Development of coated electrodes for welding of Super Duplex steel

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A R T I C L E   I N F O

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A B S T R A C T

The microstructure, tensile strength and impact strength following welding of Super-Duplex-Stainless-Steel using coated electrodes were investigated. It was observed that the lower the basicity index of the coating, the higher the reduction of oxygen in the weld pool. This can be explained by the interference of the basic elements with the oxygen reduction by the acidic elements of the coating.

The change in the microstructure of the weld indicates the different cooling rate in each zone: The cooling rate during welding, from the highest to the lowest, is as follows: Interface area >>> Root area >>> Face area.

In the face area, the highest content of austenite and lowest content of ferrite were observed, which indicates that this area has experienced the lowest cooling rate.

In the face area, the lowest content of austenite and the highest content of ferrite were observed, hence indicating that this area has experienced the highest cooling rate.

All welds showed the presence of Widmanstätten austenite, grain boundary austenite, sigma phase and large grains of ferrite.

It can be deduced that the sigma phase cannot be avoided in the welds; however, the secondary austenite can be avoided, by employing an appropriate “low heat input policy”.

It was found that the higher the content of the secondary austenite and sigma phases in the weld, the lower the yield stress, due to the brittle characteristics of these phases. However, the higher the austenite content in the weld, the higher the elongation.

1. Introduction

The Super-Duplex Stainless Steel contains roughly an equal volume fraction of austenite and ferrite, to ensure optimum properties [1]. This unique microstructure is characterized by the combination of high ductility and toughness over a wide range of temperatures, combined with the exhibition of excellent corrosion and high temperature oxidation resistance, which makes the Super Duplex Stainless Steel a leading choice of engineers looking for excellent mechanical properties combined with excellent corrosion resistance. This results in a continuing interest in Super Duplex Stainless Steels and especially in welded joints with good mechanical properties, which are determined by the weld microstructure [2, 3].

The electrode’s coating has a major influence on the microstructure as well as the tensile strength and impact strength of the weld. However, the relatively high oxygen levels in the weld metal used for rutile electrodes (compared to those of a basic coated electrode) are evident by the low impact properties.

The alloying elements (originating from the coating and rod) such as Cr, Ni, Mo, and N, control the ferrite/austenite balance, and the derived mechanical properties. In addition to the influence of the coating composition on the phase balance and the mechanical properties, it is also influenced by the heat input and cooling rate during welding. The ferrite to austenite transformation during heating affects the austenite grain size, phase fraction and homogeneity, which in turn, affects the kinetics of the phase transition in ferrite and its grain growth during cooling [4]. Moreover, at lower temperatures, the propensity for precipitation of the intermetallic phases increases, resulting in the formation of a variety of secondary phases like σ, Cr2N and its derivatives, secondary austenite, γ, R etc., mostly on the ferrite/austenite boundaries. The most important phase of these is σ, because of the large fraction often observed and its detrimental effect on the mechanical properties. Therefore, the multi-pass weld, which contains reheated regions, displays a remarkably different phase balance, partitioning of the alloying elements and consequently strength, compared to a single pass in the bulk material. Atamert et al as well as Nilsson and Chai proved that the multi-pass welds are inhomogeneous with respect to both the

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microstructure and properties [5, 6].

The characteristics of the welding process have a significant effect upon the weld microstructure, as detailed below:

Welding filler materials for Super Duplex Stainless Steel were selected to obtain a proper phase balance in the weld region, normally with a higher nickel content than the base metal, in order to promote austenite formation during the rapid cooling associated with welding, with properties that are at least equal to those of the base metal.

The electrode's coating affects the purity of the weld metal, the composition and hence the microstructure. The rutile coating contains a high amount of titanium oxide (TiO2) powder (up to 50%wt), as well as natural silicates and ferro-metal to reduce and refine the weld. The slag that is formed during the process is related to FeO-MnO-TiO2. The slag has a minor acidic reaction.

The welding parameters, such as the heat input and weld design, have a direct influence on the microstructure of the weld. The design of the weld complies with the requirements of AWS A 5.4, to ensure full penetration and to minimize the risk of burn-through. Since the amount of the harmful intermetallic phases increases as the cooling rate decreases, the control of the heat input plays a major role in the welding design – minimizing the heat input to reduce the formation of the intermetallic phase. The heat input can be controlled by proper selection of the following welding parameters: Lower current and voltage level, higher welding speed, shorter welding seams and low interpass temperature (below 150 °C).

The location of the inspected region also affects the cooling rate and microstructure. The upper surface of the melt pool is cooled by heat conduction into the underlying hot weld metal, whereas the weldment root area, although re-heated many times, is more intensively cooled by the backing plate underneath the welding assembly. Last but not least, the melt close to the seam/HAZ interface is the most rapidly cooled region. Therefore, the presence of different microstructures – ferrite & austenite, as well as intermetallics – can be expected. The thermal cycle and especially the post-welding cooling rate, namely the phase transformation kinetics, play a significant role in the austenite to ferrite transformation. This transformation affects the grain size and orientation, as well as the phase ratio and morphology of the weldment microstructure. Changes in the microstructure can affect the properties, which is more evident in the weld metal [7].

The relationship between the electrode coating and cooling rates and the microstructure, composition and mechanical properties of the weld metal are discussed below, in an effort to design a Super-Duplex Stainless Steel weldment with improved strength and toughness.

2. Experimental

Three rutile-based electrodes having the same core wire, which complies with AWS A 5.9 ER 2205 (see Table 1), were investigated.

Three rutile based electrodes having the same core wire, which complies with AWS A 5.9 ER 2205 (see Table 1), but with different coatings as listed in Table 2, were investigated.

The welding assembly used in this research, as shown in Fig. 1, was designed in accordance with the AWS A5.4 requirements. For each electrode, separate assemblies were constructed:

| Electrode Types | L100 | L127 | L208 |
|-----------------|------|------|------|
| Basicity index  | 1.4  | 1.2  | 1.1  |

Where the Basicity index is given by:

$$BI = \frac{CaO + CaF_2 + MgO + K_2O + Na_2O + Li_2O + BaO + SrO + \frac{1}{2} SiO_2}{\frac{1}{2} (Al_2O_3 + TiO_2 + ZrO_2)}$$

(1)

Coatings with a basicity ratio higher than 1.2 are considered basic coatings. Coatings with a basicity ratio below 1.2 are considered acidic (rutile) coatings [8].

The welding assembly and specimens’ location are presented in Fig. 1a and b respectively.

The welding parameters are listed in Table 3.

Welding operations were conducted using HOBART MEGA ARC 5 welding machine operated at DCEP (Direct Current Electrode Positive) polarity, where the coated welding electrode is connected to the positive electrical connector. Welding position: 1G - flat position welding of a groove. The arc was kept short in order to optimize the electrode gas protection and prevent porosity. According to AWS A5.4, the interpass temperature for stainless steel electrodes should be kept under 150 °C. However, during the development of the electrode, it was recognized that keeping the interpass temperature below 110 °C yields better results. However, due to arc instabilities in electrode L208, the current was increased to 120A and accordingly the heat input also increased.

Round tensile test bars (Ø6.25 mm, length 25 mm) were machined from the center of the weld, in order to obtain an undiluted sample. The tests were performed at room temperature on Instron 1273 with a deviation of ± 3MPa.

From each plate, five Charpy V-Notch specimens of 10*10*55 mm³ were machined, in compliance with the DIN 50 122 –ISO V standard. The specimens were sectioned perpendicularly to the welding direction, so that the notch will be positioned in the middle of the weld area. The impact tests were performed at 25 °C, 0 °C, -30 °C and -40 °C, in Schenck Trebel tensile testing machine, type RPSW 150/300.

The chemical composition of the majority of the alloying elements of the weld metal (measured in the middle of the weld track), core wire and parent plate, was determined using a SPECTROMAXx Metal Analyzer. The nitrogen and oxygen measurements were conducted using Galileo G8 ON/H analyzer.

Metallographic samples were polished with diamond slurry and silica colloidal slurry to expose the cross-section of the weld and electrolytically etched with 33 cm³ HCl, 33cm³ alcohol and 33cm³ distilled water, making the austenite phase light and the ferrite phase dark. The microstructures were analyzed at higher magnifications using a SUPRA 40 Carl Zeiss Scanning Electron Microscope (SEM). Phase identification of the weld metal was conducted using Kikuchi lines by EBSD analysis. An Energy Dispersive X-ray (EDS) analyzer coupled to the SEM instrument was used to take quantitative measurements of the composition of each phase. The chemical composition of the electrode coating and of the slag was measured utilizing the SPECTRO XRF XEPOS system.

ASTM E562-11 Standard [9] defines the calculation procedure to determine the austenite to ferrite ratios within the various weldments and parent metal. The standard deviation of the phase amount is about 2.4%.

Table 1

| Element | C  | Mn | Si | P  | S   | Cr  | Ni | Mo | Co | N  | Fe |
|---------|----|----|----|----|-----|-----|----|----|----|----|----|
| %wt     | 0.018 | 1.55 | 0.48 | 0.014 | 0.001 | 22.40 | 8.73 | 3.19 | 0.047 | 0.14 | Bal. |

Table 2

The Basicity index of the Coating.
3. Results

The compositions of the weld metal and slag are listed in Table 4.

The level of impurities in all of the welding electrodes complies with the AWS A5.4 E2594-16 standard, according to paragraph 3.6.7.3, EBSD diffraction patterns were recorded for an elongated grain located in the...

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### Table 3

| Electrode No. | Diameter [mm] | Current [A] | Voltage [V] | Welding speed [mm/min] | Heat Input [kJ/mm] | Interpass temperature [°C] |
|---------------|---------------|-------------|-------------|------------------------|---------------------|---------------------------|
| L100          | 3.25          | 110         | 26–28       | 160–200                | 0.84–1.07           | 107                       |
| L127          | 3.25          | 110         | 26–28       | 160–200                | 0.84–1.07           | 107                       |
| L208          | 3.25          | 120         | 26–29       | 160–200                | 1.044–1.17          | 107                       |

### Table 4

| Electrode | C   | Si  | Mn  | P   | S   | Cr  | Ni  | Mo  | Nb  | Cu  | N   | O   |
|-----------|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|
| L100      | 0.042 | 0.72 | 0.68 | 0.0098 | 0.0098 | 25.26 | 9.12 | 3.52 | 0.0073 | 0.135 | 0.32 | 0.931 |
| L127      | 0.032 | 0.96 | 0.9  | 0.032  | 0.013  | 25.21 | 9.96 | 3.98 | 0.037  | 0.72   | 0.35 | 0.0853 |
| L208      | 0.037 | 0.89 | 1.18 | 0.021  | 0.0093 | 27.39 | 9.57 | 3.46 | 0.0085 | 0.74   | 0.38 | 0.905 |

| Electrode | Na2O | K2O | MgO | Al2O3 | CaO | SiO2 | TiO2 | Cr2O3 | MnO | SrO |
|-----------|------|-----|-----|-------|-----|------|------|-------|-----|-----|
| L100      | -    | 4.51| -   | 6.34  | 6.64 | 25.12 | 41.30 | 24.30 | 4.6  | 1.03 |
| L127      | -    | 4.52| -   | 3.65  | 6.63 | 14.13 | 42.31 | 23.01 | 4.5  | 0.52 |
| L208      | -    | 4.52| -   | 3.20  | 6.21 | 14.12 | 41.2  | 24.28 | 4.45 | 0.32 |
Fig. 2. Diffraction analysis of a grain located in the center of weld metal L100 - A. Complete analysis, (x) microstructure (y) EBSD diffraction (z) EBSD diffraction with indexing. B. Diffraction of the gamma-phase.
middle of the L100 weld. The diffraction analysis shows the existence of γ – FCC phase, as illustrates in fig. 2a & b.

Fig. 3 presents the microstructure and EBSD diffraction at the γ – FCC grain boundary. The presence of δ-BCC, discernible in Fig. 3, indicates a unit cell close to the BCC structure, having lattice parameters of \(a = 2.87, 2.87, 2.88 \, \text{Å}\). The parameters correlate with the lattice parameters of ferrite (δ). However, it can be seen that there is a slight distortion in the lattice parameters, close to the BCT unit cell. It should be taken into consideration that the high percentage of BCC observed was transformed from BCC to BCT. The deformation or lattice distortion can be attributed to the transformation (from BCC TO FCC) strain induced during the cooling. The lattice changes can be described as a unit cell in the parent lattice, that can be deformed by the transformation strain to become a unit cell in the final microstructure. The transformation strain is a homogeneous deformation applied equally to all unit cells in a crystal. Originally it is meant only for a specific point in the lattice, i.e. the atoms on the corners of the unit cell. The transformation strain is identical to the strain resulting from continuous deformation [10].

The presence of the σ phase at the ferrite/austenite (shown in green & yellow respectively) interface is shown in Fig. 4.

EBSD analysis of the weld metal in the L127 electrode reveals the presence of the same three phases - γ-Fcc, σ phase and δ-BCT/BCC, exhibiting the same lattice parameters as in the case of the L100 counterpart. Further analysis reveals that the location of the sigma phase is adjacent to the austenite grain boundaries, as in the L100 counterpart.

EBSD analysis of the weld metal in L208 electrode reveals the presence of the same three phases - γ-Fcc, σ phase and δ-BCT/BCC, exhibiting the same lattice parameters as in the case of the L100 counterpart. Also here, the σ-phase was adjacent to the ferrite/austenite interface. However, in this case, \(\text{Cr}_2\text{N}\) and \(\text{Cr}_2\text{C}_6\) were detected (marked with a red arrow), as shown in Figs. 5 and 6, respectively. Their presence can be related to the rapid cooling of the weld metal during welding, resulting in super-saturation of the ferrite with nitrogen, promoting the formation of inter-granular nitride.
Fig. 4. EBSD color mapping for L100 weld. A – Microstructure, B – EBSD signals. Color key: green-γ-Fcc, red-σ phase, yellow – δ-BCT/BCC and white – weak signal.

Fig. 5. Diffraction analysis of weld metal L208 on the ferrite grain, Cr2N is presented (by a red arrow). A – microstructure, B- EBSD diffraction and C – EBSD diffraction with indexing.
The microstructure at the face region (Fig. 7a), the root area (Fig. 7b) and interface (Fig. 7c) in the L100 weld metal is composed of different phases, as follows:

In Fig. 7a – the face zone – Ferrite appears as large grains composing 50 ± 5% vol. Three types of austenite are discernible – grain allotriomorph austenite, Widmanstätten austenite and intergranular austenite, which precipitate in the large columnar ferrite grains.

The allotriomorph austenite, observed parallel to the ferrite boundary, composes 10 ± 1%vol. of the weld in this area. The width of the grains is about 4μm and the length ranges from 30-50 μm.

The Widmanstätten austenite grain, observed perpendicularly and adjacent to the allotriomorph austenite as clusters of close to rhombus shaped grains, composes 25 ± 3% vol. of the area. The grain size ranges from 2*8 μm² to 6*10 μm².

Intergranular austenite was also observed inside the ferrite grains, having a rounder shape, composing no more than 5%±1vol. of the area inspected. The size of these grains ranges from 6μm² to 18μm².

Grayish color sigma phase appeared between the grain boundaries of the ferrite and austenite, in less than 1%vol. of the area.

In addition, small gas pores appeared in the weld, ranging between 6-36 μm².

In Fig. 7b – the root area – Ferrite appears as large grains dominating 60 ± 5% vol. Two types of austenite were observed: allotriomorph austenite and Widmanstätten austenite; which precipitate in the large columnar ferrite grains.

Allotriomorph austenite, observed parallel to the ferrite boundary, composes 5 ± 1%vol. of the weld in this area. The width of the grains is 3–4μm and the length ranges from 30-50 μm.

The Widmanstätten austenite grains, observed perpendicularly and adjacent to the allotriomorph austenite, composes 25 ± 3% vol. of the area. The majority of grain size was in the range of 1–2 μm², and a small amount were 9*6 μm² in size.

A Sigma phase was not observed; however, 3 μm² pores were observed.

In Fig. 7c – the weld zone of the interface zone - Ferrite appeared as large laths, which dominated 60–70% vol. of the area inspected. Two types of austenite were observed: allotriomorph austenite and Widmanstätten austenite, which precipitated near large columnar ferrite grains.

Allotriomorph austenite, observed parallel to the ferrite boundary, composed 20 ± 3%vol. of the weld in this area. The width of the grains was about 4μm and the length ranged from 20-30 μm.

The Widmanstätten austenite grains appeared perpendicularly and adjacent to the allotriomorph austenite, in relatively small amounts, composing 1–2% vol. of the area. The grain size was about 2*4 μm².

The change in the microstructure differs from the other two regions due to the heat flow direction, rate and its location in the weld. The grain size in the HAZ – the plate zone in Fig. 7c – slightly grew compared to the size in the welded plates.

The phases composing the face and root regions in L127 and L208 and the results of L100 that are presented above and which are summarized in Table 5 were calculated by employing the manual point counting procedure as per ASTM E 562-11 [11].

The amount and size of the Widmanstätten, allotriomorph, intergranular and secondary austenite for L100, L127 and L208 welds are summarized in Table 6 (using ASTM E 562-11 [11]).

Three tensile test bars were tested at room temperature. The average yield stress, UTS and elongation are presented in Table 7.

The average values of 5 Charpy impact tests performed at various temperatures are summarized in Table 8.

The mechanical properties are in compliance with the requirements of AWS A5.4 E2594-16.

4. Discussion

4.1. Metal/slag composition

The weld metal composition of L100 exhibited the lowest level of Si, S and P impurities, whereas the highest level of O was found in the melt and for SiO₂ - in the slag (Table 4). These are correlated to the highest basicity index of the L100 coating – the basic slag breaks the covalent bonding of the acid metals Si, S & P in the melt, resulting in enrichment of the slag via reaction with the basic slag compounds. On the other hand, the basic slag does not effectively remove the Oxygen from the melt, as reflected in the highest O level in L100.

The recommended ratio of chromium equivalent (Creq*) to nickel equivalent (Nieq*), which are given by:

\[ \text{Creq}^* = \text{Cr} + 1.5 \text{ Si} + 1.4 \text{ Mo} + \text{Nb} - 4.99 \]  

(2)
\[ \text{Ni}_{eq}^* = \text{Ni} + 30\text{C} + 0.5 \text{Mn} + 26 (\text{N} - 0.02) + 2.77 \]  

(3)

Should be in the range of 1.5–2 [9,12]; However the higher the ratio, the higher the hot cracking susceptibility. The \( \text{Cr}_{\text{eq}}/\text{Ni}_{eq}^* \) ratios of the addressed weld metal were 2.02, 2.00 and 2.11 for L100, L127 and L208, respectively. The higher hot cracking susceptibility of L208 can be explained by the higher level of brittle phases/sigma, nitride and carbide compounds in this weld (Table 5) [13].

4.2. Cooling rate

The heat input influences the cooling rate. The higher the former, the lower the latter [7]. Accordingly, due to the higher heat input in L208 (Table 3), it cools at a slower rate, allowing more time for secondary phase precipitation, and hence the higher level of secondary/minor phases/compounds as against the counterparts, as can be seen in Table 5. In addition, the local cooling rate in the interface, root and face areas (as defined in Fig. 1a) is affected by the heat flow from the solidified region to its surroundings – the most effective heat sink is the welded plate, due to the highest mass ratio of the metal/solidified region; Whereas the face area is cooled by radiation and conduction to the hot weld bead, thus the lowest heat flow. The root area is cooled by the cold thin back plate, hence the heat extraction from the root area is in-between the two aforementioned cases. Therefore, the cooling rates can be ranked as follows: Interface area >>> Root area >>> Face area [14].

4.3. Microstructure

The solid state phase evolution during cooling from ferrite is as follows:

- Grain-boundary-austenite nucleates on the ferrite grain boundaries, then Widemanstätten austenite grows perpendicular to the newly formed austenite. During further cooling, secondary/minor phases like sigma, secondary austenite, nitride and carbides, precipitate. The phase evolution depends on the cooling rate – the higher the cooling rate, the lower the phase transition/formation/precipitation: This explains the experimental results as follows:

1. The higher cooling rate in the root, as against that in the face area, results in a higher ferrite/austenite ratio in the root.
2. The high heat input used in L208 compared with L100 and L127 results in a lower cooling rate in L208. Therefore, in L208, the Widemanstätten austenite is partially transformed into secondary austenite, which was not the case in the L100 and L127 counterparts (Table 6).
3. The same heat input in L100 and L127 resulted in similar amounts of Widemanstätten austenite (25–30%), whereas the L208, in which a higher heat input was used, resulted in decomposition of the Widemanstätten to secondary austenite, exhibiting only 10–15% of Widemanstätten austenite.
4. A small amount of intergranular austenite was discernible in the face area due to the typical slow cooling rate in this region. However, in L208, in which the cooling rate was slower, the amount of the intergranular austenite was much higher (30% Vs. 5–10%).
5. A Sigma phase was found only in the slow cooling face area in L100 and L127, whereas it was dispersed all over the cross-section of the L208 weld metal, due to the slower cooling rate. It is worthy to note that the higher Cr content in L208 (Table 4) may also contribute to the higher level of sigma and secondary austenite in this alloy.
6. In most cases, the microstructure in the face area is coarser than in the root and interface regions, due to the typical lower cooling rate in the face area.

4.4. Mechanical properties

The yield stress of L208 is lower than that of L100 and L127 due to the higher content of brittle phases in the L208 weld. The lower elongation of L127 as against this of L100 can be correlated with the higher impurity level in L127 (Table 4). The lower elongation in L208 can be attributed to the presence of brittle phases in this weld.
The relatively poor impact energy of all weld metals is, apparently, correlated with the presence, even in a very small amount, of the brittle sigma phase in all weld metals. The lower impact of L100 compared with L127 resulted from the higher proportion of O dissolved in the L100 weld metal, and the gas porosity associated with the higher O dissolution.

5. Conclusions

The effect of the heat input and electrode-coating basicity index on the chemical composition and microstructure for various locations and mechanical properties of the weld metal, was investigated for three newly developed electrodes: L100, L127 and L208.

The conclusions are:

1. As the basicity index of the coating increases, the level of impurities decreases, however, the oxygen content and gas porosity in the weld increased due to poorer reduction of oxygen in the weld pool.
2. The cooling rate of the weld depends on two factors:
   a) The heat input - the higher the heat input, the lower the cooling rate. Therefore, the cooling rate in the weld metal can be ranked as follows: L100 = L127 >> > L208.
   b) The location in the weld metal affects the cooling rate – the cooling rates can be ranked as: Interface area >> Root area >> Face area.
3. The lower the cooling rate, the longer the time of decomposition of ferrite to austenite, as well as the growth-time of the Widmanstätten austenite grains and the holding time for the intergranular austenite and secondary austenite to form. This results in a change in the microstructure throughout the weld.
   a) In the face area, where the cooling rate was the lowest, the fraction of the austenite was the highest, and this of the ferrite was the lowest. The high heat-input reduced the cooling rate and thus the formation of a secondary phase and intergranular austenite was promoted.
   b) In the interface area, where the cooling rate was the highest, the lowest content of austenite and the highest content of ferrite were observed.
4. In all welds, Widmanstätten austenite, grain boundary austenite, sigma phase and large grains of ferrite were observed. Nitride precipitation has been found only in L208 welds.
5. The study demonstrates that the sigma phase cannot be avoided in the welds. However, the secondary austenite can be avoided, by employing an adequate “low heat input policy”.
6. The higher the austenite levels in the weld, the higher the elongation properties of the weld. The higher the content of secondary austenite and sigma phases in the weld, the lower the yield stress, due to the brittle characteristics of the phases.

Declarations

Author contribution statement

Liraz Atia: Conceived and designed the experiments; Performed the experiments; Analyzed and interpreted the data; Contributed reagents, materials, analysis tools or data.
Menachem S. Bamberger: Conceived and designed the experiments; Performed the experiments; Analyzed and interpreted the data; Contributed reagents, materials, analysis tools or data; Wrote the paper.

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Additional information

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References

[1] K.H. Lo, C.H. Shek, J.K. Lai, Recent developments in stainless steels, Mater. Sci. Eng. R 65 (4-6) (2009) 39–104.
[2] J. Charles, Super Duplex Stainless Steel: Structure and Properties, Conference on DSS ’91, Les Ulis, France, Les Editions de Physique, 3, 1991, pp. 3–48.
[3] R. Gunn, Duplex Stainless Steels - Microstructure, Properties and Applications, Abington Publishing, Cambridge England, 1997, pp. 1–10.
[4] H. Hemmer, Ø. Grong, A process model for the heat affected zone microstructure evolution in duplex stainless steel weldments: Part I, Metall. Mater. Trans. A 30 (1999) 2915–2929.
[5] S. Atamert, R.C. Reed, J.E. King, Modeling of Multipass Duplex Stainless Steel weld deposit Microstructures, Duplex Stainless Steel 91' Conference, France, 1991, pp. 393–402.
[6] J.O. Nilsson, G. Chai, The physical metallurgy of duplex stainless steels, in: Porc. Duplex Stainless Steels, 2010, 2011, pp. 369–390.
[7] J.C. Lippold, Welding Metallurgy and Weldability, Ohio State University, para. 2.3, John Wiley & Sons, Inc., 2015, pp. 13–32.
[8] ASM HANDBOOK, 2015, vol.6, 2015, pp. 55–63.
[9] B. Gideon, L. Ward, D.G. Carr, Structural characterization, residual stress determination and degree of sensitization of duplex stainless steel welds, in: School of Civil Environmental, and Chemical Engineering, RMIT, Melbourne, 2009, pp. 40–59.
[10] U. Dahmen, Orientation Relationships in Precipitation Systems, Materials and Molecular Research Division, Lawrence Berkeley Laboratory, Department of Materials Science and Mineral Engineering, University of California, Berkeley, CA 94720, U.S.A, 1981.
[11] ASTM E562 Standard , Practice for Determining Volume Fraction by Systematic Manual Point Count, ASTM International, West Conshohocken, Pennsylvania, USA, 2002.
[12] S. Wen, Metallurgical evaluation of cast duplex stainless steels and their weldments, in: Materials Science and Engineering, The University of Tennessee, Knoxville, 2001, pp. 47–48.
[13] E.J. Da Cruz Jr., O.D. Franzini, I. Calliari, V.A. Ventrella, Effects of Nickel Addition on the Microstructure of Laser-Welded UNS S32750 Duplex Stainless Steel the Minerals, Metals & Materials Society and ASM International, 2019.
[14] L. Longlong, S. Chumbley, Influence of cooling rate on the ferrite prediction diagram of duplex stainless steel castings, in: The Minerals, Metals & Materials Society and ASM International, 2019.