

# Mechanical Properties of UNS S31803 (2205) Duplex SS Welds Deposited with GTAW and SMAW Methods

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**Abstract:** Duplex stainless steels have superior mechanical and corrosion properties with better toughness and ductility compared to austenitic and ferritic stainless steels due to their dual phase microstructure i.e. austenite and ferrite, which is attractive for gas pipelines. In this study, effects of different heat inputs and interpass temperatures were investigated on the mechanical and microstructural properties of UNS-S31803-(2205) duplex stainless steel welds using two different welding methods i.e. Gas Tungsten Arc Welding and Shielded Metal Arc Welding in place of conventionally used Submerged Arc Welding. UNS-S31803-(2205) duplex stainless steel pipe was joined using ER2209 (GTAW) filler wire and ESAB OK67.55 (SMAW) electrode. Pure argon was used as shielding gas and for back-root protection. Microstructural examinations and mechanical tests were carried out in accordance with PTS and ASME standards. Results showed that, two different welding methods generated a weld joint with higher strength and hardness than base metal at certain heat inputs and interpass temperatures. Weld metal properties were within the defined limits of standards and the critically selected interpass temperature and heat input which induced different cooling conditions with multiple passes produced a welded joint without the formation of deleterious sigma phase in weld metal and the heat affected zone.

**Key-words:** Duplex stainless steel; UNS S31803 (2205); GTAW; SMAW.

## 1. Introduction

Excellent corrosion resistance at low and high temperatures is one of the important factors for choosing stainless steels which are classified according to the phases in their microstructure; austenitic, martensitic, ferritic, duplex and precipitation hardening stainless steels. Among these, duplex stainless steels (with equal proportions of ferrite + austenite) have become more important in recent years. The austenite phase in these steels provides corrosion resistance and ductility, while the ferrite phase provides mechanical strength and weldability [1,2]. Due to their superior properties, duplex stainless steels has many industrial applications such as chemical, petrochemical, power, communication, gas and oil distribution pipes [3]. Offshore applications are also particularly suitable for welding duplex pipes. The most important advantage of duplex stainless steels over other types of stainless steel is that their high mechanical properties allow the use of thinner-section parts in the place of other stainless steel types. Their limitations are the reduction in resistance to pitting corrosion due to the ferritic structure formed in the heat affected zone (HAZ) after welding and the limitation of the use at temperatures between 260-300 °C due to brittleness caused by thermal ageing [4].

Duplex stainless steels are highly susceptible for the precipitation of intermetallic phases due to their fast diffusion rate in the ferrite phase and high content of alloying elements. When these steels are exposed to temperature changes between 600-1000 °C, they become sensitive to the formation of secondary intermetallic phases [5-7]. The formation of secondary intermetallic phases disrupts the ferrite-austenite balance, adversely affecting its mechanical and corrosion properties of duplex weld metal [8]. The most important secondary phase that limits the high temperature application of duplex stainless steels is the sigma ( $\sigma$ ) phase. This phase occurs between 600-950 °C and dissolves above 1000 °C. The sigma phase is an intermetallic phase of Fe-Cr-Ni-Mo with tetragonal structure [9]. If cooling at temperatures between about 1000-500 °C is carried out slowly, the sigma phase will eventually form, which could be avoided by cooling fast enough to pass through this temperature range. The elements that promote the formation of the  $\sigma$  phase are Mo and Cr. The  $\sigma$  phase preferentially occurs at the ferrite/austenite interface [6-10]. The  $\sigma$  phase particularly reduces the corrosion resistance, toughness and tensile ductility values of duplex stainless steels [11].

Gas Tungsten Arc Welding (GTAW) and Shielded Metal Arc Welding (SMAW) are among the joining methods used in the welding of duplex stainless steels. Although the weldability of duplex stainless steels is excellent, precautions should be taken to prevent significant changes in the microstructure of the heat-affected zone (HAZ) and the weld metal (WM) relative to the base metal (BM). This leads to changes in the phase balance, separation of intermetallic elements, grain growth, alloying or additive elements, or other reactions [12,13]. The weldability of duplex stainless steels is generally better than that of ferritic stainless steels but worse than that of austenitic stainless steels. A suitable ferrite/austenite balance must be achieved in the weld metal. The amount of ferrite should be

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expected to increase in the HAZ. In this respect, very low heat input and consequent rapid cooling should be avoided and the formation of the austenite phase in the HAZ should be allowed [11]. Very high cooling rates will result in a high ferrite content, nitride precipitation and ultimately low toughness and low corrosion resistance. Welding of duplex stainless steels should be carried out with controlled heat input depending on the alloy content [10,11]. Orbital GTAW welding of duplex stainless steel pipes were investigated by Sönmez et al. [11] using two different electrodes and three different welding parameters. In their specimens, austenite formation at ferrite grain boundaries, Widmanstätten plates and austenite formation within ferrite grains were observed in the weld metal. 2205 duplex stainless steel specimens were joined using GTAW using pure argon shielding gas by Taban [14] to investigate the effects of different welding cooling conditions on microstructural changes. In the natural cooling process, it was observed that the weld metal consists of plate Widmanstätten with columnar ferrite grain boundaries forming the matrix and austenite grains formed within the grain. Another important aspect of the weld metal microstructure is that the approximately equal ferrite-austenite balance in the base metal changes with the rate of cooling during or after welding. 1.4462 duplex stainless steels (DSS) with a thickness of 6.8 mm were welded by plasma arc, GTAW and plasma arc + GTAW [15]. Impact resistance tests were carried out and microstructures showed that all welds exhibited good toughness properties at low temperatures such as -60 °C. Ferrite content measurements and EDX analysis to determine the compositional variations between phases at weld metal (WM) revealed that the ferrite content varied within an acceptable range depending on the heat input which affects the microstructure and hence the toughness properties. Heat input affects the ferrite-austenite balance in the WM and HAZ, which affects the properties of the welding process. If good toughness properties are desired, the phase balance should be carefully controlled by the heat input range. 2205 grade duplex stainless steel plates were welded by GTAW and carbon dioxide laser beam welding (LBW) to compare their properties [16]. The size and microstructure of the fusion zone (FZ) were investigated for their mechanical and corrosion properties. Compared to GTAW, LBW produced welded joint with a significant reduction in FZ size and an acceptable weld profile. GTAW resulted in a ferrite-austenite balance close to that of the base metal due to the higher heat input in GTAW. However, the properties of LBW welded joint, particularly corrosion resistance were better than those of GTAW welded joint. The measured corrosion rates for LBW and GTAW joints are 0.05334 mm/year and 0.2456 mm/year, respectively. This is related to the relatively small size of both WM and HAZ produced in the case of LBW. In other words, properties of welds are significantly influenced by size of fusion zone rather than the austenite-ferrite balance produced in WM. Another study on GTAW of 2205 duplex stainless steel with a double-pass was carried out by Múnez et al. [17]. Their study indicated that a large amount of secondary austenite appeared in the weld metal due to the reheating of second pass. The coarse ferrite grains were formed near the fusion line, while other zones had similar microstructures with different austenite content. It is also found that the microhardness variation was determined by the partitioning of alloying elements (Cr, Mo, Ni and N) and precipitates such as chromium nitride. Electrochemical measurements indicated that the zone containing the fusion line was the most susceptible to pitting attack, followed by the weld metal zone.

In this study, UNS S31803 (2205) duplex stainless steel pipes were joined by GTAW and SMAW methods to investigate the effects of two different welding methods i.e. GTAW and SMAW on the mechanical and microstructural properties of the weld metals in place of conventionally applied submerged arc welding which is completed at one pass with larger HAZ and solidification behaviour.

## 2. Experimental Studies

### 2.1. Material properties and specimen preparation

In this study, the mechanical properties of welded joints welded by GTAW and SMAW were investigated based on ASME Section IX and PTS (Petronas Technical Standards). Ferrite phase percentage measurements, microstructural examinations, Charpy V-notch impact, tensile and flexural tests, hardness measurements and corrosion resistance tests of BM, WM and HAZs were carried out.

UNS S31803 (2205) duplex stainless steel was cast and rolled into plate in accordance with ASTM 240 and then supplied by the manufacturer as welded pipe in accordance with ASTM 928. The welded duplex stainless steel pipe used in the experiments was supplied in two pieces with dimensions of 21.44 × 323.9 × 250 mm. The chemical composition of the duplex stainless steel pipe is given in weight % in Table 1. The Pitting Resistance Equivalent Number PREN for the UNS S31803 (2205) steel is calculated as 38.23 using  $PREN = Cr + 3.3 Mo + 30 N$ , which is within the acceptable limits for duplex stainless steels.

**Table 1.** UNS S31803 (2205) Chemical Composition of Duplex Stainless Steel (PREN: 38.23) in wt. %.

| C      | Mn     | Si     | P      | S      | Cr     | Ni     | Mo     | N      | Cu     |
|--------|--------|--------|--------|--------|--------|--------|--------|--------|--------|
| 0.0180 | 1.7880 | 0.3050 | 0.0270 | 0.0010 | 22.916 | 4.5760 | 2.5120 | 0.1940 | 0.2060 |

ESAB OK TIGROD 2209 brand 2,4x1000 mm GTAW filler metal coded SFA / AWS A5.9 ER2209 (EN ISO 14434-A W 22 8 3 N L) was used for starting welds (root, hot, first fill) in multi-pass joints of duplex stainless steel pipes. This additional weld metal has high homogeneous corrosion resistance and high resistance to pitting and stress corrosion in chloride and hydrogen sulphide environments [18]. The chemical composition of the weld filler wires and consumables are given in Table 2. The mechanical properties of OK TIGROD 2209 filler wire are given in Table 3.

**Table 2.** Chemical composition of OK TIGROD 2209 GTA Welding Filler Wire (wt.%)

| C    | Si   | Mn  | P    | S     | Cr   | Ni  | Mo  | Cu  | N    |
|------|------|-----|------|-------|------|-----|-----|-----|------|
| 0.02 | 0.55 | 1.5 | 0.02 | <0.01 | 22.6 | 8.4 | 3.1 | 0.1 | 0.18 |

**Table 3.** Mechanical Properties of OK TIGROD 2209 GTA Welding Filler Wire.

| $R_{p0.2}$ (MPa) | $R_m$ (MPa) | % Elongation | Toughness (J) (-60 °C) |
|------------------|-------------|--------------|------------------------|
| 600              | 770         | 27           | 70                     |

SMAW method was used for filler and cover passes after GTA welding of root and hot passes. Two different electrodes of ESAB OK 67.55 (2.5x300 mm) and ESAB OK 67.55 (3.3x350 mm) coded SFA/AWS A5.4 E2209-15 (EN ISO 3581-A E 22 9 3 N L B 2 2) were used for welding filler passes. OK 67.55 duplex is a filler welding wire with basic coating for stainless steels. The weld metal produced high ductility at -60 °C with ESAB OK 67.55. The chemical compositions of 2.5 × 300 mm and 3.2 × 350 mm filler wire are given in Table 4.

**Table 4.** OK 67.55 Chemical Content of 2.5 × 300 mm and 3.2 × 350 mm SMAW Filler Wire (wt. %).

| Diameter (mm) | C    | Si   | Mn  | P      | S     | Cr   | Ni  | Mo  | Cu   | N    |
|---------------|------|------|-----|--------|-------|------|-----|-----|------|------|
| 2.5           | 0.04 | 0.65 | 0.9 | 0.0017 | 0.003 | 23.3 | 9   | 3.1 | 0.11 | 0.14 |
| 3.2           | 0.04 | 0.53 | 0.8 | 0.0018 | 0.004 | 23.5 | 9.3 | 3.3 | 0.19 | 0.16 |

The general mechanical properties of OK 67.55 weld filler metal are given in Table 5.

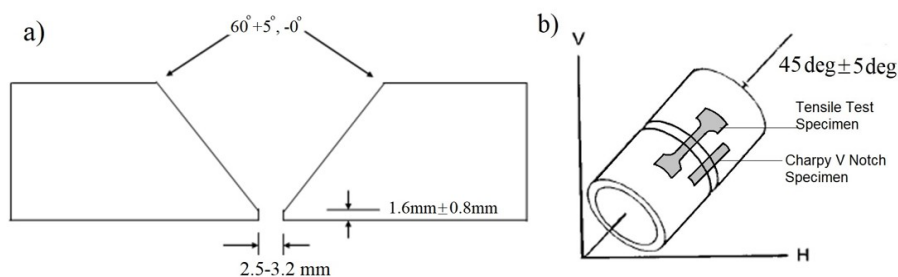
**Table 5.** OK 67.55 Some Mechanical Properties of 2.5 × 300 mm and 3.2 × 350 mm Shielded Metal Arc Welding Wire.

| $R_{p0.2}$ (MPa) | $R_m$ (MPa) | % Elongation | Toughness (J) (-60 °C) |
|------------------|-------------|--------------|------------------------|
| 650              | 800         | 28           | 65                     |

GTAW/SMAW machine (ESAB brand Origo GTAW 3001i model with TA24 control panel) was used to join duplex stainless pipes, generating 300 A current output at 22 V (35% circuit rate), 240 A current 19.6 V (60% duty cycle) and it has the capacity to give 200 A current and 18 V voltage values at 100% duty cycle.

## 2.2. Positioning of the test sample prior the welding process

The UNS S31803 (2205) duplex stainless welded pipe was cut into 21.44 × 323.9 × 250 mm specimens and the weld grooves were machined. The weld grooves are aligned with the help of gauge and scale. Figure 1a shows the detail of the weld groove. It is then fixed to the work table in the 6G position specified in the ASME Section IX. Figure 1b shows the 6G position with the positions of tensile test specimens and Charpy V notch test specimens.

**Figure 1.** (a) Weld Groove; (b) 6G Welding Position.

The worktable was carbon steel and is not normally suitable for contact with duplex stainless steel. However, the contact surface of the specimen is not important as it would be cut out during the extraction of the test specimens.

## 2.3. Joining test specimens and welding parameters

According to PTS 20.191, the oxygen content is required to be 0.05% at the highest volume of weld metal. For this reason, a barrier was formed on both sides of the test specimen and sealed with paper tape on the surface to be welded and

Ar gas was purged with a flow rate of 25 L/min with 99.9% purity, thus providing a sealed root protection. This process continued until the welding is terminated. After the pre-weld preparation, starting from the root pass, one hot pass and one filling pass were performed by GTAW using 99.9% pure Ar shielding gas with a flow rate of 15 L/min using red colour coded WT20 2.4 × 175 mm 2% Thorium alloyed Tungsten electrode. After three welding passes completed with GTAW, the welding process was continued with the SMAW. The weld pass layers and details of weld passes of the joining performed with GTAW and SMAW are shown in Figure 2.

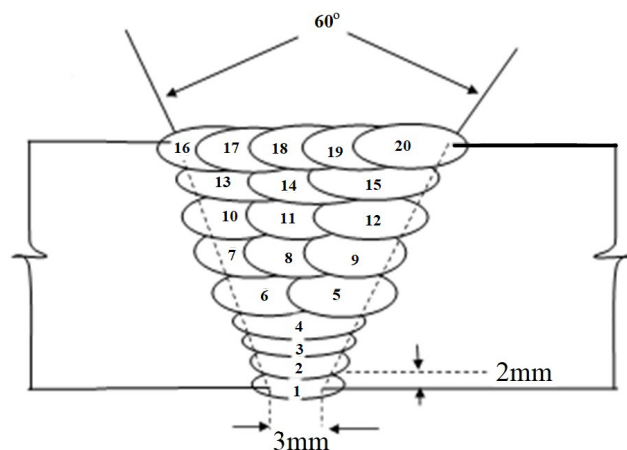


Figure 2. Welding Layers and Passes.

The GTAW and SMAW processes used in the study are shown in Table 6, with parameters such as weld current (A), weld voltage (V), welding time (s) and inter-pass temperature (°C) recorded for each pass of welds. The inter-pass temperatures were strictly followed as per the requirements of the PTS document and therefore the predetermined temperatures were measured using a thermocouple attached to a hand held temperature gauge and waited until the predetermined temperature was reached and the next pass was started. Welding parameters were strictly monitored by the welding supervisor and the welding current, welding voltage and welding speed were recorded and used to calculate the heat input, which is the result of the total energy transferred to the weld zone (kJ/min) divided by the welding speed (mm/min).

Table 6. Welding parameters related to welding method and consumables.

| Weld Layer | Welding Pass | Welding Method | Welding Position | Welding Filler Wire Diameter (mm) | Weld Filler Wire Code | Electrode Polarity | Welding Current (Amperes) | Welding Voltage (V) | Weld Time (sec) | Pipe Diameter (mm) | Advance Rate (mm/min) | Heat Input (kJ/mm) | Temperature Between Passes (°C) |
|------------|--------------|----------------|------------------|-----------------------------------|-----------------------|--------------------|---------------------------|---------------------|-----------------|--------------------|-----------------------|--------------------|---------------------------------|
| 1          | 1            | GTAW           | 6G               | 2.4                               | ESAB OK TIGROD 2209   | DC-                | 100                       | 10.6                | 1480            | 1017               | 41.23                 | 1.08               | -                               |
| 2          | 2            | GTAW           | 6G               | 2.4                               | ESAB OK TIGROD 2209   | DC-                | 150                       | 11.8                | 745             | 1017               | 81.91                 | 0.91               | 75 °C                           |
| 3          | 3            | GTAW           | 6G               | 2.4                               | ESAB OK TIGROD 2209   | DC-                | 150                       | 12                  | 720             | 1017               | 84.75                 | 0.89               | 120 °C                          |
| 4          | 4            | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 85                        | 23.1                | 738             | 1017               | 82.68                 | 1.14               | 85 °C                           |
| 5          | 5            | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 85                        | 24                  | 680             | 1017               | 89.74                 | 1.09               | 110 °C                          |
| 5          | 6            | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 85                        | 23.2                | 640             | 1017               | 95.34                 | 0.99               | 100 °C                          |
| 6          | 7            | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 80                        | 23.9                | 678             | 1017               | 90                    | 1.02               | 80 °C                           |
| 6          | 8            | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 80                        | 24.2                | 686             | 1017               | 88.95                 | 1.04               | 95 °C                           |
| 6          | 9            | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 80                        | 23.4                | 704             | 1017               | 86.68                 | 1.04               | 100 °C                          |
| 7          | 10           | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 83                        | 24.1                | 690             | 1017               | 88.43                 | 1.09               | 115 °C                          |
| 7          | 11           | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 83                        | 23.3                | 710             | 1017               | 85.94                 | 1.08               | 90 °C                           |
| 7          | 12           | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 83                        | 23.4                | 687             | 1017               | 88.82                 | 1.05               | 90 °C                           |
| 8          | 13           | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 92                        | 23.6                | 640             | 1017               | 95.34                 | 1.09               | 110 °C                          |
| 8          | 14           | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 92                        | 24.6                | 632             | 1017               | 96.55                 | 1.13               | 110 °C                          |
| 8          | 15           | SMAW           | 6G               | 3.2                               | ESAB OK 67.55         | DC +               | 79                        | 24.4                | 678             | 1017               | 90                    | 1.03               | 115 °C                          |
| 9          | 16           | SMAW           | 6G               | 2.5                               | ESAB OK 67.55         | DC +               | 79                        | 23.9                | 681             | 1017               | 89.60                 | 1.01               | 120 °C                          |
| 9          | 17           | SMAW           | 6G               | 2.5                               | ESAB OK 67.55         | DC +               | 75                        | 22.4                | 902             | 1017               | 67.65                 | 1.19               | 130 °C                          |
| 9          | 18           | SMAW           | 6G               | 2.5                               | ESAB OK 67.55         | DC +               | 75                        | 22.2                | 895             | 1017               | 68.18                 | 1.17               | 125 °C                          |
| 9          | 19           | SMAW           | 6G               | 2.5                               | ESAB OK 67.55         | DC +               | 75                        | 22.2                | 890             | 1017               | 68.56                 | 1.17               | 110 °C                          |
| 9          | 20           | SMAW           | 6G               | 2.5                               | ESAB OK 67.55         | DC +               | 75                        | 22.6                | 876             | 1017               | 69.66                 | 1.17               | 120 °C                          |

## 2.4. Testing and analysis

### 2.4.1. NDT (Non Destructive Testing)

Radiographic (RT) and liquid penetrant (LP) tests were employed to inspect the welds for defects in joined welds. After cleaning the area to be inspected with cleaner spray (Spotcheck SKC S), the penetrant liquid (Spotcheck SKL-SP1 spray) was applied and left for 20 minutes. Careful intermediate washing was carried out with a cleaner moisten cloth after waiting for the penetrant fluid to diffuse into the surface flaws and later, a developer (Spotcheck SKD S2 spray) was used before the examination of the surfaces. Specimens were prepared in the dimensions and numbers specified in ASME Section IX and PTS standards to examine the microstructures of the specimens, and determine the amount of ferrite and mechanical properties. RT and LP tests revealed no defects in the weld zone.

### 2.4.2. Microstructural study

The specimen with the base metal and the weld zone was removed from the welded pipe in accordance with the ASME Section IX standard, as shown in Figure 3.

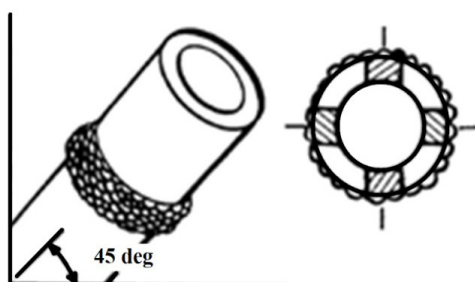


Figure 3. ASME Section IX Sampling Areas for Hardness.

The metallographic samples were ground with 400, 600, 800, 1200G sandpapers, and then polished with alumina oxide paste (1 micron powder size), respectively. The etching process was completed with an etchant containing equal volumes of  $\text{HNO}_3 + \text{HCl} + \text{H}_2\text{O}$  mixture. The microstructures of the samples were examined using Nikon Eclipse MA100 optical microscope and Olympus BX60 optical microscope inverted metal microscopes for 100X and 500X magnifications. LEO VP series electron microscope was used to characterise the fracture surface images in Secondary Electron (SE) and Back Scattered Electron (BSE) mode.

### 2.4.3. Determination of delta ferrite in weld metal and base metal

In accordance with ASTM E562 standard, the  $\delta$  ferrite phase contents were determined by taking five point measurements from each zone of welds, including the right HAZ, weld metal and left HAZ by magnetic method, i.e. measurements from a total of 15 points. The  $\delta$  ferrite phase quantity is determined by the magnetic ferrite tester device i.e. Feritscope FMP30. The instrument can make two different measurements:  $\delta$  ferrite value and ferrite number. The measuring range is 0.1% - 80%  $\delta$  ferrite (0.1-110 FN).

### 2.4.4. Mechanical tests

The effect of the two different welding methods used on the mechanical properties was determined by mechanical tests carried out under four headings, i.e. Charpy V-notch impact test, tensile test, flexure tests and hardness measurements. The toughness characteristics of duplex stainless steel were investigated. Test specimens were cut to ASME Section IX. These specimens were then welded to  $10 \times 10 \times 55$  mm plates and three specimens were taken from each zone according to ASTM A370 and ASTM E23. Following the removal of the specimens, the areas to be tested were divided into five regions in total according to PTS 20.191 and PTS 31.40.20.34 standards; weld centre cap and root, HAZ, HAZ+2 mm and HAZ+5 mm. The 15 specimens were tested in the Instron 450 MPX 3630 notch impact tester at a temperature of  $-50 \pm 1$  °C using inert nitrogen in accordance with the working and manufacturing design.

Two tensile test specimens are prepared in accordance with the ASME Section IX. Two specimens with cross-sectional sizes of  $19.99 \times 19.1$  mm and  $19.96 \times 19 \pm 0.8$  mm were tested on the Instron 5989 L3619 tensile and flexural tester with a rate of 3 mm/min at 20.1°C. The instrument has a capacity of 600 kN and a measurement accuracy of  $\pm 0.5\%$ .

The flexural test specimens were prepared as four point bending specimens. Specimen dimensions are defined as  $10 \times 21.4 \times 150$  mm in accordance with ASME Section IX. The tests were carried out at  $20 \pm 1$ ° mandrel with a diameter of 40 mm, the distance between mandrels was 69 mm and flexural angle was 180°; Instron 5989 L3619 tensile tester was used for testing of flexural strength.

Hardness measurements were carried out in accordance with ASME Section IX from a total of 45 points, including weld metal, HAZ and base material joined by two welding methods. Vickers hardness tests were performed with QNESS Q 10M Automatic Vickers Hardness Tester with a load capacity of HV50gr-10kg.

### 3. Results and Discussion

In duplex stainless steels, any welding process which exerts thermal cycling can result in significant reduction in the ratio of optimum microstructure, which in turn reduces the strength values of both the weld metal and its subzones. All duplex stainless steels solidify as primary ferrite and austenite and if the processing temperature falls below the ferrite transition temperature, the ferrite will eventually transform to austenite. This phase change plays an important role in the ferrite/austenite balance in the final weld metal microstructure. The methods used to determine the phases and the effects of GTAW and SMAW on the properties of 2205 type duplex stainless steels are specified in ASTM A923. According to this standard, the method used to determine the phases in the weld metal is by microstructural examination. In addition, the Feritscope magnetic phase tester was used to determine the magnetic character of these phases present in the weld metal. The delta ferrite phase, which has a magnetic character, formed in the weld metal, was determined by both Feritscope and Image Analysis (phase analysis) methods and the results were compared between them. ASTM A923 recommends that the effect of unwanted detrimental phases in the weld metal on the mechanical properties of duplex stainless steel should be determined by impact testing. In addition, hardness measurements of different phases in the base metal and weld metal were also carried out.

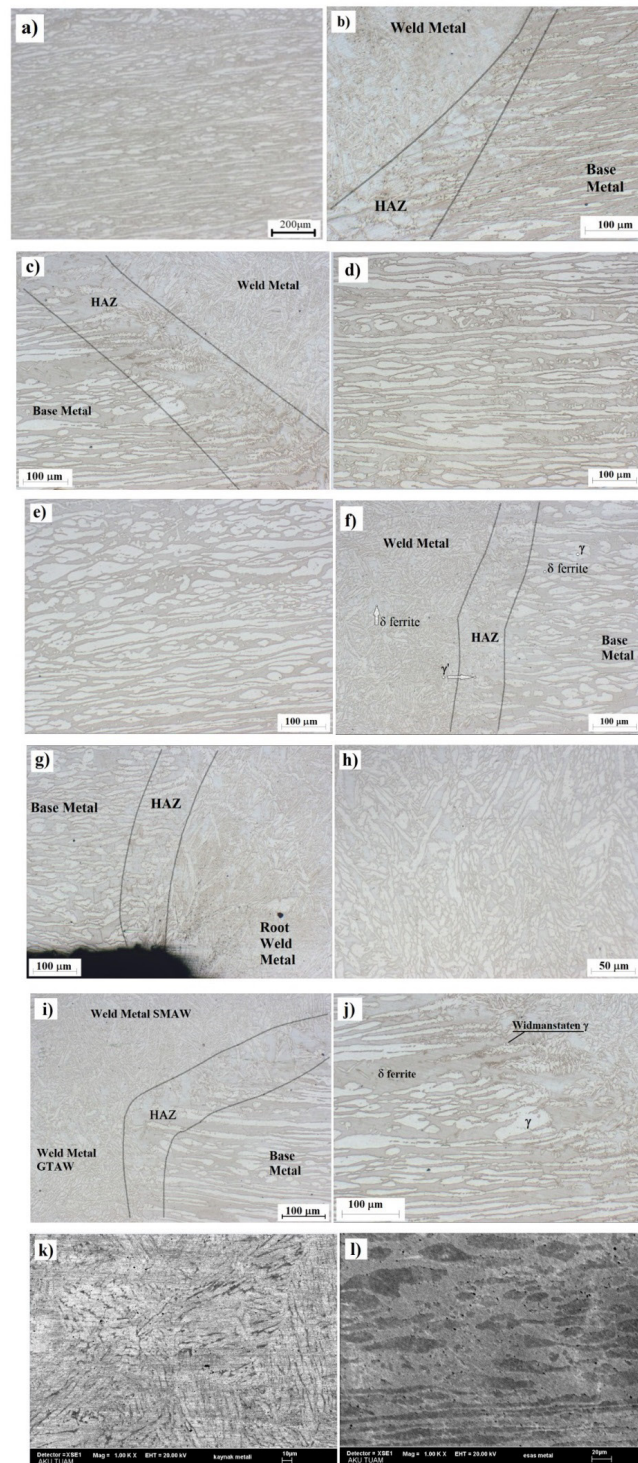
#### 3.1. Microstructural examination of base and weld metals

Phase analyses were conducted to determine the phases present in the base metal and weld metal. The phase ratios were determined on the basis of the volumetric ratio of the phases comprising the structure, employing metallographic images. The magnetic phases were verified by measuring them with ferrite measurement techniques.

The microstructure of as received UNS S31803 (2205) duplex stainless steel is given in Figure 4a. The light coloured regions shown in Figures 4a-j are austenite, while the darker regions are  $\delta$ -ferrite. Austenite and  $\delta$ -ferrite appear to be oriented towards the rolling direction in the base metal part of the images. Microstructural features include elongated appearance of austenite lamellae ranging from a few micrometers to a few hundred micrometers (Figure 4a-j) and it is mostly segregated and aligned along in the same direction (for example in Figure 4b). HAZ microstructures contain mostly equiaxed austenite grains and the amount of delta ferrite appears to increase near weld fusion zone (FZ). Optical images with HAZ show that the microstructure consists of elongated grains along the HAZ-Weld Metal direction. It becomes evident that there is a decomposition of delta ferrite into secondary austenite grains along the heat affected zone (HAZ) with finer secondary austenite phases. In general grains however, extended across several melt pools, indicating an epitaxial growth mechanism.

One of the main parameters affecting the ferrite-austenite ratio in the multi-pass weld metal is the type of microstructure of the welds. The microstructures given in Figure 4 are from the cap pass, filler pass, HAZ, HAZ+2, HAZ+5 deposited by the GTAW and SMAW methods, respectively. As can be seen Figures 4a-i,  $\delta$ -ferrite is generally seen in dark colour in the base metal and in the weld metal. Light coloured structures refer to the austenite phase. In the base metal microstructure, elongated grains formed due to rolling which extend up to the HAZ where  $\delta$ -ferrite grains begin to recover and ferrite grains begin to acquire a coarse-grained appearance towards the solid liquid interface (Figures 4b, c, f, g, and j). From the transition zone, the liquid weld metal continues to solidify by nucleation from  $\delta$ -ferrite grains. As the solidifying interface moves towards the weld metal from the HAZ, the secondary austenitic microstructure feature is observed to be elongated in the form of dendrite arms. Based on PTS 31.40.20.34, 0.5% intermetallic phase or secondary phase precipitates (sigma phase, etc.) cannot have physical contact with each other in an area of 1 mm<sup>2</sup> by volume, but at the same time there is no continuous film or layer formation at the grain boundaries in images given in Figure 4, indicating of such phases. Therefore, the microstructures in different parts of the weld zone show that the formation of micro features in accordance with the standards coincides with the expected structures. In the welded SAF 2205 duplex stainless steels,  $\gamma$  was found to be formed in the form of Widmanstätten structure in the melt zone [19]. In the case of high cooling rate, the amount of delta ferrite phase formed in the weld zone is very high. The deterioration of the phase balance that takes place in this zone is due to low N content in the weld zone compared to the base metal [19]. In the weld bead shown in Figure 4j it was observed that the  $\gamma$  phase was formed in the form of a Widmanstätten structure. Figures 4k and l illustrate the BSE image of the weld filler metal and base metal microstructures, which exhibit no discernible colour difference in composition with respect to the sigma intermetallic phase, which is characterised by a high concentration of Cr/Mo [6-8] (dark phases are associated with a lower atomic number; the bright white colouration observed along the phase/grain boundaries or triple points is indicative of the presence of the sigma phase). It is evident that the base metal and weld metal lack visible grains. However, a phase boundary is discernible in the base metal. The grain shape in some weld metal is predominantly elongated and irregular, extending to the centre of the weld. The grains observed in the heat-affected zone (HAZ) are comparatively regular in size, with an average diameter of approximately 44.57  $\mu$ m, with a standard deviation of 7.17  $\mu$ m. The distance between elongated austenite formations in the base metal is 22.818  $\pm$  7.12  $\mu$ m, with an average width of 30.75  $\pm$  8.26  $\mu$ m. The average length of elongated austenite grains in base metal is 90.89  $\pm$  38.8  $\mu$ m and 4.13  $\pm$  1.43  $\mu$ m in weld metals. It is evident that the average spacing between the gamma grains in base metal is relatively short and the length-to-thickness ratio is above 3.8 for base metal and 4.11 for weld metal austenite plates, indicating a highly elongated or plate like morphology. Base metal morphology may be attributed to the rolling to size but weld metal morphology is related to high cooling rate of weld pool. As the heat source is displaced from the weld zone, the effect of the heat is diminished. This results in structural differences between the weld zone and the base metal. In general,  $\delta$  ferrite transforms to Widmanstätten  $\gamma$  +  $\delta$  ferrite phases at a high cooling rate and/or regular austenite and  $\delta$ -ferrite phases at a lower cooling rate in the heat-affected zone (HAZ). In the weld metal,  $\delta$  ferrite transforms to plate-like secondary  $\gamma$  +  $\delta$  ferrite at different ratios of occurrence with respect to cooling rate.





**Figure 4.** (a) Microstructures of Base Metal; (b) SMAW, WM+HAZ; (c) SMAW Filler Pass, HAZ + WM; (d) HAZ+2 (left); (e) HAZ+5 (left); (f) Hot Pass Weld Metal, HAZ (left); (g) Root Pass (GTAW) + HAZ; (h) Root Pass (GTAW); (i) GTAW Root Left Bottom and SMAW Bead; (j) Widmanstätten  $\gamma$  Structure Formed in HAZ and Weld Metal and BSE images of (k) weld metal (filler) and (l) base metal.

From the solid liquid interface, the molten weld metal continues its solidification by epitaxial nucleation on  $\delta$ -ferrite grains and advances by competitive growth inside the bulk weld metal [20]. The austenite in the form of dendrite arms extends from the grain boundaries into the within the weld metal. During cooling, the secondary intermetallic phases are nucleated in the areas exposed to the 650-900°C range. At 900°C, the  $\sigma$  phase forms at the  $\delta/\gamma$  grain boundaries and grows into  $\delta$ -ferrite, causing the overall hardness values to increase. At 750 °C, secondary intermetallic phases are formed at a lower rate and in smaller size. The formation occurs within the boundaries of  $\delta/\gamma$  grain boundaries. At 650°C, the time required for the formation of secondary intermetallic phases at the  $\delta/\gamma$  and  $\delta/\delta$  phase boundaries is high [21]. The formation of secondary Widmanstätten type  $\gamma$  phase (Figure 4) during cooling after welding can be observed in the parts of the HAZ close to the weld metal. However, since the welding process is carried out by controlling the heat input

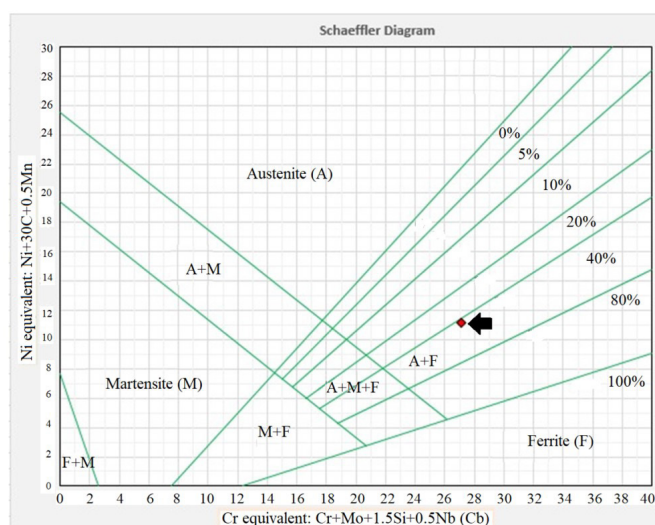
and the temperature between the weld passes during the welding operation, the undesired secondary intermetallic phases such as sigma phase that may occur may have been either avoided or possibly minimised. In addition, considering that the previous weld is heat treated in the multiple passes, it is concurred that any secondary intermetallic phases that may occur during the first weld bead are likely to be dissolved during the new weld pass and the remaining intermetallic type secondary phase formations are largely eliminated in the mean time. Hence, secondary intermetallic phase formations were not allowed to occur during the welding process with controlled heat input through controlling the interpass temperatures. This interpass temperature and heat input cause a build-up of heat in the weld bead and therefore the HAZ widths are also affected. As can be seen in Table 7, the HAZ widths increase in length from the root pass to the fill passes, except for the cap pass. The hot pass above the root pass is the first SMAW weld and is deposited with the lowest interpass temperature of 85 °C. The subsequent fill passes are deposited at higher interpass temperatures, resulting in greater HAZ widths. The cap pass, on the other hand, tends to cool faster as it is the last weld deposited on a larger volume of metal, i.e. root, hot and filler passes, resulting in a slightly smaller HAZ width compared to the filler passes.

**Table 7.** The HAZ widths of different zones from root pass to cap pass.

| Line Number   | HAZ Widths Avg. (µm) |
|---------------|----------------------|
| Root pass HAZ | 167.95±35.1          |
| Hot pass HAZ  | 140.75±26.2          |
| Fill pass HAZ | 333.30±55.6          |
| Cap pass HAZ  | 231.97±68.5          |

### 3.2. Investigation of ferrite measurement results of duplex stainless steel

The amount of phases in the base material and weld metal can be determined by microstructural characterisation and magnetic measurement techniques. In addition, the Schaeffler, Delong and WRC prediction diagrams developed to estimate the amount of phases that make up the microstructure of the weld metal are also used. These diagrams may be preferred in the evaluation process as they provide an estimate assuming that the welding parameters and chemical composition have not changed. The weld metal location calculation is based on the Schaeffler diagram and is shown in Figure 5.



**Figure 5.** The Location of % 30 UNS S31803 (2205) and %70 Filler Metal Composition in Schaeffler Diagram in Red Diamond [22].

According to PTS 31.40.20.34, the ferrite content of the weld metal and HAZ is acceptable in the range of 35-65%. The results of the ferrite meter measurements are given in Table 8.

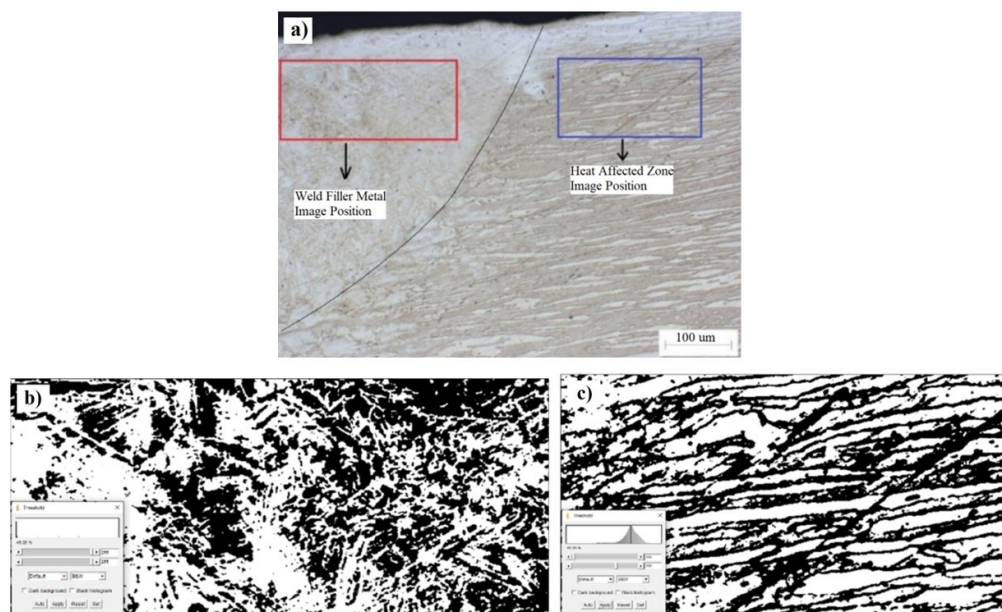
**Table 8.** Ferrite Phase Quantity Measurement Results (vol. %).

| Region under review | Ferrite Value (%) |      |      |      |         |
|---------------------|-------------------|------|------|------|---------|
|                     | 1                 | 2    | 3    | 4    | 5       |
| Right HAZ           | 43                | 43.7 | 44.4 | 45.1 | 44.2    |
| Weld Metal          | 42.6              | 43.2 | 44   | 42.1 | 43.8    |
| Left HAZ            | 44.2              | 44.9 | 43   | 43.7 | 44.6    |
|                     |                   |      |      |      | Average |
|                     |                   |      |      |      | 44.08   |
|                     |                   |      |      |      | 43.14   |
|                     |                   |      |      |      | 44.08   |

As a result, the ferrite percentages were within the desired range and the test was considered successful. A study was also carried out using Image J program to verify the ferrite test because the ferrite meter measurement is made from the outer surface of the material. For this study, on Figure 6a, zones as close as possible to the cover pass were selected in order to correctly evaluate the comparison. Analysis results are given in Figure 6c and d. The selected areas are shown in Figure 6a.



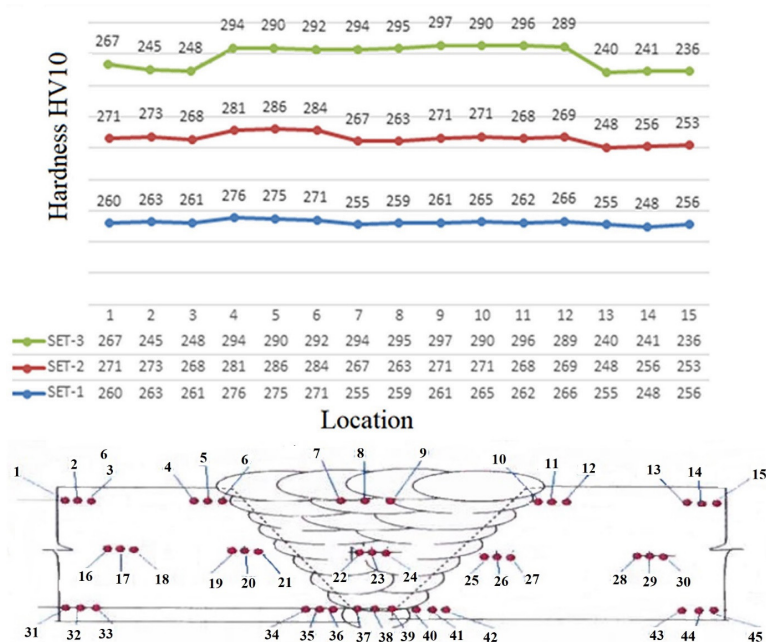
After evaluation using the Image J program, the weld filler metal is shown in Figure 7b with a % ferrite ratio of approximately 45.28% and the HAZ is shown in Figure 6c with the ferrite ratio of approximately 46.33%. As can be seen when magnetic measurements and phase analysis by Image J are considered together, the results of both methods are close to each other. In metallographic phase analysis, the accuracy of 95% is expressed according to the ASTM E 562 standard. For magnetic measurement techniques, the reliability rate varies depending on the measuring capability of the instrument. Magnetic methods are preferred to metallographic methods because they are non-destructive, fast and practical.



**Figure 6.** (a) Image J Program Based Image of Cover Pass; (b) Weld Filler Metal Processed with Image J Program; (c) HAZ (Heat Affected Zone) Processed with Image J Program.

### 3.3. Hardness test results in duplex stainless steel

Weld joints produced with different welding methods, heat input and hence cooling rates changes due to interpass process. The effect of different ferrite-austenite ratios on the hardness of the joint due to these variations was determined by hardness measurements on a total of 45 points from the regions shown in Figure 7. The evaluation of the points measured according to the ASME Section IX is made according to PTS 31.40.20.34 and in the case of duplex stainless steels the main material 290 HV<sub>10</sub>, ITAB and 330 HV<sub>10</sub> for weld metal are the highest permissible hardness values. The hardness test values are given in Figure 7.



**Figure 7.** Hardness Measurement Values and Their Locations.

Vickers hardness values of the welded material are high in the weld bead and in the parts of the HAZ adjacent to the weld. The  $\gamma$  phase in the Widmanstätten structure in this region has caused the hardness value to increase. Sathiya et al. [23] explained that the hardness value obtained in the weld zone is higher than the HAZ due to the presence of a higher amount of secondary austenite phase in the weld zone. HAZ hardness values show that a large decrease in hardness value is observed due possibly to the high amount of  $\delta$ -ferrite phase observed in this region. A slight change in the hardness values between base metal and HAZ, points 3 and 4, are observed in the HAZ. Microstructural examination indicates that secondary intermetallic phases were precipitated in these zones, and literature suggests that hardness values increase due to the formation of secondary intermetallic phases [24]. It is also claimed by Santos et al. [24] in a study of friction mixing welding of 2205 duplex stainless steels, that the hardness values of austenite after welding were higher than delta ferrite due to the fact that austenite completed solidification in a much finer grained structure than delta ferrite. As the cooling rate in welds increased, a small increase in the hardness of the delta ferrite phase in the weld was also observed. Similar to secondary austenite, precipitation of secondary intermetallic phases and local differences in delta ferrite grain size in the weld metal are believed to cause this condition [25]. Researchers have identified the reasons why the hardness of secondary austenite is higher than delta ferrite: solid solution hardening by C, N, Cr, Mo, Ni, etc [26]. With this mechanism, they attributed the coexistence of two phases with increased hardness, as well as the mechanism of increase in strength due to grain thinning and the increase in internal stress due to the difference in thermal shrinkage properties of the two phases on cooling from annealing temperature. However, considering the effect of the nitrogen atom in solid solution, the increased nitrogen content may have caused the austenite to be stronger than ferrite, preferably causing the formation of intermediate solid solution hardening [26]. The hardness profiles of the welded specimens showed that hardness values of the weld metal and HAZ were found to be higher than that of base metal. As seen in Figure 4, the secondary austenite phase ( $\gamma'$ ) is finer grained than delta ferrite (matrix phase) across the weld metal and the HAZ. As a result, the increased cooling rate increased the hardness of both secondary austenite in weld metal and Widmanstätten secondary austenite phases in the HAZ due possibly to their fine and dendritic appearance in addition to the matrix hardening from elevated cooling rate.

### 3.4. Tensile, flexural and notch impact test results

In the study, tensile tests were performed on specimens prepared in accordance to standards. According to PTS 31.40.20.34, the lowest permissible base material yield strength is 488 MPa, the ultimate stress is 620 MPa and the elongation % is 25. The tensile test results of the specimens are given in Table 9. The tensile samples failed not from the weld metal and HAZ, but from the base metal. It was concluded that the mechanical properties of specimens were above the limits, exceeding the minimum yield and ultimate tensile strengths specified in the standards.

**Table 9.** Tensile Test Results of Weld Metal (Average).

| Cross-sectional Area (mm <sup>2</sup> ) | R <sub>eH</sub> (N/mm <sup>2</sup> ) | R <sub>p0.2%</sub> (N/mm <sup>2</sup> ) | F <sub>max</sub> (kN) | R <sub>m</sub> (N/mm <sup>2</sup> ) | % Elongation |
|-----------------------------------------|--------------------------------------|-----------------------------------------|-----------------------|-------------------------------------|--------------|
| 381                                     | 586                                  | 573                                     | 285                   | 747                                 | 30           |

Four prepared side bending specimens were tested using the bending test parameters given in Table 10. As a result of the test, no tearing and cracking was detected, no welding defects such as cold adhesion were found. Due to high elongation value, it is not surprising to observe structural defects resulting from excessive matrix hardness or brittleness which may sometimes result from the formation of intermetallic compounds such as chi or sigma phases [27,28] of which authors have not observed in this study.

**Table 10.** Bending Test Results (Specimen size: 10 × 21.4 × 150 mm).

| Sample No | Test Type    | Mandrel Diameter (mm) | Distance Between Mandrels (mm) | Tilt Angle | Results (P: pass, F: Fail) |
|-----------|--------------|-----------------------|--------------------------------|------------|----------------------------|
| 01        | Side bending | 40                    | 69                             | 180°       | P                          |
| 02        | Side bending | 40                    | 69                             | 180°       | P                          |
| 03        | Side bending | 40                    | 69                             | 180°       | P                          |
| 04        | Side bending | 40                    | 69                             | 180°       | P                          |

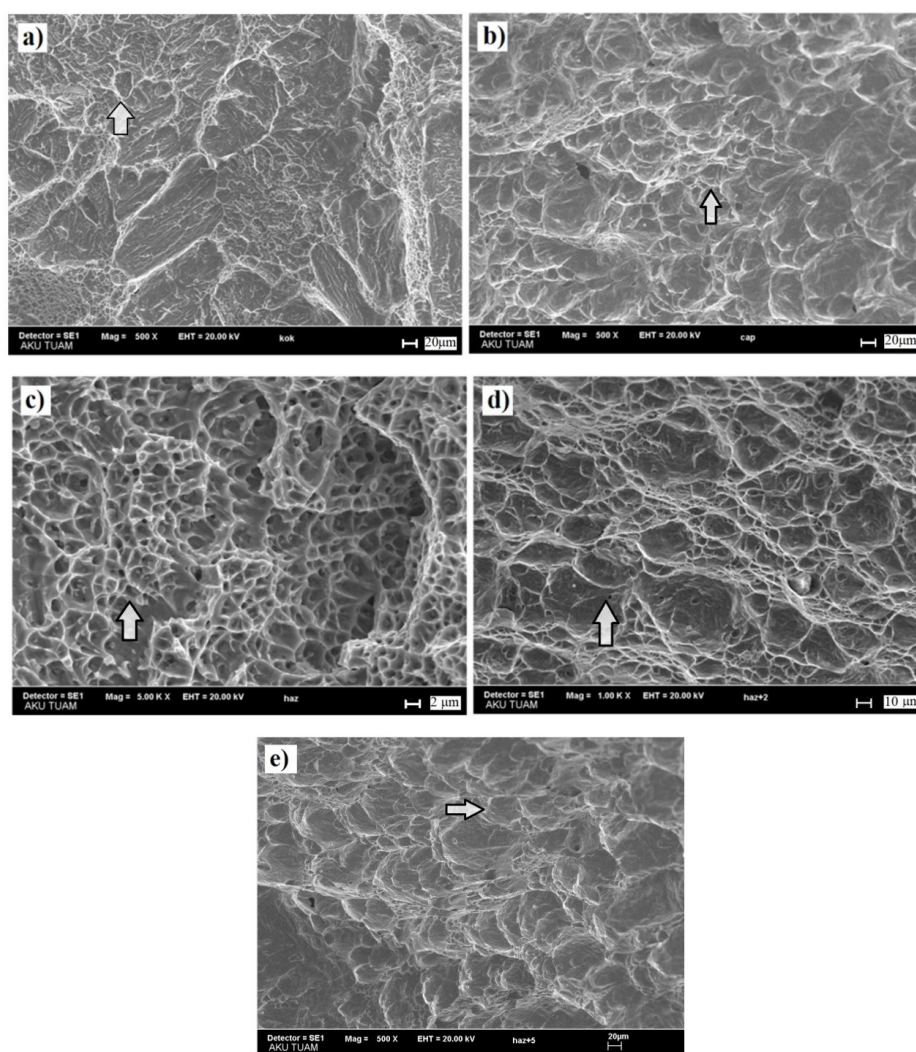
Charpy-V notch impact test was performed to measure the toughness values of the samples prepared in accordance with the standards at -50 degrees. The test locations and impact test results are given in Table 10. The test results indicate that impact tests show variations in relation to test locations. Although the test results are considered to be successful, the results were close to the minimum desired values defined by the standards. In Table 11, it can be seen that the most affected test location was root weld, which may be related to having the lowest hardness values in the weld zone (Table 11) and the lowest delta ferrite content (Table 8). Considering that root welds are repeatedly heated and cooled, the change in austenite morphology may be the reason for the lower impact value compared to the cover pass, which is the last weld deposited in the weld. In the HAZ, the impact values are relatively high compared to the weld metal locations; HAZ+5 location is the highest as

it is almost in base metal, showing the highest impact toughness as much as reported in the literature with zero amount of intermetallic phases [29]. HAZ+2 and narrow zone HAZ have almost similar results compared to other parts of weld zone, showing that grain coarsening and excessive microstructural variations contribute to the similar impact toughness values.

**Table 11.** Notch Impact Test Results of Various Parts of Weld.

| Tested Part | Average Impact Toughness (J/cm <sup>2</sup> ) | Average Test result (J) |
|-------------|-----------------------------------------------|-------------------------|
| Cover       | 93                                            | 75                      |
| Root        | 66                                            | 52                      |
| HAZ         | 206                                           | 165                     |
| HAZ+2 mm    | 195                                           | 156                     |
| HAZ+5 mm    | 391                                           | 313                     |

The fracture surfaces of the notch impact specimens from different parts of the weld zone and base metal are given in Figure 8. SEM images of fracture surfaces showed that, all failures were ductile in nature. It is generally assumed that in root pass, the cracks propagate intergranularly and large crevices are observed during the tensile testing. When the SEM images of the samples were examined, it was observed that the structure was ductile fractured as a result of the micro cavities (dimples) being small and deep. Since the austenite phase in the FCC lattice structure does not have a ductile to brittle transition, it has been observed in the results that it gives good toughness and ductility even at cryogenic temperatures, and the ductile brittle fracture transitivity of the  $\delta$ -ferrite phase in the BCC lattice structure is one of the reasons why the notch impact energies of samples taken from the same regions are different from each other.



**Figure 8.** (a) Root Pass Fracture Surface; (b) Cap Pass Fracture Surface; (c) HAZ Fracture Surface; (d) HAZ+2 Fracture Surface; (e) HAZ+5 Fracture Surface Image (Fracture dimples are marked by an arrow).

#### 4. Conclusions

The following conclusions can be drawn from this study which was conducted to determine the effects of using two different welding methods i.e. GTAW and SMAW on mechanical and physical properties of UNS S31803 duplex stainless steel.

- Due to the need to perform the welding process with controlled heat input, the interpass temperatures were controlled and the temperatures that would cause secondary phase formations were not allowed to occur, therefore, no such phases were observed;
- GTAW and SMAW methods were successfully applied in a thick walled duplex stainless steel which is usually joined by submerged arc welding;
- As the cooling rate was slow due to having multiple passes and interpass process, the sufficient time was allowed for delta ferrite to be converted to secondary austenite in the weld filler metal;
- It was observed that the cooling conditions due to different methods and multiple passes caused an increase in the hardness values of the austenite and delta ferrite phases;
- It was observed that the hardness was highest in the weld and HAZ made by GTAW method in the root pass and the hardness was lower and close to the parent metal in the weld and HAZ made by the SMAW method in the cover pass. The main reason for this inter-regional hardness change is believed to be caused by secondary austenite formation resulting from higher cooling rate;
- The weld metal after welding is in duplex structure and does not contain different phases as determined by both magnetic and image analysis methods;
- All tensile specimens failed from the base metal, which implies that the weld joint deposited by two different welding methods is sufficient in terms of strength and even more durable than the parent material;
- SEM-BSE and optical analysis of weld metal, HAZ and base metal microstructures indicated that there was no observed formation of secondary phases along the phase boundaries and grain boundaries;
- When the fractured tensile specimens were examined, it was found that the specimens failed by ductile mode and therefore their ductility were high and there was no tearing and opening from the weld zone in bent specimens;
- In addition, the impact values of the weld metal were higher than those required by standards.

#### Authors' contributions

The authors state that Başar Ersegün Çelik carried out experiments and drafted the article, Şükrü Talaş designed the experimental procedure and critically read and finalised the article.

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