32101 and S32750 corresponds to ductile fracture, which coincides with the elastic behaviour observed in the mechanical tests. The difference appears in the case of welded joints in steel UNS S32760; although it has a cellular fracture surface, ductility in the tensile tests was less than that of the base metal. A possible explanation is precipitation of harmful phases that promoted a decrease in ductility. In fact, corrosion tests by immersion in FeCl₃ carried out by Santos [25] indicated a weight loss much greater than 10 mm·md (milligrams per square decimetre per day) for the welded joint in steel S32760, which according to standard ASTM A923-08 indicates the presence of harmful intermetallic phases in super DSSs.

Maps of Vickers hardness (HV₀.₂/₁₅) by microindentation are shown in Figure 7. The real increase in hardness (maps on the left) and the relative increase in hardness (maps on the right) are shown for all the joints. For example, the hardness of the BM of DSS UNS S32101 is 270 HV₀.₂/₁₅, and in the SZ-AS bottom region, it is around 370 HV₀.₂/₁₅ (Figure 7(a)). The hardness map on the left (Figure 7(b)) shows an increase for the same region (SZ-AS) of the order of 100 HV₀.₂/₁₅; therefore, this map describes the proportional increase in hardness (ΔHV₀.₂/₁₅). Some general aspects may be mentioned in

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*Figure 5.* (a) Test specimen for mechanical tests and (b) force–displacement curve for steel UNS S32760.

Notes: Junta soldada = Welded joint; Força (kN) = Force (kN); Metal de base = Base metal; Junta soldada = Welded joint; Deslocamento (mm) = Displacement (mm).
On the other hand, the bottom part of the SZ of DSS UNS S32101 showed an increase in hardness of the same order as the others. This increase in hardness was attributed to the marked refinement of average grain size in the SZ-AS.

The mechanical tests made it possible to determine the increase in strength and hardness in the processed region. There are four mechanisms of hardening of metals, based on the restriction of the movement of dislocations: by solid solution, by particles of second phase, by work hardening and by grain refinement. Solid solution does not apply, because in FSW there is no change in composition in the stir zone. In the case of DSSs and SDSSs, there is no formation of precipitates with the characteristics of size and distribution that can make advance of dislocations difficult, similar to what happens for example in aluminium alloys, thus ruling out the hypothesis of hardening by particles of second phase. Work hardening is another possible cause of the increase in joint strength, due to the increase in dislocation density during deformation, causing their interaction and blocking; on the other hand, during recrystallization, there is elimination of dislocations, which cancels work hardening. Therefore, in the stir zone, the effects of the work hardening disappear with recrystallization, despite the fact that in the thermo-mechanically affected zone there is partial recrystallization of the deformed metal, the high dislocation density being maintained.

Table 3. Longitudinal tensile tests on welded joints [22].

| UNS         | EL (MPa) | UTS (MPa) | ELO (%) |
|-------------|----------|-----------|---------|
| 32101-BM    | 530*     | 700*      | 30*     |
| 32101-SZ    | 607 ± 2  | 798 ± 11  | 37 ± 1  |
| 32750-BM    | 550*     | 795*      | 20*     |
| 32750-SZ    | 749 ± 12 | 912 ± 3   | 34 ± 1  |
| 32760-BM    | 619*     | 871*      | 25*     |
| 32760-SZ    | 743 ± 34 | 908 ± 67  | 13 ± 4  |

Note: EL: elastic limit; UTS: ultimate tensile strength; ELO: percentage elongation.
*Data from the Outokumpu inspection certificate [23].

On the other hand, particular. It can clearly be seen that the SZ exhibits greater hardness than the BM. A hardness distribution can be seen in the SZ, and the AS shows greater hardness owing to the greater degree of grain refinement in this region. This degree of refinement is related to the greater degree of deformation in this region. The bottom region of the SZ-AS may be mentioned in particular; this had the highest hardness values in the welded joint. This asymmetry is intrinsic to the process, which, during welding, creates gradients of temperatures and strains that promote dynamic recrystallization in some regions, namely the SZ-AS, which had the highest levels of strain and, possibly, higher temperatures.

The highest levels of hardness occurred for SDSSs UNS S32750 and S32760, followed by DSS UNS S32205 and, lastly, S32101. In fact, the SDSSs have higher hardness than DSS UNS S32205, followed by UNS S32101.

![Figure 6](image)

Figure 6. Fractographs obtained by SEM in the SZ of the welded joint (a) UNS S32101, (b) S32750 and (c) S32760 showing the typical cellular morphology for ductile fracture after tensile testing of longitudinal specimens.
which would respond by work hardening of the metal in this region. Regarding the reduction in grain size, this was verified in the stir zone, as a consequence of dynamic recrystallization of both phases – ferrite and austenite – promoted by the combined effect of pronounced deformation and high temperature during the FSW process. As well as the amount of ferrite and austenite, which changes very little during the process, it is considered that the increase in hardness in the whole welded joint mainly occurred by the mechanism of grain size reduction, also known as Hall-Petch.

The hardness maps, besides confirming the significant increase in strength in the welded region, relative to the base metal, reveal the significant changes in properties between the advancing side and the retreating side in the joint. This difference corresponds exactly to the change in average grain size present in both regions, due to the residence time at high temperature of the recrystallized grains from the moment of their formation up to their displacement to the opposite side of the joint. This residence time and the temperature above 1000 °C [26,27] promote grain growth. Furthermore, Su et al. [28] suggest that passage of the metal close to the tool promotes the formation of new grains by the mechanism of recrystallization, due to the heterogeneous deformation induced by the tool under the recrystallized structure and the fact that after emergence of the new grains, the growth phase begins, still under the thermo-mechanical effects of the tool.

4. Conclusions

The development of welding parameters made it possible to achieve satisfactory welding conditions for the four steels studied in this work. The most suitable parameters were rotational speed of 200 rpm, speed of advance of 100 mm/min and producing the welded joints with force control, and the axial force being about 36 kN. This combination also made it possible to use a tool made of PCBN-40%W-Re composite with maximum temperature of about 850 °C and lateral forces less than 10 kN (F_x and F_y), prolonging the useful life of the tool.

The welded joints had excellent surface finish and full penetration, with improvement in mechanical performance. This improvement was associated with the high degree of refinement in the SZ. This refinement was more intensive in the SZ-AS owing to the higher levels of strains to which this region was subjected. The mechanisms of microstructural restoration in the welded joint were dynamic recovery and continuous dynamic recrystallization for ferrite and discontinuous dynamic recrystallization for austenite owing to the pronounced deformation at high temperatures. For steel S32760, there was a decrease in ductility, which may be related to the presence of harmful secondary phases.

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