Microstructural and mechanical characterization of the Ti-containing TWIP steel welded joint produced by GTAW process

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Abstract

The aim of this work is to study the microstructural and mechanical changes generated in a welded joint of Ti-containing TWIP steel by Gas Tungsten Arc Welding (GTAW) process. For this purpose the welding parameters that allowed the union of 5.6 mm thick butt plates with full penetration and without filler material were investigated. Microstructural changes were examined by light optical metallography. Segregation and second-phase precipitated particles were investigated by scanning electron microscopy and electron-dispersive spectroscopy. Phase transformations were evaluated using X-ray diffraction. Finally, the mechanical behavior was assessed by Vickers microhardness and micro-tensile tests. In general, the welded joint showed dendritic structure formation in the fusion zone (FZ) and equiaxed grain growth in the heat affected zone (HAZ). The FZ showed high degree of segregation, where Mn, Si and C segregated in the interdendritic regions, while Al preferentially segregated in dendritic areas. On the other hand, no effect on the austenite stability was noticed. The welded joint shows an increase in microhardness, which is associated with the formation of precipitated particles. Finally, the micro-tensile results showed a decrease in UTS in the FZ, while the welding interface and HAZ showed an increase in UTS compared to the base material.

Keywords: Twinning induced plasticity (TWIP) steel; Ti microalloying element; Welding; Heat input; Microstructure.

1. Introduction

One of the main objectives of the automotive industry is to improve vehicle safety providing larger space, enhanced performance, better reliability and comfort. As a result, the weight in the vehicles has increased, which in turns increases CO₂ emissions to the environment. Accordingly weight reduction issue is a priority in the automobile manufacturing [1]. Consequently, the automotive industry is interested in the application of high-strength steels and/or light-weight alloys [2]. Technological advances regarding these new steels present advantages compared to conventional ferritic-pearlitic steels. Nowadays, fully austenitic steels or austenite-containing multi-phase steels, known as Advanced High-Strength Steels (AHSS) are developed and introduced in the body-in-white car structure. These steels present an optimal combination of strength and formability [3] among which Twinning-Induced Plasticity (TWIP) steels are one of the most representative and promising ones. TWIP steels present high-Mn content and small additions of alloying elements such as Al, Si and C, allowing an austenitic microstructure at room temperature. These steels are characterized by low stacking fault energy (SFE) that allows twinning as a deformation controlling mechanism (the so-called TWIP effect). This mechanism promotes an excellent work-hardening

rate as a result of the generation of deformation twins, which gradually reduce the effective sliding distance of the dislocations, resulting in a sort of dynamic Hall-Petch effect [1-4], combining tensile strength up to 1100 MPa and elongation up to 90% simultaneously [5]. Currently TWIP steels are designed with the following chemical contents: 15-25 wt% Mn, 0-3 wt% Si, 0-3 wt% Al, 0-1 wt% C and eventually secondary additions of Cr, Cu, N, Nb, Ti and/or V [3]. However, the industrial applications of TWIP steels has been limited due to, among other facts, the lack of knowledge on their welding behavior. Indeed preliminary results reports indicate that the high Mn and C content generate undesirable effects in the different welding zones [6-10]. Since good weldability is required to guarantee the use of the TWIP steels in industrial applications, these undesirable effects should be considered in order to take benefit of the high potential of TWIP steels in the automobile structures [8]. The most important problems reported with welding of TWIP steels are related to segregation, dendritic grain formation in the fusion zone (FZ), hot cracking, reduction of mechanical properties and occasionally phase transformations [7, 9, 10]. The latter can be associated to the high degree of segregation during melting, which in turns enriches the liquid phase with Mn and C, generating instability of the austenitic phase leading to unexpected phase transformations, such as eutectic (Fe,Mn)₃C cementite, (γ-M₃C) eutectic phase, δ α-ferrite or ε-martensite [4, 11]. Different authors have also studied the embrittlement in Zn-coated TWIP steels, e.g. Ashiri et al. [8] reported that TWIP steels are susceptible to liquid metal embrittlement (LME) when they are subjected to the resistance spot welding process. Saha et al. [12] reported that the cracks are located in the heat affected zone (HAZ) adjacent to the fusion line and fully filled with solidified liquid metal from the weld nugget. Therefore, the presence of a low melting liquid film along grain boundaries generates lack of plastic compatibility between adjacent grains enhancing the crack appearance. On the other hand, the reduction of mechanical properties in the FZ and HAZ is also generated by microstructural changes as a consequence of the heat input in the welding processes. These changes are due to formation of dendritic grains in FZ and equiaxed grains growth in the HAZ [6]. In consequence one way to minimize these problems is the use of low heat input welding processes.

Because TWIP steels in the annealed condition display relatively low yield stresses, microalloying elements, carbide and nitride formers, can be added. These elements influence the microstructure and mechanical properties by solid solution and the formation of precipitated phases. Reyes-Calderon et al. [13] studied the effect of microalloying elements such as Nb, V and Ti on the hot flow behavior of TWIP steel. They determined that their addition generates a slight increase in the flow stresses at high temperature being the Ti microalloyed TWIP steel the one exhibiting the highest stress values. They also found that additions of V and Ti in TWIP steels promoted grain refinement. It is very important to keep in mind that microalloying particles can pin the grain

boundaries and avoid grain growth, but these particles can be dissolved at high temperatures, allowing then grain growth. This large grain size promotes in turns a delay on the onset of dynamic recrystallization (DRX), a softening mechanism taking place at high temperatures which is also a grain refiner [14]. Research works on microalloyed TWIP steels welding are very scarce. It is for this reason that the present work aims primarily to study the microstructure and mechanical properties of a Ti microalloyed TWIP (TWIP-Ti) steel welded by the gas tungsten arc (GTAW) process.

2. Experimental details

2.1. Base material

The present Fe-22Mn-1.8Al-1.2Si-0.57C-0.014N-0.022Ti (wt%) Ti-containing TWIP steel was melted in the Foundry Lab of the Metallurgical Research Institute-UMSNH (Mexico) using high purity raw materials, in a 25 kg capacity induction furnace. Rectangular specimens were extracted from the ingots and homogenized at 1200 °C and then hot-rolled to 60 % thickness reduction. They were slow cooled down to room temperature. Additionally, specimens were reheated at 1200 °C and hot-rolled to 50 % thickness reduction, followed by solution heat treatment at 1200 °C and then water quenched. The heat and thermomechanical treatments were performed in order to generate a microstructural conditioning of the steel, modifying the dendritic structure of the solidification, and obtaining a refined equiaxial austenitic grain microstructure. Mechanical properties of the studied TWIP-Ti steel in the as-solutioned condition were yield strength (YS) of 380 MPa, ultimate tensile strength (UTS) of 739 MPa, elongation at fracture of 78% and Vickers microhardness of 395 ± 19.4 HV_{0.5}.

2.2. Welding Parameters

The dimensions of the pair of plates for weldability test were 85 mm in length, 45 mm in width and 5.6 mm in thickness. The welding joint was carried out using the Gas Tungsten Arc Welding (GTAW) semi-automatic process without filler material and butt joint. In this case, direct current with reversed polarity, arc-length of 1.5 mm, tungsten electrode EWTh-1 of 1.587 mm diameter and Ar as shielding gas were used. The operating variables and heat input used in the weld joint are shown in Table 1. Equation 1 [15] was use to calculate the heat input:

$$Q = \frac{VI}{v} f \dots\dots\dots(1)$$

where Q is the heat input (kJ/mm), V is the voltage (V), I is the electric current (A), v is the displacement speed (mm/s) and f is the electric arc efficiency.

Table 1. Operating variables used in the GTAW process of the TWIP-Ti steel.

Electric current (A)	Voltage (V)	Displacement speed (mm/s)	Heat input (kJ/mm)
91	10.92	1.16	0.574

Additionally, the temperature distribution through the HAZ was measured at 7, 27, 37 and 42 mm from the welding interface by means of K-type thermocouples.

2.3. Characterization in as-solutioned and post-welding condition

The microstructural analysis of the TWIP-Ti steel in as-solutioned and post-welding condition was carried out using light optical microscopy (LOM). For this purpose, samples were prepared through chemical etching, where the interest areas were revealed by using nital 4% for 5 s and a solution of $\text{Na}_2\text{S}_2\text{O}_5$ (2 g) in 100 ml of distilled water for 15 s. The average grain size in the HAZ was measured using the Sigma Scan Pro 8 software. Segregation and precipitated particles in the welded joint was determined using point and elemental mapping chemical analysis by Electron Dispersive Spectroscopy (EDS) using a Scanning electron microscope (SEM) JEOL JSM-7600F. Phase transformations were evaluated by X-ray diffraction (XRD) using Cu-K radiation ($\lambda = 1.5402 \text{ \AA}$) in a D5000 Siemens diffractometer with the following operating conditions: 2θ range from 20 to 120° , step of 0.02° and time of 1 s per step.

Vickers microhardness tests in as-solutioned and post-welding condition were performed using a Wilson microhardness tester with load of 500 g, by indentations along the cross section of the weld bead, in the middle of the sample thickness. The distance between indentations was of 250 μm in the FZ, 200 μm in the HAZ and 300 μm in the base material.

Micro-tensile tests were also carried out in the as-solutioned and post-welding condition (particularly in the FZ, close to the welding interface (WI) and in the HAZ areas) with specimens of the geometry and dimensions shown in **Figure 1**. The specimens were prepared by wire electrical discharge machining (EDM). The micro-tensile tests were conducted in a Microtest DEBEN testing machine at a constant speed of 2 mm/min.

Figure 1. a) Geometry and dimensions of the specimens used for the micro-tensile test. b) Specimens extraction location.

3. Results and discussion

3.1. Base material microstructure

The microstructure of the TWIP-Ti steel in the as-solutioned condition consists of fully austenite equiaxial grains with annealing twins (see **Figure 2a**). The average grain size was $64 \pm 35 \mu\text{m}$. A heterogeneous distribution can be observed as grain sizes range from 15 to $286 \mu\text{m}$, which could be associated with the presence of abnormal grains. The statistical analysis showed that grain size has a log-normal distribution, as shown in **Figure 2b**, with a marked right hand tail as consequence of the latter abnormal grains. These heterogeneous grain size distribution can be associated with the interference between the recrystallization (annealing) kinetics and the precipitation dissolution of the microalloying elements [16]. It is well-known that Ti allows the formation of precipitates at high temperatures, which can inhibit grain growth [17]. Mejia et al. [18] showed the effect of Ti as a grain refiner in TWIP steels, reporting that a TWIP steel microalloyed with Ti exhibited a grain size of $63 \pm 32 \mu\text{m}$ while the TWIP steel without microalloying element showed a grain size of $104 \pm 23 \mu\text{m}$ after austenization at high temperatures. Most probably the present annealing temperature was not enough to completely dissolve the TiN particles, and therefore some abnormal grain growth took place. This explains the apparently large standard deviation in the grain size distribution.

Figure 2. Microstructure of the studied TWIP-Ti steel in as-solutioned condition. a) LOM micrograph, and b) Grain size histogram.

3.2. Post-welding microstructure

The welding of the TWIP-Ti steel revealed a uniform joint, where the welding bead presented a complete penetration and good aspect, free of porosity and cracking. The different zones generated during the welding process are showed in **Figure 3**, where three clear zones, namely the fusion zone (FZ), the welding interface (WI) and the heat affected zone (HAZ) are clearly depicted. The FZ shows a fully austenitic dendritic microstructure formed by columnar and equiaxial grains. The equiaxed dendritic grains are mostly noticed in the center of the weld bead, while columnar dendritic grains are perpendicularly oriented to the WI, i.e., the transition region between melted and non-melted areas (see **Figures 3a** and **3b**). It is particularly interesting to notice the relatively small width of this FZ, $\sim 3000 \mu\text{m}$, which is due to the low heat input used during the welding process.

Figure 3. Microstructures of the TWIP-Ti steel welded joint. a) Welding cord section, b) Dendritic grain in the FZ, c) Welding interface (WI), and d) HAZ extension of equiaxial grain growth.

On the other hand, the microstructural changes in the HAZ consist of coarsened austenitic grains, as compared to those of the TWIP-Ti steel in the as-solutioned condition (see **Figure 3c and 3d**). The HAZ presented grain sizes up to 155 μm, very close to the transition region between melted and non-melted region (see **Figure 3d**), due to the grain growth caused by the heat generated during the welding process. Due to the austenitic nature of this steel, the HAZ length can not be simply derived by observing areas where microstructural phase change may have taken place. Alternatively it can be derived by optical observation of the grain size and hardness profiles. In this sense, **Table 2** shows the average grain size at different distances from the weld bead center. As expected, the zone adjacent to the welding interface showed the largest grain size of $76 \pm 32 \mu\text{m}$ (1.82 mm from centerline of the weld bead). From that point the grain size was rapidly lowering down to achieve a

similar value to the one of the base material (taking account of the large standard deviation). In other words, and taking account of the grain size profile, the HAZ extension seems to be very narrow ($\sim 450 \mu\text{m}$), as shown in **Figure 3d**. On the other hand, the smallest grain size, corresponding to $50 \pm 27 \mu\text{m}$, was observed at 5 mm from the weld bead center (see **Table 2**).

Table 2. Average grain size at different distances from the weld bead center.

Distance from weld bead center (μm)	-25000	-14000	-5000	-3370	-2520	-1820	0	5000	14000	25000
Average grain size (μm)	64	72	60	67	67	76		50	61	56
Standard deviation	25	37	34	38	36	32		27	24	28

It is worth noting that the grain size on both sides of the weld bead showed slight differences, which can be associated to the at least two factors: i) Heterogeneity of the TWIP-Ti steel grain size in the as-solutioned condition, and, ii) A non-uniform heat input distribution. It is well-known that the solidification of metallic alloys involves the formation of crystals with a preferred growth direction due to heat flux from the system to the surroundings, which promotes instability of the solidification front as a result of constitutional supercooling in the molten metal and the dendritic structure formation generated [19, 20]. Furthermore, the constitutional supercooling in the center of the weld bead is larger than that close to the weld interface. Thus, equiaxed dendrites nucleated and grown from the weld centerline, blocking columnar dendritic growth [21]. On the other hand, different research works refer to the excessive equiaxial grain growth in the HAZ of TWIP steels, using different fusion welding processes such as laser and resistance spot welding [21, 22]. It is worth highlighting that although the HAZ of TWIP-Ti steel showed grain growth, this growth was not excessive, which can be associated to the low heat input used in the semi-automatic GTAW process, and the grain growth inhibition effect of Ti as a microalloying element. [17, 18].

3.3. Segregation and precipitated particles

Figure 4 displays the chemical composition profiles and elemental mapping images of the alloying elements in the FZ of the TWIP-Ti steel welded joint. Mn, Al, Si and C were considered in this analysis. It is a general observation that Al segregated towards the interior of the dendrites, while Mn, Si and C segregated towards the interdendritic zone (see **Figure 4a**). On the other hand, the variations of the alloying elements were quantified by means of EDS point analysis and compared

with the TWIP-Ti steel in as-solutioned condition. As expected, the high Mn content in this steel caused high chemical variation around to 6.2 % with a minimum and maximum value of 18.1 and 24.3 % Mn, while the Al, Si and C presented variations corresponding to 0.2 % (minimum and maximum value of 1.42 and 1.62 % Al), 0.67 % (minimum and maximum value of 0.88 and 1.55 % Si) and 0.13 % (minimum and maximum value of 0.44 and 0.57 % C), respectively (see **Figure 4b**). It is well-known that during solidification, the dendrite core pushes out the alloying elements with distribution coefficients (k) lower than unity (i.e. Mn, C and Si) to the interdendritic region [23]. It has been reported that the coefficient k for the Mn and C segregation are approximately 0.75 and 0.28, respectively [24]. This segregation is particularly important in TWIP steel welding, because is favored by the high content of alloying elements [25]. However, at high solidification rates a finer dendritic structure and less segregation is generated [26]. Therefore, the use of the lowest heat input in TWIP steels fusion welding processes is strongly recommended.

Figure 4. Distribution of the alloying elements in the fusion zone (FZ) of the TWIP-Ti steel welded joint. a) Elemental mapping of chemical elements distribution, b) Concentration profiles for Mn, Al, Si and C, c) Mn segregation pattern, and d) Si segregation pattern.

In addition, precipitated particles were observed in the different welding zones. These particles were identified as AlN, Ti (C,N), MnS and TiC (see **Figures 4a** and **5**). However, the distribution of these particles was different in each zone, i.e., AlN, Ti (C,N) and MnS are more noticeable in the FZ and HAZ (i.e., the narrow region with larger grain size than the base material), located along the interdendritic boundaries. On the other hand, TiC particles were observed within the austenitic matrix. No need to say that precipitation in the grain boundaries can induce embrittlement [27]. The morphology and chemical composition of the particles precipitated in the TWIP-Ti steel weld joint can be seen in **Figure 5**.

Figure 5. Precipitated particles in the welded joint of TWIP-Ti steel. a) AlN, b) Ti(C,N), c) MnS, and d) TiC particles.

3.4. Phase transformations

Figure 6 shows the X-ray diffraction patterns in as-solutioned and post-welding conditions. Characteristic peaks belonging to the γ -austenite (FCC) phase are present in both conditions, which coincide with the 2θ angles at 44° , 50° , 74° , 89° and 95° , which correspond to the crystallographic planes (111), (200), (220), (311) and (222), respectively. No evidence was found of phase

transformations such as formation of δ α -ferrite, ϵ -martensite, cementite (Fe,Mn)₃C and eutectic phases (γ -M₃C) [4, 28, 29] during the welding process. In consequence the studied TWIP-Ti steel has fully austenitic microstructure (see **Figure 6**). This means also that although the fusion welding processes can produce segregation and/or evaporation of some alloying elements (e.g., Mn and C), this phenomena was not high enough here to generate unexpected phase transformations in TWIP steels. Otherwise, poor zones in C-Mn can promote a low austenite stability [28], which in turns it is related to the dependence of the stacking fault energy (SFE) on temperature and chemical composition [4].

Figure 6. X-ray diffraction patterns of TWIP-Ti steel in as-solutioned and post-welding condition.

3.5. Microhardness.

The average value and profile microhardness of the current TWIP-Ti steel in as-solutioned and post-welding condition are shown in **Figure 7**. The present TWIP-Ti steel shows an average microhardness value of $395 \pm 19.4 \text{ HV}_{0.5}$ in the as-solutioned condition. Higher values are observed in the FZ and HAZ. This behavior can be explained in terms of precipitation aging in the TWIP-Ti steel, because the solution heat treatment at 1200 °C and subsequent water quenching, generated unstable super-saturated solid solution. Consequently, the welding process provides the thermal energy necessary to form precipitated phases as a function of the peak temperature, cooling rate and chemical composition [30]. According to the results obtained by SEM-EDS, the hardening generated in the TWIP-Ti steel joint could be attributed to the formation of TiC, Ti(C,N) and AlN precipitated particles. Similarly to the grain size distribution, the hardness distribution on both sides of the weld bead is non-uniform. On the left side of the FZ (see **Figure 7**), higher hardness values

can be observed in the vicinity of the welding bead. On the other hand, the lower hardness value is also generated on the left side of the FZ at 25 mm. These differences are again associated to the type of microstructure and the grain size present in each zone. It is well-known, that softening in the FZ is associated to the formation of dendritic microstructure and lower carbon content due to the decarburization during welding process [31]. However, the changes generated in the HAZ were generated in solid state. Out of the FZ, the hardness values still remain higher than in the base material, indicating that in fact, the HAZ, in terms of modification of alteration mechanical properties (in this case due to precipitation) is much larger than expected according to the grain size evaluation. Indeed the HAZ length according to this criterion would be as large as the welded sample.

Figure 7. Average value and profile microhardness of the TWIP-Ti steel in as-solutioned and post-welding condition.

Figure 8 shows the thermal cycles recorded with the thermocouples in the HAZ of the welded joint. As well, peak temperatures and cooling rates are listed in **Table 3**. It is well-known that heat distribution through the HAZ generates important microstructural changes such as recrystallization, grain growth, phase transformations, formation of precipitates, residual stresses and stress relaxation [22]. However, in those conditions where the heat input is low and the heat flow is effective (high cooling rate), less microstructural changes are produced by effect the heat input [22].

Figure 8. Thermal cycle in the welded joint of TWIP-Ti steel through GTAW process.

Table 3. Peak temperature and cooling rate in the HAZ of the TWIP-Ti steel welded joint.

Thermocouple	Distance from the WI (mm)	Peak temperature (°C)	Cooling rate (°C/s)
T1	7	688.7	2.24
T2	27	460.1	0.82
T3	37	264.0	0.19
T4	42	210.4	0.14

It is worth noting that according to **Table 3** and the hardness profile (see **Figure 7**), it seems that temperatures as low as 210 °C are promoting some precipitation hardening, which should be studied in more detail to determine the aging mechanisms where Ti microalloying element plays an important role.

In order to get a better understating of the microstructures and precipitated phase in the affected areas of the welding process, the equilibrium phase diagram of the present TWIP steel was calculated through the software JMatPro® v9.1. As shown in **Figure 9a**, austenite begins to solidify at 1390 °C, being completed at 1289 °C, maintaining its stability up to room temperature. In addition, second phases formation is also observed and according to the thermodynamic calculations, the main precipitated phases are the following ones: AlN, Ti(C, N), TiC, MnS and

M₂P compounds (see **Figure 9b**) which are basically in accordance with the experimental evidences already reported above. AlN starts forming at 1521 °C, before the solidification ends (approximately at 1300 °C). From this temperature, the TiN precipitation begins to increase as the temperature decreases, until 1190 °C where the formation of Ti(C,N) and TiC begins. On the other hand, the MnS formation can be observed from approximately 1300 °C and reaches its stability in a very short temperature range. Additionally, the formation of M₂P particles starts forming at approximately 530 °C. These particles were not observed by SEM-EDS, most probably because their precipitation kinetics is very slow, and no time to their nucleation and growth is available. However, the percentage and distribution of precipitated particles in the different welding zones can be different due to the heating and cooling rates involved. This is clearly shown in **Figure 10**, where the HAZ displays a greater amount of particles, which are distributed in the austenitic grain matrix (see **Figure 10b**), in comparison with the FZ and the TWIP-Ti steel in as-solutioned condition.

Figure 9. JMatPro® v9.1 predictions of TWIP-Ti steel. a) Equilibrium phase diagram, and b) Second-phases prediction.

Figure 10. Precipitated particles in TWIP-Ti steel. a) As-solutioned condition, b) Fusion zone (FZ), and c) Heat affected zone (HAZ-5 mm).

3.6. Micro-tensile tests.

The micro-tensile tests from samples extracted in different welding zones of the TWIP-Ti steel in the as-solutioned (base material) and post-welding condition are shown in **Figure 11**. The stress and strain values of the TWIP-Ti steel in as-solutioned condition were used as reference for the mechanical behaviour analysis in the different welding zones. In this condition the base material exhibited a YS of 385 MPa and UTS of 739 MPa with an elongation to rupture of 78 %, while the sample extracted from the FZ showed a YS of 372 MPa and UTS of 669 MPa with an elongation to rupture of 79.9 %. In general, this zone exhibited the lowest values of UTS, which corroborates that the high heat input in the TWIP-Ti steel welded joint significantly affects the tensile strength. Contrary to the FZ, the zone closer to the WI exhibited the highest values of YS of 426 MPa and UTS of 802 MPa with elongation to rupture of 76 %. The UTS increased 8.2 % while the elongation decreased 2.5 % at the weld interface with respect to the base material, as shown in **Figure 11**. On the other hand, the HAZ samples extracted at 5 and 30 mm from the weld bead center also showed an increase in the UTS with values between 760-768 MPa, which is 3.9 %

higher than in the base material. In general, the HAZ samples showed an increase in the elongation to rupture with values between 79.5-82 % at 10 mm from the weld bead center, which is 5 % greater than the base material. However, at 5 mm from the weld bead center, the elongation decreased 6.4 %, being this zone the most critical one, with an elongation to rupture of 73 %, the lowest one in all areas tested. On the other hand, the greatest elongation to rupture was exhibited at 30 mm from the weld bead center with a value of 82 % (see **Figure 11**). It is important to note that the YS in the HAZ is higher compared to the BM and FZ, but lower than near the WI (see **Figure 11b**). The differences in UTS and elongation to rupture can be associated to different factors such as chemical composition variation, solidification condition and thermal process experienced by the welding [27]. This effect can be explained in terms of either solid solution or second-phase particles strengthening (i.e. the interaction of dislocations with different obstacles) [32].

Figure 11. Micro-tensile tests in TWIP-Ti steel in as-solutioned (base material) and post-welding condition. a) Engineering stress-strain curves, and b) YS, UTS and elongation to rupture.

In this way, the dendritic microstructure in the FZ and the presence of particles such as AlN, MnS and Ti (C,N) located in the dendritic grain boundaries, explains the lowest strength of this zone. Contrary, equiaxial grains in the HAZ with the presence of fine particles of Ti (C,N) and TiC distributed inside the grains explains the excellent mechanical behavior in this region. In any case, the large ductility values, and the absence of brittleness in the welded areas is an indication of the excellent mechanical behavior that welded joints of TWIP steels can present, provided correct welding parameters are employed. This fact is associated to the intrinsic good work hardening ability of this type of steels.

Figure 12 shows the SEM micrographs of the fractured surfaces. The type fracture shown by the micro-tensile specimens in as-solutioned and post-welding conditions was ductile failure as

illustrated by the evident nucleation, growth and coalescence of dimples and voids. This behavior is clearly associated to the high ductility shown in the micro-tensile tests. On the other hand, specimens with greater deformation exhibited a larger number of dimples (see **Figures 12a** and **12b**) than samples depicting less deformation capacity (see **Figures 12c** and **12d**).

Figure 12. SEM micrographs of fracture surface of TWIP-Ti steel: a) As-solutioned condition, b) Fusion zone, c) FZ/HAZ interface, and d) HAZ at 30 mm from the weld bead center.

Figure 13 shows SEM images of the microstructural evolution during micro-tensile tests in the base material condition and different welding zones (FZ and HAZ). Deformation twins are readily shown as thin bands of a few tenths of micrometers wide. However, deformation twins were not evident in all austenitic grains which can be explained in terms of the crystallographic misorientation between the grains and the applied external force. On the other hand, it can be seen that some zones have different systems and density of twinning, e.g., in the as-solutioned condition has a parallel bands twinning system (see **Figures 13a** and **13b**) while in the FZ shows a ladder-shaped twinning pattern, i.e., secondary twins appears within the primary one (see **Figures 13c** and **13d**). In the case of the HAZ, a combination of twinning patterns is shown, i.e., parallel and ladder-

shaped bands (see **Figures 13e and 13f**). Furthermore, the presence of precipitated particles and cavities was also observed in the austenitic matrix. These cavities may be the result of the detachment of precipitated particles due to plastic deformation. Typically, fusion welding processes cause loss of the mechanical properties. However, in the welding of the studied TWIP-Ti steel was observed that mechanical properties do not suffer a significant loss, which can be associated to the twinning effect that was generated independently of the different microstructures of the welded joint and the TiC and Ti (C,N) particles formation. Wang et al. [33] suggest that according to the generated structures by the deformation twins, there are three types of grains in plastically deformed TWIP steels: i) Type I grains that contain few or no deformation twins, ii) Type II grains that exhibit a well-developed twin structure along one active twinning system, and iii) Type III grains that have well-developed twin structure along more than one active twinning systems. This depends of the type of the microstructure of the TWIP steel, i.e., equiaxial, columnar or dendritic, as well as the orientation and size of grains, which is evident in the different welding zones of the studied TWIP-Ti steel, particularly in the FZ and HAZ where the mechanical behavior is different.

Figure 13. Deformation twinning in micro-tensile specimens in the as-solutioned and post-welding conditions. a), b) As-solutioned; c), d) Fusion zone; e), f) Heat affected zone.

Conclusions

Based on results obtained from the metallographic, structural and mechanical characterization of the Ti-containing TWIP steel welded joint, the following conclusions can be drawn:

1. The high content of alloying elements in the TWIP-Ti steel causes the formation of the dendritic structure and Mn and C segregation in the fusion zone (FZ).
2. The heat input in welded joint by the GTAW process caused very limited grain growth in the heat affected zone (HAZ).
3. No phase transformations of austenite were observed with the welding parameters used to join the present TWIP-Ti steel, which is indicative of good weldability.
4. Second-phase particles (TiC, Ti(C,N) and AlN) were formed in the FZ and HAZ of the welded joint promoting precipitation hardening effect.
5. The microstructural changes in the FZ of the welded joint generated a slight decrease in the ultimate tensile strength (UTS) compared to the base material, while the elongation to rupture slightly increased.
6. In general, the heat affected zone (HAZ) exhibits excellent ultimate tensile strength (UTS) and elongation to rupture, which is associated to the deformation mechanism by twinning and the fine precipitated particles formation.
7. The elongation to rupture values in the different weld zones are little affected and remain high, which guarantees a high mechanical performance of the welded joint.
8. No brittle behavior was observed in the welded area, a feature associated to the intrinsic work hardening ability of these steels. This behavior is also an indication of the optimal behavior of TWIP steels to welding processes when appropriated welding parameters are selected.

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References

- [1] B.C. De Cooman, K. Chin and J. Kim, *High Mn TWIP steels for automotive applications*. In: New Trends and Developments in Automotive System Engineering (Eds: M. Chiaberge), InTech, Shanghai, **2011**, pp. 101-128.
- [2] M. Tisza, *Key Eng. Mater.* **2014**, 381, 137.

- [3] B.C. De Cooman, Y. Estrin and S.K. Kim, *Acta Mater.* **2018**, 142, 283.
- [4] O. Grässel, L. Krüger, G. Frommeyer and L.W. Meyer, *Int. J. Plast.* **2000**, 16, 1391.
- [5] M.Y. Demeri, in *Advanced High-Strength Steels: Science, Technology, and Applications*, First ed. (Eds; ASM International), **2013**, pp. 1-312.
- [6] I. Mejía, H. Hernández-Belmontes and C. Maldonado, *MRS Adv.* **2017**, 2, 3899.
- [7] J. Yoo, B. Kim, Y. Park and C. Lee, *J. Mater. Sci.* **2015**, 50, 279.
- [8] R. Ashiri, M. Shamanian, H.R. Salimijazi, M.A. Haque, J. Bae, C. Ji, K. Chin and Y. Park, *Scr. Mater.* **2016**, 114, 41.
- [9] H. Hernández-Belmontes, I. Mejía and C. Maldonado, *MRS Proc.* **2016**, 1812, 35.
- [10] L. Mujica, S. Weber, C. Thomy and F. Vollertsen, *Sci. Technol. Weld. Joining.* **2009**, 14, 517.
- [11] L.M. Roncery, S. Weber and W. Theisen, *Scr. Mater.* **2012**, 66, 997.
- [12] D.C. Saha, I. Chang and Y. Park, *Mater. Charact.* **2014**, 93, 40.
- [13] F. Reyes-Calderón, I. Mejía, A. Boulaajaj and J.M. Cabrera, *Mater. Sci. Eng. A.* **2013**, 560, 552.
- [14] A.I. Fernández, P. Uranga, B. López and J.M. Rodríguez-Ibabe, *Mater. Sci. Eng. A.* **2003**, 361, 367.
- [15] S. Kou, in *Welding Metallurgy*, Second ed. (Eds: John Wiley & Sons), **2003**, pp. 1-480.
- [16] M. Sabet, A. Zarei-Hanzaki and S. Khoddam, *J. Eng. Mater. Technol.* **2009**, 131, 44502.
- [17] B. Mintz and D.N. Crowther, *Int. Mater. Rev.* **2010**, 55, 168.
- [18] I. Mejía, F. Reyes-Calderón and J.M. Cabrera, *Mater. Sci. Eng. A.* **2015**, 644, 374.
- [19] W. Kurz and D.J. Fisher, *Fundamentals of solidification*, Fourth ed. (Eds: Trans Tech Publications Ltd), **1998**, pp. 1-305.
- [20] K. Easterling, *Introduction to the Physical Metallurgy of Welding*, Second ed. (Eds: Butterworth-Heinemann), **1992**, pp. 1-282.
- [21] W. Guo, Z. Wan, Q. Jia, L. Ma, H. Zhang, C. Tan and P. Peng, *Opt Laser Technol.* **2020**, 124, 105961.
- [22] D.C. Saha, S. Han, K.G. Chin, I. Choi and Y.D. Park, *Steel Res. Int.* **2012**, 83, 352.
- [23] D. Kim, K. Han, B. Lee, I. Han, J.H. Park and C. Lee, *Metall. Mater. Trans. A.* **2014**, 45, 2046.
- [24] X. Zhenming, L. Gao, G. Qingfeng and J. Qi-Chuan, *Mater. Res. Bull.* **2004**, 39, 457.
- [25] T. DebRoy and S.A. David, *Rev. Mod. Phys.* **1995**, 67, 85.
- [26] H. Mae, X. Teng, Y. Bai and T. Wierzbicki, *Int. J. of Solids Struct.* **2008**, 45, 1430.
- [27] A.E. Salas-Reyes, I. Mejía, A. Bedolla-Jacuinde, A. Boulaajaj, J. Calvo and J.M. Cabrera, *Mater. Sci. Eng. A.* **2014**, 611, 77.

- [28] C. Lee, Y. Jaehong, S. Kim, Y. Park and J. Choi, *Characteristics of the hot cracking and segregation behaviour in the high manganese steels welds*. In: Proceedings of the 1st International Conference on High Manganese Steels, Seoul. F-7, **2011**.
- [29] G. Frommeyer, E.J. Drewes and B. Engl, *Rev. Met. Paris*. **2000**, 97, 1245.
- [30] A. Dumay, J.-P. Chateau, S. Allain, S. Migot and O. Bouaziz, *Mater. Sci. Eng. A*. **2008**, 483-484, 184.
- [31] R.E. Smallman and A.H.W. Ngan, *Physycal Metallurgy and Advanced Materials Engineering*, Seventh ed. (Eds: Butterworth-Heinemann), **2007**, pp. 1-672.
- [32] J.C. Lippold, *Welding Metallurgy and Weldability*, First ed. (Eds: John Wiley & Sons), Inc., Hoboken, New Jersey, **2015**, pp. 1-442.
- [33] D. Wang, K. Wang, Z. Li, X. Wang, X. Wang and F. Han, *Mater. Sci. Eng. A*. **2015**, 636, 396.

Table captions:

Table 1. Operating variables used in the GTAW process of the TWIP-Ti steel.

Table 2. Average grain size at different distances from the weld bead center.

Table 3. Peak temperature and cooling rate in the HAZ of the TWIP-Ti steel welded joint.

Figure captions:

Figure 1. a) Geometry and dimensions of the specimens used for the micro-tensile test. b) Specimens extraction location.

Figure 2. Microstructure of the studied TWIP-Ti steel in as-solutioned condition. a) LOM micrograph, and b) Grain size histogram.

Figure 3. Microstructures of the TWIP-Ti steel welded joint. a) Welding cord section, b) Dendritic grain in the FZ, c) Welding interface (WI), and d) HAZ extension of equiaxial grain growth.

Figure 4. Distribution of the alloying elements in the fusion zone (FZ) of the TWIP-Ti steel welded joint. a) Elemental mapping of chemical elements distribution, b) Concentration profiles for Mn, Al, Si and C, c) Mn segregation pattern, and d) Si segregation pattern.

Figure 5. Precipitated particles in the welded joint of TWIP-Ti steel. a) AlN, b) Ti(C,N), c) MnS, and d) TiC particles.

Figure 6. X-ray diffraction patterns of TWIP-Ti steel in as-solutioned and post-welding condition.

Figure 7. Average value and profile microhardness of the TWIP-Ti steel in as-solutioned and post-welding condition.

Figure 8. Thermal cycle in the welded joint of TWIP-Ti steel through GTAW process.

Figure 9. JMatPro® v9.1 predictions of TWIP-Ti steel. a) Equilibrium phase diagram, and b) Second-phases prediction.

Figure 10. Precipitated particles in TWIP-Ti steel. a) As-solutioned condition, b) Fusion zone (FZ), and c) Heat affected zone (HAZ-5 mm).

Figure 11. Micro-tensile tests in TWIP-Ti steel in as-solutioned (base material) and post-welding condition. a) Engineering stress-strain curves, and b) YS, UTS and elongation to rupture.

Figure 12. SEM micrographs of fracture surface of TWIP-Ti steel: a) As-solutioned condition, b) Fusion zone, c) FZ/HAZ interface, and d) HAZ at 30 mm from the weld bead center.

Figure 13. Deformation twinning in micro-tensile specimens in the as-solutioned and post-welding conditions. a), b) As-solutioned; c), d) Fusion zone; e), f) Heat affected zone.

This research work studies microstructural and mechanical changes of Ti-containing TWIP steel welded joint by GTAW semi-automatic process in 5.6 mm thick butt plates without filler material and full-penetration. No phase transformations occur, which is indicative of good weldability. TiC and Ti(C,N) precipitate in FZ and HAZ. Welded joint exhibits excellent UTS and elongation, which is associated to twinning deformation.