Towards a map of solidification cracking risk in laser welding of austenitic stainless steels

María-Asunción Valiente Bermejo*, Tarasankar DebRoyb, Kjell Hurtiga, Leif Karlssona, Lars-Erik Svensssonb

aDepartment of Engineering Science, University West, Gustava Melins Gata 2, Trollhättan SE-461 86, Sweden
bDepartment of Materials Science and Engineering, The Pennsylvania State University, University Park, State College PA 16801, USA

Abstract

In this work, two series of specimens with Hammar and Svensson’s Cr- and Ni-equivalents (Creq+Nieq) = 35 and 45 wt% were used to cover a wide range of austenitic grades. These were laser welded with different energy inputs achieving cooling rates in the range of 10³ ºC/s to 10⁴ ºC/s. As high cooling rates and rapid solidification conditions could favour fully austenitic solidification and therefore raise susceptibility to solidification cracking, the solidification modes of the laser welded specimens were compared to the ones experienced by the same alloys under arc welding conditions. It was found that high cooling rates experienced in laser welding promoted fully austenitic solidification for a wider range of compositions, for example specimens with (Creq+Nieq) = 35% under arc welding cooling conditions at 10 ºC/s showed fully austenitic solidification up to Creq/Nieq = 1.30, whilst the same specimens laser cooled at 10³ ºC/s showed fully austenitic solidification up to Creq/Nieq = 1.50 and those cooled at 10⁴ ºC/s showed it up to Creq/Nieq = 1.68. Therefore, high cooling rates extended the solidification cracking risk to a wider range of Creq/Nieq values. This work also compares the cooling rates experimentally determined by thermocouples to the computed cooling rates calculated by a highly-advanced computational model. The distance between the thermocouple’s wires and the thermal resistance of thermocouples together with the small size of the weld pools proved to be practical limitations in the experimental determination of cooling rates. However, an excellent agreement was found between computed and experimental solidus isotherms at high energy input settings. For low energy input settings cooling rate was in the order of magnitude of 10⁴ ºC/s, whilst for high energy input settings cooling rate was found to be in the order of magnitude of 10³ ºC/s.

Keywords: laser beam welding (LBW); Ytterbium fibre laser; austenitic stainless steel; ferrite; solidification modes; cooling rate.

* Corresponding author. Tel.: +34 667776172; E-mail address: asun.valiente@hv.se
1. Introduction

About 1400 million tonnes of stainless steels are produced in the world per year and more than 50% are austenitic stainless steels. It is well recognised that laser beam welding is a very versatile joining process which can be easily automated and is capable to achieve high productivity and quality welds in different alloys. However, there are still some concerns about its use for welding of austenitic stainless steels.

Solidification cracking can be experienced by austenitic stainless steels during welding, in particular when weld solidifies exclusively as austenite. Causes have been extensively studied by Kujanpää et al. (1979, 1980, 1986), Kujanpää (1985), Lippold et al. (1982), Brooks (1992), Li et al. (1999) and Katayama et al. (1985) and it has been demonstrated that during fully austenitic solidification, impurities such as sulphur and phosphorus tend to segregate to the liquid phase and form low melting point eutectics. The distribution and nature of these eutectics along the grain boundaries at the last stages of solidification, together with the solidification shrinkage and the restraining forces are considered to be the main causes of solidification cracking. Therefore, solidification cracking is more likely for fully austenitic (A) and austenitic-ferritic (AF) solidification. In both solidification modes, austenite is the primary solidification phase, but in AF some ferrite is formed in the austenite boundaries because of the eutectic reaction experienced by the last-solidifying interdendritic liquid. Weld metals with primary ferritic solidification modes: ferritic-austenitic (FA) and fully ferritic (F) are less prone to solidification cracking because the solubility of impurities in ferrite phase is higher.

Laser beam welding (LBW), as a low energy input welding process can result in high cooling rates. It is well-known that high cooling rates can promote austenite as primary solidification phase. This phenomenon was observed and studied by Elmer et al. (1990, 1991), Lippold (1994), Fukumoto et al. (1998) and Iamboliev et al. (2003). Therefore, an austenitic alloy that under arc welding conditions solidifies as FA and does not present a risk of solidification cracking, when laser welded can shift to primary austenitic solidification and can become prone to cracking. Consequently, the study of the transition between primary austenitic and primary ferritic solidification modes is of utmost importance under low energy laser welding conditions.

Traditionally, the solidification mode has been related to the parameter chromium equivalent vs. nickel equivalent ratio $C_{eq}/N_{eq}$, (being $C_{eq}= Cr + 1.37Mo$ and $N_{eq}= Ni + 0.31Mn + 22C + 14.2N$ Hammar and Svensson’s equivalents). However, it was recently found by Valiente Bermejo (2012-a) that the coexistence AF-FA depends also on the overall alloy content. For arc welding conditions and an overall alloy content of $(C_{eq}+N_{eq}) = 30$ wt%, the critical $C_{eq}/N_{eq}$ ratio was between 1.38 and 1.55, while in case of $(C_{eq}+N_{eq}) = 40$ wt% the critical $C_{eq}/N_{eq}$ ratio was between 1.28 and 1.32. Previous studies by Katayama et al. (1984), Elmer et al. (1989), Lippold (1994), Fukumoto et al. (1999) and Brooks et al. (2003) were conducted to investigate the effect of low energy welding processes on the transition between solidification modes but none of them considered the effect of the overall alloy content. Further information and comparison among Hammar and Svensson’s equivalents and other Cr- and Ni-equivalents related to the transition between solidification modes and ferrite content prediction was published by Valiente Bermejo (2012-a, 2012-b, 2012-c).

Due to the experimental difficulties of measuring cooling rates in low energy laser welds, traditionally cooling rates have been estimated by different correlations such as the ones proposed by Flemings (1974), Esaka et al. (1988), Katayama et al. (1984) and Volkova et al. (2003). These correlations relate some thermal variables of the process with the resulting dendrite morphology, normally with the Dendrite Arm Spacing (DAS). However, in this study an attempt was made to measure cooling rates experimentally. In addition, an advanced computational method developed by one of the authors has been applied to calculate cooling rates.

Differently from previous work, the transition between solidification modes was evaluated at fixed overall alloy contents $(C_{eq}+N_{eq})$ and related to cooling rate. The combination of chemical compositions and cooling rates that cause fully austenitic solidification in the specimens evaluated, will serve as a basis to prepare a map of solidification cracking risk for laser welding of austenitics. This approach includes both the influences of the alloying content and the cooling rate and will thereby be of great help for the industry in project design stages.
2. Experimental work

2.1. Alloy selection

The overall alloy contents were fixed at \((\text{Cr}_{\text{eq}} + \text{Ni}_{\text{eq}}) = 35\%\) and at 45%. Eight samples with a fixed 35% alloy content but with \(\text{Cr}_{\text{eq}}/\text{Ni}_{\text{eq}}\) ratios from 1.23 to 2.04 were prepared and nine samples with a fixed 45% alloy content but with \(\text{Cr}_{\text{eq}}/\text{Ni}_{\text{eq}}\) ratios from 1.22 to 1.85. The range of ratios was selected so that the theoretical transition between solidification modes predicted by Katayama et al. (1984) was fully covered and included also lower and higher ratios. Each sample was prepared by melting combinations of gas tungsten arc (GTA) wires (AWS SFA5.9 ER310, ER312 and AWS SFA5.18 ER70S-6) with a total batch weight of 50 g using an electric arc furnace according to ASTM E1306-07 in a pure argon atmosphere. The final button-shaped alloys were cut and their chemical composition analyzed by Optical Emission Spectroscopy (Table 1).

| Alloy | C   | Mn  | Si  | S   | P   | Cr   | Ni   | N   | Mo | \(\text{Cr}_{\text{eq}}/\text{Ni}_{\text{eq}}\) | \(\text{Cr}_{\text{eq}} + \text{Ni}_{\text{eq}}\) |
|-------|-----|-----|-----|-----|-----|------|------|-----|----|---------------------------------|-----------------------------------|
| 1     | 0.114 | 1.63 | 0.55 | 0.007 | 0.017 | 19.303 | 12.356 | 0.029 | 0.084 | 1.23                        | 35                                |
| 2     | 0.088 | 1.63 | 0.55 | 0.006 | 0.017 | 19.587 | 12.024 | 0.046 | 0.081 | 1.30                        | 35                                |
| 3     | 0.100 | 1.65 | 0.54 | 0.006 | 0.017 | 20.127 | 11.307 | 0.041 | 0.077 | 1.39                        | 35                                |
| 4     | 0.095 | 1.66 | 0.53 | 0.006 | 0.016 | 20.636 | 10.728 | 0.039 | 0.073 | 1.49                        | 35                                |
| 5     | 0.119 | 1.67 | 0.53 | 0.006 | 0.015 | 20.625 | 10.047 | 0.043 | 0.069 | 1.50                        | 35                                |
| 6     | 0.099 | 1.68 | 0.54 | 0.005 | 0.015 | 21.604 | 9.540  | 0.045 | 0.064 | 1.68                        | 35                                |
| 7     | 0.107 | 1.71 | 0.53 | 0.005 | 0.017 | 22.640 | 8.240  | 0.054 | 0.056 | 1.91                        | 35                                |
| 8     | 0.105 | 1.73 | 0.53 | 0.006 | 0.018 | 23.232 | 7.610  | 0.069 | 0.051 | 2.04                        | 35                                |
| 9     | 0.095 | 1.65 | 0.449 | 0.005 | 0.019 | 24.411 | 16.804 | 0.050 | 0.108 | 1.22                        | 45                                |
| 10    | 0.099 | 1.67 | 0.439 | 0.004 | 0.018 | 24.752 | 16.339 | 0.050 | 0.105 | 1.26                        | 45                                |
| 11    | 0.099 | 1.66 | 0.448 | 0.004 | 0.019 | 25.042 | 16.014 | 0.051 | 0.103 | 1.30                        | 45                                |
| 12    | 0.102 | 1.66 | 0.445 | 0.005 | 0.019 | 25.422 | 15.512 | 0.064 | 0.100 | 1.33                        | 45                                |
| 13    | 0.117 | 1.68 | 0.442 | 0.005 | 0.019 | 25.788 | 15.126 | 0.055 | 0.097 | 1.36                        | 45                                |
| 14    | 0.098 | 1.69 | 0.449 | 0.005 | 0.020 | 25.990 | 14.969 | 0.077 | 0.096 | 1.39                        | 45                                |
| 15    | 0.111 | 1.72 | 0.438 | 0.004 | 0.018 | 26.799 | 13.754 | 0.063 | 0.089 | 1.53                        | 45                                |
| 16    | 0.113 | 1.76 | 0.438 | 0.003 | 0.018 | 28.108 | 12.274 | 0.080 | 0.079 | 1.72                        | 45                                |
| 17    | 0.133 | 1.76 | 0.435 | 0.004 | 0.019 | 29.196 | 10.961 | 0.097 | 0.071 | 1.85                        | 45                                |

2.2. Laser welding procedure

An Ytterbium Fibre Continuous Wave laser (YLR-6000-S) was used for laser welding of the specimens. Pure argon (99.997%) was used as shielding gas.

Some pre-trials were conducted in order to select the combination of laser variables with low (L) and high (H) energy inputs, which would ensure conduction and keyhole welding modes. Table 2 shows the final parameters and settings used for laser welding.

| Settings | Energy input (J/mm) | Welding mode | Collimating (mm) | Focal length (mm) | Foc. pos. (mm) | Spot size in focus (mm) | Fiber ø (μm) | Nominal Power (W) | Welding speed (mm/s) |
|----------|---------------------|---------------|------------------|-------------------|----------------|------------------------|-------------|------------------|---------------------|
| L        | 75                  | Conduction    | 200              | 200               | 0              | 0.8                    | 800         | 750              | 10                  |
| H        | 110                 | Keyhole       | 160              | 300               | 0              | 1.125                  | 600         | 2200             | 20                  |
2.3. Cooling rate determination

Three different methods were used to determine cooling rates in the laser welds, i.e. an experimental determination by using thermocouples, a computational model that considers conduction and convection for the high energy input welds and finally a correlation based on the resulting dendrite morphology.

2.3.1. Experimental determination

Thermal cycles experienced during welding were recorded by K-type thermocouples (Nickel-chromium alloy/Nickel-aluminium alloy) connected to a multi-channel thermocouples module that was regulated by a control system developed in-house and based on LabVIEW® programming. The temperature recording interval was set to 0.01 seconds.

Two commercial alloys were used: stainless steel alloy 304 (UNS S30400) representing the specimens whose alloy content is 35% whilst duplex stainless steel alloy 2205 (UNS S32205) represented the specimens with 45% alloy content.

Four experiments were conducted by combination of the two commercial alloys and the two welding settings (L and H described in Table 2): L-304, H-304, L-2205 and H-2205. Each welding experiment was repeated five times for data consistency and the results presented in the next section are representative for the thermal cycle of each experiment.

Each experiment included four different locations to register temperatures and six thermocouples. Figure 1 shows in full detail the thermocouples’ location and references. At 0.5 mm below the surface in the centreline of the joint, one thermocouple (TC0) was located inside a 1.5 mm ø hole. On the top surface, at 1 mm distance from the centreline two thermocouples (TC2 & TC3) were placed, and also on the top surface but at 2 mm from the centreline two more thermocouples were located (TC1 & TC4). Finally, one thermocouple (TC5) was placed at the centreline of the joint on the top surface. It was decided to double the number of thermocouples initially planned on the top surface, to have better chances to get data at the designated locations.

Fig. 1. Positioning and location of thermocouples.

2.3.2. Computational model

A heat transfer and liquid metal flow model was used to calculate temperature fields and cooling rates for the welding of 304 and 2205 stainless steels under the high energy input conditions (110 J/mm). The model has been extensively tested for the keyhole mode welding of Ta, Ti-6Al-4V, V, 304 stainless steel and a structural steel for various combinations of welding speed and power [by Rai et al. (2007a) and Rai et al. (2008)]. The different shape and size of the weld pool for different materials and welding conditions were satisfactorily predicted by the model. Welding conditions also represented different heat transfer mechanisms, i.e., conduction and convection dominated heat transfer modes in the weld pool. Computational efficiency of the numerical model was achieved by assuming a
quasi-steady state behavior of the keyhole shape and the flow of heat and liquid metal in the weld pool. A detailed
description of the model is available in the literature [Rai et al. (2007b) and Ribic et al. (2011)].

2.3.3. Dendrite morphology correlation
As previously mentioned, some correlations were found for specific grades of stainless steels and are based on the
relationship between some thermal variables of the process and the resulting dendrite morphology, normally the
primary dendrite arm spacing (PDAS) and secondary dendrite arm spacing (SDAS). In this work, Katayama et al.
(1984) correlation [SDAS = 25 (CR)^{0.28}] was used for cooling rate (CR) estimation of alloys 304 and 2205.

2.4. Microstructural characterization
Once the specimens were laser welded they were cut and transverse cross-sections were ground and polished
according to standard metallographic preparation procedures. The etchants used were Lichtenegger-Blöch (at 35-40°C
between 3.5 min. to 4 min.) and electrolytic etching (40% NaOH, 5V, 2 s). Optical microscopy (Leitz Aristomet,
Olympus BX60M and Leica MEF4AM) was used for microstructural characterization.

3. Results and discussion

3.1. Cooling rates
It was not possible to obtain readings from the thermocouple placed in the centerline on the top surface in any
experiment. It was intended to register temperatures upper solidus with that thermocouple, but the high heat density
of the laser beam caused an immediate burn off of the thermocouple.

To avoid a direct contact between the thermocouple and the laser beam but to try to register temperatures in the
melting range, a thermocouple was placed in the centerline of the joint but 0.5 mm below the surface. The distance
below the surface was decided after a preliminary evaluation of the cross-section profiles of the laser welds, being 0.5
mm a convenient depth considering that average depths obtained with the low energy input settings was about 0.4 mm
and with the high energy input about 2 mm. Therefore, it is expected that the thermocouple under the surface could
be in contact with the molten metal in the welds prepared with the high energy input settings and that it could be very
close to the molten metal or in the fusion boundary of the welds prepared with the low energy input settings.

Figure 2 shows the thermal cycles registered during laser welding of alloy 304, representing the 35% alloy
specimens, under high (H) and low (L) energy input. For the low energy input experiments, all four thermocouples at
a distance from the centerline gave readings, while for the high energy input experiments, the signal was lost in two
of them but there were data from one thermocouple at each distance from the centerline.

![Fig. 2. Thermal cycles during welding of alloy 304. Left: High energy input (110 J/mm). Right: Low energy input (75 J/mm)](image)

Figure 3 shows the thermal cycles registered during laser welding of alloy 2205, representing the 45% alloy
specimens, under high and low energy input settings. In both experiments one thermocouple at a distance from the
centerline was not giving any signal. A practical conclusion to be drawn from these experiments is the importance to
place more than one thermocouple in the designated positions to be studied, as there are several experimental reasons that may cause a failure in the thermocouple.

Neither in the H experiment nor in the L experiment, was solidus temperature for alloy 304 (1400°C) or for alloy 2205 (1419°C) registered by the thermocouples. In experiment H for alloy 304 (Fig.2), maximum temperatures registered were 1163°C at 1 mm distance from the centerline on the top surface and 1148°C at 0.5 mm below the surface in the centerline. These temperatures indicate that both locations were not melted and that they belong to or are very close to the fusion boundary, whilst in experiment L for alloy 304 (Fig.2), all temperatures registered indicate that locations are in the HAZ: thermocouples at 1 mm distance from the centerline registered the highest temperatures (738°C and 706°C), maximum temperature at 0.5 mm below the surface was 426°C and at 2 mm from the centerline maximum temperatures were 306°C and 471°C.

Similarly, in experiment H for alloy 2205 (Fig.3), maximum temperatures registered were 1301°C at 1 mm distance from the centerline on the top surface and 1224°C at 0.5 mm below the surface in the centerline. These temperatures indicate that both locations belong to or are very close to the fusion boundary. In experiment L with alloy 2205, all temperatures registered indicate that measurement locations are in the HAZ (Fig.3).

Experimental cooling rates and computed cooling rates for alloy 304 (Cr_{eq}+Ni_{eq}=35%) at high energy input are shown in Table 3. A reasonable agreement was found between the experimental and computed cooling rates at 0.5 mm below the surface in the centerline. At 1 mm from centerline, they look quite similar to the computed cooling rates in the centerline in the range of temperatures between 500-927°C and are in the range of magnitude of 10^3 ºC/s.

Experimental cooling rates and computed cooling rates for alloy 2205 (Cr_{eq}+Ni_{eq}=45%) at high energy input are shown in Table 4. There is a reasonable agreement between the experimental and computed cooling rates in the range of temperatures between 500-1127°C and it is also in the range of magnitude of 10^3 ºC/s. Also an excellent agreement is found between computed and experimental solidus isotherms at high energy input settings for alloy 304 and alloy 2205 (Fig.4-5). The experimentally observed position of the fusion boundary should be compared to the combined maximum extent of the individual isotherms computed at different positions along the weld length.

According to both experimental cross-sections (Figs. 4-5), weld profiles were 2 mm depth and 3 mm width, which mean that the fusion boundary should be found at 1.5 mm from the centerline. Therefore, it was expected that the thermocouples in the centerline at 0.5 mm below the surface and at 1 mm from the centerline registered melting and even upper liquidus temperatures, but thermocouples located in those positions registered temperatures close to or in the fusion boundary. Some reasons might explain these differences, on one hand, the introduction of the hole for insertion of the thermocouple wires affects the local geometry of the weld pool, but more importantly, the practical attachment of the two wires composing a thermocouple involves a minimum separation of the wires that is typically 1-1.5 mm. It needs to be considered that in small laser weld pools like the ones presented in this study for H conditions (2 mm depth x 3 mm width) there is an average temperature gradient of about 900°C/mm between the centre of the weld and the fusion boundary, but that is even more critical for the laser welds in L conditions (1.8 mm depth x 0.4 mm width), as the average temperature gradient is about 6600°C/mm between the centre of the weld and the fusion boundary, therefore, a very short gap between the wires in the attachment of the thermocouples highly influences the temperature registered.
Table 3. Experimental and computed cooling rates for alloy 304 \([\text{Cr}_{eq}+\text{Ni}_{eq}=35\%]\) at high energy input settings

| Type            | Location                                      | Cooling rates below solidus (ºC/s) | Cooling rate solidification (ºC/s) | Cooling rate upper liquidus (ºC/s) |
|-----------------|-----------------------------------------------|-----------------------------------|-----------------------------------|-----------------------------------|
|                 |                                               | 500-800ºC                        | 800-927ºC                        | 927-1127ºC                       | 1127-1400ºC                     | 1400-1454ºC | 1454-1727ºC |
| Experimental    | 0.5 mm below surface, centerline              | 625                              | 1963                             | 1902                             |                                 |            |            |
| Computed        | 0.5 mm below surface, fusion boundary         | 619                              | 1171                             | 1527                             | 2126                             |            |            |
| Computed        | 0.5 mm below surface, centerline              | 647                              | 1238                             | 1835                             | 2698                             | 444        | 2232       |
| Experimental    | 1 mm from centerline, top surface             | 700                              | 1345                             | 854                              |                                 |            |            |
| Computed        | Fusion boundary, top surface                  | 610                              | 1062                             | 2062                             | 4064                             |            |            |
| Computed        | Centerline, top surface                       | 647                              | 1228                             | 1821                             | 3253                             | 398        | 2037       |

Table 4. Experimental and computed cooling rates for alloy 2205 \([\text{Cr}_{eq}+\text{Ni}_{eq}=45\%]\) at high energy input settings

| Type            | Location                                      | Cooling rates below solidus (ºC/s) | Cooling rate solidification (ºC/s) | Cooling rate upper liquidus (ºC/s) |
|-----------------|-----------------------------------------------|-----------------------------------|-----------------------------------|-----------------------------------|
|                 |                                               | 500-800ºC                        | 800-927ºC                        | 927-1127ºC                       | 1127-1400ºC                     | 1400-1454ºC | 1454-1727ºC |
| Experimental    | 0.5 mm below surface, centerline              | 455                              | 1444                             | 2208                             | 1104                             |            |            |
| Computed        | 0.5 mm below surface, centerline              | 655                              | 1234                             | 1841                             | 2681                             | 637        | 2141       |
| Experimental    | 1 mm from centerline, top surface             | 394                              | 1234                             | 1862                             | 1620                             |            |            |
| Computed        | Fusion boundary, top surface                  | 622                              | 1108                             | 1917                             | 4522                             |            |            |
| Computed        | Centerline, top surface                       | 647                              | 1234                             | 1793                             | 3026                             | 638        | 2110       |

Table 5 shows the cooling rates estimated by Katayama’s correlation, and comparing them with the average values of experimental and computed cooling rates (Tables 3 and 4), they are in the same order of magnitude: about \(10^3\) ºC/s for the high energy input welds and about \(10^4\) ºC/s for the low energy input welds.

Table 5. Cooling rates estimated by Katayama’s correlation

| \((\text{Cr}_{eq}+\text{Ni}_{eq})\) | Settings reference | SDAS (µm) | Cooling rate \((C_R)\) (ºC/s) |
|-----------------------------------|--------------------|-----------|-----------------------------|
| 35%                               | H                  | 2.43±0.46 | 4125                        |
| 35%                               | L                  | 1.76±0.40 | 13056                       |
| 45%                               | H                  | 2.55±0.40 | 3473                        |
| 45%                               | L                  | 1.60±0.26 | 18350                       |
### 3.2. Solidification modes

Table 6 shows the solidification modes for the alloys (Cr\(_{eq}\)+Ni\(_{eq}\)=35% and 45%) prepared under three different conditions: first in an electric arc furnace cooled at 10 °C/s (according to ASTM E1306-2007 at 30 V, 550A, 60 s melting time, 50 g of material, 3 min cooling time, Ar shielded), second, laser welded with high energy input settings cooled at about 10\(^3\) °C/s and third, laser welded with low energy input settings and cooled at about 10\(^4\) °C/s.

| Alloy | Cr\(_{eq}\)+Ni\(_{eq}\) | Cr\(_{eq}/\)Ni\(_{eq}\) | 10°C/s | H (10\(^3\)°C/s) | L (10\(^4\)°C/s) |
|-------|----------------|----------------|--------|----------------|----------------|
| 1     | 35             | 1.23           | A      | A              | A              |
| 2     | 35             | 1.30           | A      | A              | A              |
| 3     | 35             | 1.39           | AF/FA  | A              | A              |
| 4     | 35             | 1.49           | AF/FA  | A              | A              |
| 5     | 35             | 1.50           | AF/FA  | A              | A              |
| 6     | 35             | 1.68           | FA     | AF/FA          | A              |
| 7     | 35             | 1.91           | FA     | AF/FA          | A              |
| 8     | 35             | 2.04           | FA/F   | F              | 17             | 45 | 1.85 | FA | FA/F | F |

For Cr\(_{eq}\)+Ni\(_{eq}\)= 35% series and at arc furnace cooling conditions, fully austenitic solidification (A) is found in specimens up to Cr\(_{eq}/\)Ni\(_{eq}\)= 1.30, however, when laser welded fully austenitic solidification is found in a wider range of compositions, as it was found in specimens up to Cr\(_{eq}/\)Ni\(_{eq}\)=1.91. The same trend is observed for Cr\(_{eq}\)+Ni\(_{eq}\)= 45% series: fully austenitic solidification is found in specimens up to Cr\(_{eq}/\)Ni\(_{eq}\)= 1.26 under arc welding conditions, whilst in laser welded specimens the range of compositions is extended up to Cr\(_{eq}/\)Ni\(_{eq}\)= 1.53.

To illustrate the shift in the solidification modes caused by cooling conditions, figure 6 shows the AF/FA solidification modes in alloy 3 under arc furnace cooling conditions whilst figure 7 shows the fully austenitic solidification (A) in the same alloy but cooled at 10\(^3\) °C/s.

The shift in the solidification mode to primary austenitic under rapid cooling conditions was earlier documented in several studies [(Lippold et al. (1994), Fukumoto et al. (1998), Iamboliev et al. (2003), Elmer et al. (1990, 1991)].
stability of austenite as primary solidification phase increases compared to ferrite because of the increased dendrite tip undercooling, and in this work it was proved that laser welding and consequently higher cooling rates promoted fully austenitic solidification for a wider range of Cr\textsubscript{eq}/Ni\textsubscript{eq} ratios, and as previously mentioned in the introduction chapter, fully austenitic solidification makes the specimens susceptible to solidification cracking.

The experimental results presented in this work together with recent investigations involving more series of (Cr\textsubscript{eq}+Ni\textsubscript{eq}) values laser welded at different energy inputs will be a step forward towards the preparation of a map of solidification cracking risk in laser welding of austenitic stainless steels.

Another result worth noticing is that the coexistence AF-FA in arc welded specimens with Cr\textsubscript{eq}+Ni\textsubscript{eq}= 35% was observed at Cr\textsubscript{eq}/Ni\textsubscript{eq} ratios between 1.39 and 1.50 whilst for Cr\textsubscript{eq}+Ni\textsubscript{eq}= 45%, the coexistence is found at lower Cr\textsubscript{eq}/Ni\textsubscript{eq} ratios and at a narrower interval, from 1.30 to 1.39. This trend is in agreement with previous works involving other alloy contents (by Valiente Bermejo, 2012). However, when the same alloys experienced rapid cooling in laser welding, the coexistence takes place at higher Cr\textsubscript{eq}/Ni\textsubscript{eq} values, i.e., from 1.50 to 1.91 for 35% alloying content and from 1.39 to 1.53 for 45% alloying content.

4. Conclusions

- It was proved that high cooling rates promote fully austenitic solidification for a wider range of Cr\textsubscript{eq}/Ni\textsubscript{eq} values thereby making these alloys more susceptible to solidification cracking.
- At higher cooling rates, the transition between primary ferritic and primary austenitic solidification takes place at higher Cr\textsubscript{eq}/Ni\textsubscript{eq} values. However, the transition does not occur at a single Cr\textsubscript{eq}/Ni\textsubscript{eq} ratio and both solidification modes coexist in a range of Cr\textsubscript{eq}/Ni\textsubscript{eq} values.
- A reasonable agreement was found between the experimental, computed and dendrite morphology correlation methods for the high energy input settings: 10^3 °C/s. Dendrite arm spacing was used in the estimation of cooling rates for the low energy welds and 10^4 °C/s was the average value.
- Practical limitations were found in the experimental determination of cooling rates: the thermal resistance of thermocouples and the influence of the distance between the thermocouple’s wires attachment on the temperatures registered.
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