The effect of heat input on microstructure and HAZ expansion in dissimilar joints between API5L X80 / DSS 2205 steels using thermal cycles

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ABSTRACT: In this research, the effect of the shielded metal arc welding (SMAW) process heat input upon the microstructure and development of the heat-affected zone in the dissimilar joint of API 5L X80/DS5 2205 steels was investigated by recording the thermal cycles with thermocouple implantation in the perpendicular direction of the weld line. The filler metal used (electrode) is DSS 2209. The microstructure of the base and weld metals and their interfaces at different heat inputs were investigated using the scanning electron microscopy/energy-dispersive spectroscopy analysis technique (SEM/EDS) and optical microscopy (OM). The results indicated that the interface between the base metals and the weld metal has excellent consistency and that there is no evidence of cracks at different heat inputs. By increasing the heat input, it was determined that the amount of secondary austenite in the weld metal and heat-affected zone of 2205 steel had been increased. There occurred an epitaxial growth at the interface of 2209/2205, and there were a fine transition zone and Type II boundaries at the interface of the X80 steel. The thermal cycle results determined that the temperature peak in the areas away from the fusion line had increased by increasing the heat input and that the heat-affected zone of the two base metals, particularly the X80 steel, had been extended further.

KEYWORDS: Austenite; Heat affected zone; Microstructure; Steel; Thermal cycles; Transition zone; Welding

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RESUMEN: Efecto de la corriente aplicada en la microestructura y el cordón de soldadura en la unión de los Aceros AP15L X80/DSS 2205 utilizando ciclos térmicos. En esta investigación, se ha investigado el efecto de la entrada de calor durante el proceso de soldadura por arco de metal blindado (SMAW) sobre la microestructura y la zona afectada por el calor en la unión disimilar de los aceros API 5L X80/DS5 2205, mediante el registro de los ciclos térmicos a tra-

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vés de la implantación de un termopar en la dirección perpendicular de la línea de soldadura. El metal de aporte utilizado (electrodo) es DSS 2209. Se ha estudiado la microestructura de los metales base y el de soldadura y sus interfaces, para diferentes aportes de calor, combinando microscopía electrónica de barrido, microanálisis de espectroscopía de dispersión de energía (SEM/EDS) y microscopía óptica (OM). Los resultados indicaron que la interfase entre los metales base y el metal de soldadura tiene una excelente consistencia y no hay evidencia de grietas para los diferentes aportes de calor. Al aumentar la entrada de calor, se ha observado que se incrementa la cantidad de austenita secundaria en el metal de soldadura y en la zona afectada por el calor del acero 2205. Hubo un crecimiento epitaxial en la interfaz de 2209/2205, y hubo una zona de transición fina y límites Tipo II en la interfaz de 2209/API 5L X80. En la zona afectada por el calor del acero X80 se detectaron áreas que contenían granos gruesos, finos y parcialmente finos. Los resultados del ciclo térmico determinaron que el pico de temperatura en las áreas alejadas de la línea de fusión habían aumentado al aumentar la entrada de calor y que la zona afectada por el calor de los dos metales base, particularmente el acero X80, se había extendido aún más.

PALABRAS CLAVE: Acero; Austenita; Ciclos térmicos; Microestructura; Soldadura; Zona afectada por el calor; Zona de transición

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1. INTRODUCTION

API (American Petroleum Institute) steels are high-strength low alloy steels (HSLA), which have proper mechanical properties (strength, ductility, and hardness) because of the micro-alloving elements added to the composition and applied thermo-mechanical process (Barbaro et al., 2002). API 5L X80 (X80) steel is micro-alloyed steel with high strength. This steel due to its high toughness and the possibility of producing pipes with higher diameters and lower thickness compared to the previous grades (such as API, X70, and X65), causes the reducing costs and increases transmission efficiency. This steel is extensively employed in oil and gas transmission pipelines, marine installations, extraction risers, hydrocarbons transportation, ships, and submarines (Aydin and Nelson, 2013). HSLA steel pipes such as X80 are corroded in humid and sultry environments such as areas near the shores and rivers due to environmental corrosion, as well as cyclic drying and wetting so there can be a continuous problem with the corrosion of these pipes in these areas (Ren et al., 2009). The corrosion of these alloys can result in safety hazards and environmental pollution caused by the oil and gas leakage and increase the maintenance costs (Han et al., 2012). Austenitic stainless steels like 304L or 316L tend to be employed as a replacement for HSLA steels in critical areas such as coasts. Austenitic stainless steels are also ordinarily sensitive to stress corrosion cracking (SCC) in chloride and marine environments. Nowadays, duplex stainless steels such as DSS2205 (2205) are good options instead of austenitic stainless steel. They are extensively used due to their good mechanical properties and excellent corrosion resistance, particularly SCC in chloride and marine environments. Thus, duplex stainless steels are considered significant alternatives to HSLA steels in critical areas such as coasts, as well as wet and chloride environments (Wang *et al.*, 2012). Therefore, the dissimilar welding of HSLA steels like X80 and duplex stainless steels such as DSS 2205 is necessary and can be widely utilized.

The solidification microstructure of the weld metal and grain size has great effects on the mechanical properties, corrosion behavior, and sensitivity to weld hot cracking (Kou, 2003). There are various metallurgical phenomena such as dilution, unmixed zone, transition zone, and grain growth, particularly in the dissimilar weld joints, which depend thoroughly upon the heat input and the type of welding process (Lipold and Kotecki, 2005; Wang et al., 2011). In the dissimilar joints of stainless steel to carbon steel and low alloy steels, a transition zone is formed between the weld metal and the base metal. Depending on the heat input and welding parameters, they may be wide or narrow and their properties differ from the adjacent zones (Lipold and Kotecki, 2005). Studies have explained that the formation of the transition zone causes the Type II boundaries to be created in these joints (Srinivasan et al., 2006a; Srinivasan et al., 2006b). These boundaries are created due to the crystal structure difference and the change in the initial solidification behavior to account for the change in the chemical composition gradients in the direction perpendicular to the melting boundary (Wang et al., 2009; Wang et al., 2012). HAZ of the HSLA steels tends to consist of three areas: supercritical (AC3 to the melting point), inter-critical (AC1-AC3), and subcritical (below AC1), respectively. The supercritical area is composed of both the coarse-grained zone (1100 °C to the melting point) and the sub-grains zone (1100-AC3 °C), whose microstructures influence the weld properties. In these steels, the HAZ size is a criterion for the microstructural changes, which are frequently controlled by the heat input and welding process variables (Farhat, 2007). Among the HAZ parts, the coarse-grained zone (CGHAZ) endures the most severe thermal cycles and has the highest hardenability due to the coarse grain size. The fine-grained HAZ of these steels is sensitive to softening, and the hardness of this area decreases by increasing the

heat input (Kim et al., 2005; Farhat, 2007). The heat input is one of the most important parameters affecting the HAZ and weld metal microstructure, mechanical properties, and corrosion behavior in the duplex stainless steel (Yousefieh et al., 2012). The high heat input reduces the cooling rate and forms the secondary austenite and undesirable phases such as the sigma phase. The low heat input due to the high cooling rate slows down the ferrite to austenite transformation favoring the presence of ferrite in the final microstructure. By decreasing the temperature during the cooling below ferrite solvus, the excess nitrogen in the ferrite in solid solution tries to escape. Some of the nitrogen diffuses to the austenite and encourages its further formation, but due to the high cooling rate during the welding, all of the nitrogen in solid solution cannot diffuse to the austenite. Hence, the rest of the nitrogen can deposit as an undesirable phase such as chromium nitride (Cr_N) in the ferrite matrix. The formation of this chromium- nitride increases brittleness and reduces the corrosion resistance in HAZ of these steels (Lipold and Kotecki, 2005; Yousefieh et al., 2011).

In this research, the effect of the shielded metal arc welding (SMAW) process heat input on the microstructure of different zones of dissimilar weld joints of API 5LX80 / DSS2205 and HAZ expansion caused by various thermal cycles using thermocouple implantation has been studied.

2. EXPERIMENTAL PROCEDURE

In this research, API 5L X80 and DSS 2205 duplex stainless steel sheets with the dimensions of $100 \times 50 \times 8$ were used as base metals. The joint

design was prepared as a single V-groove 70° with a 1 mm root face. The samples were welded using the SMAW process by controlling the interpass temperature up to about 150 °C (to reduce distortion and stress) at three different heat inputs of 0.54, 0.70, and 0.81 KJ·mm⁻¹ by the E2209 electrode. Table 1 shows the chemical composition of base and filler metals, and Table 2 shows the welding parameters. The heat input of the welding has been calculated by the Eq. (1) (Lipold and Kotecki, 2005):

$$H.I=\eta \frac{V.I}{1000S}$$
(1)

where the V, I, S, and η are the welding voltage, welding current, welding speed, and arc efficiency, respectively

In this study, 10 K-type thermocouples were placed at equal distances perpendicularly to the joint edge of the two base metals (5 thermocouples each) to measure the temperature at different distances from the weld line and record the thermal cycles. The holes with a 1.5 mm diameter and a depth of 6 mm at a 1 mm distance from the upper edge and 0.5 mm from each other were drilled behind the samples to introduce the thermocouples. Figure 1 shows the schematic of the holes' position and location of planting the thermocouples and the scheme of the joint design. A 16-channel Field Logger data logger made by Venus Co. has been used to record the temperature-time changes.

For the microstructural studies, at first, the proper samples from the weldments including the weld metal, the interfaces, and the two base metals were separated.

Base С Cr Ni Р Ν V Ti Nb Cu Mo Mn S Si **UNS S31803** 0.03 22 5.8 3 2 0.02 0.03 0.79 0.14 _ (DSS 2205) 0.07 0.02 0.009 API 5L X80 0.25 0.29 1.8 0.001 0.007 0.25 0.004 0.003 0.01 0.03 AWS E2209-17 0.03 22.6 9 3 0.9 0.16 1 _ _ -_ --

TABLE 1. Chemical composition of base and filler metals (wt. %) with Fe to balance

TABLE 2. Welding parameters								
C 1-	Electrode	Electrode	A	V-14	Welding	Average Heat		
Sample	Туре	Dia.	Ampere	voltage	Speed	Input		
No.	(AWS)	(mm)	(A)	(V)	(mm·s ⁻¹)	(kJ·mm ⁻¹)		
1	E2209	2.5	75	20	2.24	0.54		
2	E2209	2.5	90	20	2.17	0.7		
3	E2209	2.5	125	20	2.45	0.81		



FIGURE 1. Schematic of the joint design and thermocouples location.

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Material	Etching Solution + Chemical	Immersion Time (s)	
	Composition	ОМ	SEM
	Marble	0	40
DSS 2205 +	10grCuS0 ₄ +50cc HCl+50cc H ₂ O	8	40
Weld Metal	Beraha	2	
	$1 \text{grK}_2 \text{S}_2 \text{O}_5 + 40 \text{cc} \text{Hcl} + 80 \text{cc} \text{H}_2 \text{O}$	3	-
A DI 51 V00	Nital 2%	12%	
AFI 5L X80	98cc+Ethanol+2cc HNO ₃	15 45	

Afterwards, the entire samples were polished with 60 to 2000 sandpaper using a diamond paste of 1 μ m and then 0.25 μ m after sanding. Subsequently, the polished samples were etched using the solutions, and etching conditions described in Table 3. The microstructure of different zones of the joints was investigated using an optical microscope and a scanning electron microscope equipped with an EDS point analysis system. To measure the amount of ferrite in the base and weld metals, a feritscope device "Feritscope FMP30" equipment from Helmut-Fischer was used. The Vickers micro-hardness test was performed on the weldments under a load of 0.1 Kg and a time of 10 seconds to obtain the hardness profile across the joint zone.

3. RESULTS AND DISCUSSION

3.1. Microstructure of base metals

Figure 2 shows the images taken using optical microscopy (OM) and scanning electron microscopy (SEM) of the base metals microstructures. Figure 2a reveals that the microstructure of the 2205 steel includes a ferritic matrix (dark) with austenitic grains of different morphologies (bright yellow). It expects that austenite is seen white color under an optical microscope, but the yellow color that is observed in images can be due to the light intensity of the microscope and the method of taking photos The image shows the presence of simi-



FIGURE 2. Microstructure of the base metals; a) OM image of 2205, b) SEM image of 2205, c) OM image of X80, and d) SEM image of X80. The microstructures corresponding to images a) and b) have been etched with Beraha solution and those corresponding to images c) and d) with Nital solution.

lar amounts of ferrite and austenite. These have been measured using the ferrite scope (51.1% ferrite - 48.9% austenite). The solidification of duplex stainless steels, depending on the ratio of Cr _/Ni_{eq}, can be fully ferritic or ferritic-austenitic. Based on the pseudo-binary diagrams of iron and chromium (Ghasemi Banadkouki and Dunne, 2006), if the ratio of Cr _q/Ni _>1.85, the solidification would be fully ferritic (mode F). According to Table 1, and considering the WRC-1992 diagram equations (Cr _eq Cr + Mo + 0.7Nb and Ni _= Ni + 35C + 20 N + 0.25 Cu) the ratio Cr _q/Ni _steel follows a fully ferritic mode. The transformation sequence for these duplex stainless steels would be as follows:

$$L \rightarrow L + F \rightarrow F \rightarrow F + A$$

Solidification of these steels starts with the ferrite formation and no austenite forms until the end of solidification. By decreasing the temperature below the ferrite solvus, the transformation of ferrite to austenite begins. First austenite nucleates at the ferrite grain boundaries and with further cooling it starts growing at the expense of the ferrite matrix. Though the solubility of N in ferrite is greater than that of carbon, the nominal nitrogen content is still greater than the maximum solubility value (Stein *et al.*, 2013). Therefore, the diffusion of N is expected to happen to the austenite, stabilizing this phase. The SEM image in Figure 2b shows the distribution of austenite and ferrite phases in the microstructure after the solidification.

Figure 2 (c-d) display the OM and SEM images taken from the X80 steel. The figure shows that the predominant microstructure contains acicular ferrite and granular bainite (ferrite grains with Austenite-Martensite (M/A) constituents). This alloy microstructure is a function of the chemical composition, production process, and thermo-mechanical treatment, performed during the processing of the steel. Rapid cooling and lack of full austenite decomposition in the final stages of the X80 steel production lead to having stable retained austenite zones. In addition to rapid cooling, carbon enrichment of the austenite due to carbon diffusion away from the ferrite increases the stability of the austenite, which promotes the presence of M/A constituents (Ghasemi Banadkouki and Dunne, 2006). Figure 2c shows the microstructure of the X80 steel in more detail in the SEM image. In this image, acicular ferrites, granular bainite grains, and M/A constituents are clearly visible. This type of microstructure with mostly granular morphology and M/A constituents resembles that of granular Bainite (Zajac et al., 2009). It has been reported that the precipitation of niobium-titanium carbonitrides and vanadium carbides may occur in the microstructure during the processing of this steel (Sarkar and Panda, 2012). However, a detailed transmission electron microscopy analysis of the microstructure should be carried out to detect these precipitates which generally have sized on the nanometer scale. This characterization is out of the scope of this investigation.



FIGURE 3. Optical images of weld metal (E2209) microstructure at the heat inputs of: a) 0.54 kJ·mm⁻¹, b) 0.70 kJ·mm⁻¹ and c) 0.81 kJ·mm⁻¹. The microstructures were etched with marble solution.

3.2. Weld metal microstructure

Figure 3 shows the microstructure of the E2209 weld metal at different heat inputs. The figure discloses that the final structure of the weld metal consists of ferrite (dark) and austenite (bright yellow). It expects that austenite is seen white color under an optical microscope, but the yellow color that is observed in images can be due to the light intensity of the microscope and the method of taking photos. The weld metal solidification starts with the formation of the ferrite phase and until the end of solidification only ferrite is stable ($Cr_{eq}/Ni_{eq}=1.93$). The transformation to austenite begins when the temperature is decreased under the ferrite solvus. First, austenite nucleates at the ferrite grain boundaries and grows along the grain boundaries. As the cooling continues, extra nitrogen diffuses from ferrite toward the austenite and encourages the austenite formation, as well as causes precipitation of the chromium nitrides in the ferrite grains (Lipold and Kotecki, 2005). New austenite grains grow similarly to a Widmanstatten-like structure (similar to the ferrite Widmanstatten in terms of appearance and growth mechanism) from the ferrite boundaries toward the center of the ferrite grains. Intergranular nucleation, also observed within ferrite grains often takes place at nitride precipitates or at the interface between Widmanstatten austenite plates and the ferrite matrix. Comparing the images in Fig. 3 indicates clearly that the amount of weld metal austenite (yellow phase) has increased by increasing the heat input and this amount is also greater than for the base metal 2205 (Fig. 2a). The Feritescope results (vol.% of ferrite and austenite) provided in Table 4 also demonstrate an increase in the amount of austenite in the weld metal by increasing the heat input.

There is a great difference between the values of ferrite and austenite in the microstructure of weld metal (Table 4) compared to base metal 2205 (closely equivalent ratio of ferrite and austenite) for the following reasons. First, the amount of nickel which is an austenite phase stabilizer, higher in the E2209 weld electrode compared to the DSS 2205 base metal (Table 1); hence, higher amounts of nickel cause the stability and increase in the amount of austenite in the weld metal (2209). Second, the weld metal

TABLE 4. Ferrite scope results of weld metal at different heat inputs

Heat Input	FN (Ferrite Number)	Vol% Ferrite (FN×0.7)	Vol% Austenite
0.54 kJ·mm ⁻¹	55.0 ± 2.8	38.5 ± 1.9	61.5 ± 3.1
0.70 kJ·mm⁻¹	49.0 ± 2.4	34.3 ± 1.7	65.7 ± 3.3
0.81 kJ·mm ⁻¹	45.0 ± 2.2	31.5 ± 1.6	68.5 ± 3.3



FIGURE 4. Optical images of the microstructure found in the heat affected zone (HAZ) and the interfaces between the base (2205) and the weld metal (2209) at heat inputs of a) 0.54 kJ·mm⁻¹, b) 0.70 kJ·mm⁻¹, c) 0.84 kJ·mm⁻¹. The microstructures were etched with Marbel solution.

cooling rate during the solidification decreases by increasing the heat input, because more time is provided to the steel during the transformation of ferrite to austenite below the ferrite solvus. This increases the vol.% of austenite in the final microstructure. As it can be observed in Fig. 3, the amount of the transgranular austenite has increased by increasing the heat input and the Widmanstätten-like austenite has decreased. The increase in the heat input causes a reduction in the cooling rate and an increase in the time available for ferrite-to-austenite transformation, thereby creating more opportunities and proper conditions to form the austenite. Due to the complete occupy of the ferrite boundaries by inter-

granular austenite, new austenite grains within ferrite grains start to nucleate and grow (intragranular austenite). In additioni, with the reduction of the cooling rate, the probability of Widmanstätten-like austenite formation decreases.

3.3. Heat affected zone microstructure

The microstructure of the HAZ and interface between the 2205 base metal and the E22209 weld metal at various heat inputs has been presented in Fig. 4. The figure reveals that the joint between the weld metal and the 2205 steel is continuous at all heat inputs. There is no evidence of discontinuity, such as transition zones, unmixed zone, and cracks. At this interface, the epitaxial growth is explicitly visible at all heat inputs. When the base and weld metals have the same crystal structure and similar chemical composition, the growth mechanism will experience an epitaxial growth near the fusion line. In this case, the base metal grains connected to the fusion line operate as the primary nucleus, and the weld pool due to the direct contact with these grains starts to solidify without the need to nucleate. Then, the formed solid grows inside the weld pool without any problems. The epitaxial growth has also been recognized in dissimilar weld joints (Kou, 2003).

It is obvious from Fig. 4 that the HAZ microstructure of the 2205 steel contains large ferrite grains. The austenitic grains have nucleated at the ferrite grain boundaries covering them and have also started growing similar to Widmanstätten plates toward the center of the grains. In addition, there are some amounts of intragranular austenite (the secondary austenite) within the ferrite grains. The coarse-grained zone near the fusion line is the result of reaching temperatures above the ferrite solvus (the single-phase ferrite zone) during the welding thermal cycles. In this zone, the austenite phase, and the precipitates have been dissolved, and ferrite grains grow without any significant obstruction. After reaching the peak temperature, the microstructure cools quickly; without giving the alloy enough time for the ferrite-to-austenite transformation to take place. Therefore, it is reasonable that in the HAZ an increase in the ferrite volume fraction takes place compared to the volume fraction of austenite. As the temperature is decreased during the cooling below ferrite solvus, the excess nitrogen in the solid solution in the ferrite tries to scape. Some of the nitrogen diffuses to the austenite and encourages its further formation. The rest of the nitrogen forms as chromium nitride precipitates in the ferrite matrix. These precipitates are suitable places for the nucle-



FIGURE 5. Welding thermal cycles of 2205 base metal heat affected zone at different heat inputs of a) 0.54 kJ·mm⁻¹, b) 0.70 kJ·mm⁻¹, c) 0.81 kJ·mm⁻¹.

ation of the austenite inside ferrite grains. In addition, the formation of chromium nitride decreases the chromium in solid solution in the matrix reducing the Creq / Nieq ratio and favoring the austenite formation conditions inside the ferrite grains (Jana, 1992; Lipold and Kotecki, 2005).

Comparing Fig. 4 (a-c) also indicates that the amount of austenite in HAZ has been increased by increasing the heat input. The cooling rate is decreased by increasing the heat input and more time is provided for ferrite-to-austenite transformation, which increases the amount of austenite in the HAZ. As it is shown in these images, the HAZ has become wider and the width of this zone has been increased by increasing the heat input. Because the temperature increase extends at distances further away from the fusion line. The effect of the heat input on the temperature distribution in the HAZ and the expansion of this area by the influence of the thermal cycles have been shown in Fig. 5. These plots reveal that the temperatures peak reached increases as the fusion line is approached. The duration of exposure to these temperatures has been longer than that in the areas far away from the fusion line. Hence, the dissolution of the precipitates and the austenite is more pronounced, and the ferrite grains have grown significantly. In the zones further away from the fusion line, where the temperature has not been increased sufficiently, the precipitates dissolution and retransformation of the austenite phase have been less pronounced and, the ferrite grains have not grown. It has also been observed that the temperature peak at the distances away from the fusion line has been increased by increasing the heat input. These areas have experienced higher temperatures; consequently, while the precipitates and the austenites volume fraction decreases at distances away from the fusion line the ferrite grains become larger, and the width of HAZ also increases, which correlates well with the microstructures shown in Fig. 4. It should be mentioned that the comparatively low heat conductivity of the 2205 steel has made the HAZ of this steel narrower.

Figure 6 shows the HAZ microstructure of the X80 steel at various heat inputs. This figure indicates that the interface of the X80 steel with the weld metal is absolutely continuous and free from any cracks. The figure shows that there is a narrow transition zone near the fusion at all heat inputs. The presence of this transition zone can be due to the difference in the crystal structure of the weld metal and base metal, the concentration gradient at the fusion boundary, and the high growth rate (high solidification rate) during welding (Lipold and Kotecki, 2005). The type II boundaries can form at a distance of a few microns from the fusion line within the weld metals. These boundaries are parallel to the fusion boundary and separate the weld metal from the transition



FIGURE 6. Optical and SEM images of interfaces of X80/2209 at different heat inputs, a) and b) $0.54 \text{ kJ} \cdot \text{mm}^{-1}$, c) $0.7 \text{ kJ} \cdot \text{mm}^{-1}$, d) $0.84 \text{ kJ} \cdot \text{mm}^{-1}$. The microstructures in the images a-c have been etched with Marble and with Nital solutions.

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area. The Type II boundaries do not generally exist under normal solidification conditions and are frequently formed where the base metal is in the austenite phase stability temperature range (Nelson et al., 1999; Nelson et al., 2000). Comparing Fig. 6 (a-c), and Fig. 6d indicates that the transition zone width has become slightly wider with an increase in the heat input. The dilution of the base metal in the weld metal has increased by increasing the heat input, thereby expanding the width of the partially melted zone. In addition, by increasing the heat input, the maximum temperature in the HAZ has increased and caused an increasing stability duration of the austenite-austenite interface, thereby decreasing the temperature gradient. Hence, the grain boundaries' migration is also raised, and the transition zone becomes wider.

Figure 7 clearly indicates that increasing the heat input leads to the HAZ development and grain coarsening in this area, which could be related to the higher heat conductivity of the X80 steel compared to the 2205 duplex stainless steel. From the images, the three zones with different grain sizes: the coarse grain, fine grain, and partial fine grain (nearly similar to the base metal) can be recognized. These zones have been shown separately and more clearly in Fig. 7. The partial fine-grained zone is near the base metal and after it, there is the fine-grained area that has been formed by normalizing (cooling the weld joints in the air) during welding. The coarse-grained area is adjacent to the fusion line which has been extended to the fine-grained area. The grain growth in this area is caused by the re-heating undergone by the samples during the different weld passes. The coarsegrained zone is heated to a temperature higher than AC3 due to the welding heat, the austenite is formed at this temperature, and the precipitates could be dissolved depending on their solubility product. Accordingly, there would be no barrier to preventing the austenite grain growth. The transformation of austenite to ferrite has occurred during the subsequent cooling of this area and the final microstructure is in the form of large ferrite grains, which are likely to form as widmanstätten ferrite due to the high cooling rate in this zone. The fine-grained zone undergoes lower temperatures (below AC3). In this zone, precipitates do not play any relevant role, the activation energy is not high enough and the hard impingement by the presence of ferrite has a more pronounced influence on the mobility of the grain boundaries during cooling. When the temperature decreases the austenite present in the microstructure transforms to ferrite. The ferrite morphology does not change much, which is relatively similar to the X80 base metal.

SEM-EDS linear analysis was performed along the width of the interface between the base metals and the weld metal. At the interface of 2209 / X80, a



FIGURE 7. Heat affected zone microstructure of the X80 steel with more clarity (Fig. 6); a) coarse grain zone, b) finegrained zone and c) partially fine-grained zone. The microstructures in all images have been etched with Nital solution.

chemical composition gradient was obvious (Fig. 8). Figure 8a shows that in a thin area of about 10 μ m in thickness, there is a chemical compositional gradient from the X80 base metal to the 2209 weld metal, caused by the dilution of the X80 base metal. The difference between the melting point and the heat conductivity of the X80 and weld metal does not allow for their sufficient flow and perfect mixture. Thus, the weld metal melts and solidifies without experiencing much dilution (slight mixing of the base and weld metals). Figure 8 reveals that the thickness of the transition zone increases (from 10 to about 17 μ m) by an increase in the heat input, which shows



FIGURE 8. EDS results in the X80/2209 interfaces at a heat input of a) 0.54 kJ·mm⁻¹, b) 0.70 kJ·mm⁻¹and c) 0.84 kJ·mm⁻¹.

good correspondence with the results achieved by the microstructural studies of these interfaces. Other researchers have reported similar results concerning the transition zone development by a rise in the heat input (Wang *et al.*, 2012; Han *et al.*, 2012; Srinivasan *et al.*, 2006b).

Figure 9 shows the results achieved by studying the effect of the heat input on the microstructure and HAZ expansion in the X80 steel using thermal cycles. The figure shows that the temperature peak in the HAZ area increases by an increase in the heat input. Each thermocouple experiences higher temperatures for a longer period of time, which extends the width of the heat-affected zone, and the grains become coarser in this zone. The figure shows that the areas further away from the fusion line are also exposed to higher temperatures by increasing the heat transferred to HAZ due to the higher heat conductivity of the X80 steel; thus the width of HAZ increases. Furthermore, the exposure time of the coarse-grained zone in the austenitic single-phase area (Fig. 9) increases by increasing the heat input and provides more time for the austenite grains to grow. The part of the base metal that is exposed to the recrystallization temperature expands due to the higher heat input; accordingly, the width of the fine-grained zone raises by elevating the heat input. The results of the temperature recording during the welding thermal cycles show that temperature changes and the HAZ expansion agree with the microstructural observations.

Figure 10 shows the hardness profile of the X80 / 2205 joint. The figure depicts that the weld metal hardness is relatively uniform except near the fusion



Figure 10. Hardness distribution in different areas of X80/2209/2205 joint at various heat inputs of a) 0.54 kJ·mm⁻¹, b) 0.70 kJ·mm⁻¹, and c) 0.81 kJ·mm⁻¹.



FIGURE 9. Welding thermal cycles of X80 base metal heat affected zone at different heat inputs of a) 0.54 kJ·mm⁻¹, b) 0.70 kJ·mm⁻¹ and c) 0.81 kJ·mm⁻¹.

boundary and has been augmented by enhancing the heat input. The transformation of ferrite to austenite has been elevated by increasing the heat input due to increasing the temperature and decreasing the cooling rate; hence, an increase in the austenite amount in the weld metal causes the hardness to be increased slightly. More changes have been made in the hardness near the fusion lines due to increasing the base metal dilution. By increasing the heat input due to decreasing in cooling rate and increases in the intragranular austenite, hardness near the fusion line in the HAZ of the 2205 steel has increased. The wide hardness variations in the HAZ of the X80 steel are due to the microstructural changes and the formation of a multi-region HAZ (The partial finegrained, the fine-grained, and the coarse-grained zones). The hardness of the coarse grain zone near the fusion line is decreased due to the grain coarsening and a change in the ferrite morphology from acicular to the allotriomorphic. A softening takes place due to the partial recrystallization and new grain nucleation, which decreases hardness in the fine-grain zone. In the partial fine-grain zone, the hardness has not changed much due to insignificant structural changes compared to the base metal. The variation of hardness in different joint zones (Fig. 10) shows good correspondence with the results achieved by thermal cycles and microstructural studies.

4. CONCLUSION

- The DSS 2209 weld metal has a two-phase ferrite-austenite microstructure in which there is a ferrite matrix with the grain boundary austenite, widmanstätten-like austenite, and transgranular austenite. A rise in the heat input from to increases the amount of austenite (especially the transgranular type).
- The interfaces of the base and weld metals are continuous and lack any cracks at all the heat inputs. There are no defects or discontinuities at the interface of 2209/2205, while a transition zone and Type II boundaries are observed at the X80/2209 interface. The width of the transition zone is increased from 10 to 17 microns by augmenting the heat input.
- The HAZ of 2205 steel consists of a twophase ferrite-austenite matrix in which there has formed secondary austenite. Three coarse-grained, fine-grained, and partial finegrained zones were observed in the HAZ of the X80 steel. More secondary austenite was observed in HAZ of 2205 steel by increasing the heat input.
- The results achieved by thermal cycles indicate that the temperature peak is increased by increasing the heat input and that the areas

far away from the fusion line undergo higher temperatures due to their exposure to high temperatures for a longer period of time. It was distinguished that the grains become coarser and that the heat-affected zone becomes wider.

The results of hardness changes in the cross-section of the joint reveals good correspondence with the microstructural studies and the results obtained by thermal cycles.

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