



Hot cracking in laser welding of equiaxed and directionally solidified nickel-based alloy IN792: Experimental and numerical study

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ABSTRACT

In the new sustainability-based scenario emerging in our era, welding, and repairing nickel-based alloys and superalloys play a crucial role in avoiding the replacement of very expensive components. Laser beam welding is an effective technique that can be used for this purpose. However, when considering nickel-based alloys, laser welding is challenging since the integrity of the joints might be compromised by the occurrence of welding cracking.

The present work aims to investigate the hot-cracking susceptibility of the Nickel-based alloy Inconel 792, in both equiaxed and directionally solidified forms. For a deep comprehension of the phenomena, a numerical model was developed to complement the experimental results, revealing that a concentration of stresses related to the solidification process of the fusion zone is mitigated by the intrinsic characteristics of the directionally solidified microstructure. Indeed, the issues related to grain boundary liquation in the heat-affected zone and solidification cracks in the fusion zone are aggravated in the equiaxed joint.

1. Introduction

In the frame of sustainable metallurgy, one of the strategies used to save resources and energy is to enhance the product life, potentially using repairing techniques to avoid a part replacement. This strategy is particularly valuable when applied to nickel-based alloys or superalloys where the cost of the raw material and the production energy are particularly high [1]. Among the repair methods, high energy density welding techniques, such as laser or electron beam welding, have many advantages, such as low invasiveness due to the low heat input, the possibility to deposit filler materials with a unique tool (laser deposition welding), welding of dissimilar alloys, etc. [2,3]. However, some limitations exist, mainly connected to the susceptibility of some alloys to solidification and liquation cracking (SC and LC) as a consequence of the

welding procedure. This is an intriguing challenge in welding and repairing, above all with nickel-based alloys and some aluminum alloys.

SC and LC susceptibility depends on various factors. Liquation cracks occur in the heat-affected zone (HAZ), and the phenomenon is caused by the liquation of low-melting constituents at the grain boundaries due to the proximity of the heat source, thus resulting in intergranular separation. The combined effect of the solidification shrinkage and residual stresses contribute to the cracking mechanism. On the other hand, solidification cracking occurs in the fusion zone (FZ), and it is related to the inability to compensate for the volume reduction in the weld during solidification. Solidification cracks are also intergranular, and they can be either longitudinal or transversal to the weld bead. Usually, SC forms and grows immediately behind the weld pool, in the mushy zone, because of an unfortunate and complex combination of thermal,

Abbreviations: $\beta_{(1,2)}$, Laser-beam absorptivity; BoP, Beads on Plate; BSE, Backscattered Electron; CSZ, Crack Susceptibility Zone; CWM, Computational Welding Mechanics; DS, Directionally solidified; E, Equiaxed; EBSD, Electron Backscattered Diffraction; EDS, Energy Dispersive Spectroscopy; FZ, Fusion zone; $H_{(1,2)}$, Heat source height parameters; HAZ, Heat-affected zone; IPF, Inverse Pole Figure; KAM, Kernel Average Misorientation; LC, Liquation cracking; OM, Optical microscope; PM, Parent material; P, Laser power; p , Material property (generic); PHT, Preheating temperature; $r_{(e,m,i)}$, Heat source radius; SC, Solidification cracking; SE, Secondary Electrons; SEM, Scanning Electron Microscope; v , Welding speed.

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metallurgical, and mechanical factors.

Numerous works are present in the literature about this topic, and many strategies have been proposed to avoid liquation and solidification cracking. With regards to liquation cracking, it must be noted that laser beam welding is particularly appropriate since it induces small HAZ extension, thus reducing the probability of HAZ fissuring [4]. David et al. [5] reported that the microstructure of the parent material (PM) strongly influences the liquation cracking susceptibility, and the authors claimed that a reduction in the grain size has a beneficial effect. A similar conclusion is drawn in [6] where it is suggested to reduce the heat input rather than reduce the welding speed to avoid grain coarsening. Pre-welding heat treatments are also a valuable strategy to control the presence of low melting constituents and unstable secondary phases, although extra care is required to correctly set the parameters [7].

The solidification cracking mechanism is mostly related to the solidification temperature range [5] and the extension of the so-called mushy zone where solidification cracks occur. The length of the mushy zone depends mainly on the composition of the alloy, and it can be reduced by increasing the solidification rate [4]. Hu and Richardson [8], after studying the mechanism of transverse solidification cracking in laser welding of high-strength aluminum alloys, proposed the following mitigating actions: reducing the welding speed and using an additional heat source to reduce the longitudinal thermal stress in the semi-solid region during cooling. Other strategies, described in [9], involve grain refiners, laser beam oscillation, and ultrasonic vibration. Coniglio and Cross made an excellent review of possible causes and mitigating actions against solidification cracking collected in three articles [10–12]. They addressed the influence of welding speed on different phenomena related to solidification cracking, such as stress cells surrounding the weld pool, grain structure, and mushy zone extent. Increasing the welding speed has twofold effects on SC. From one side, it enhances SC by increasing the crack-susceptible zone (CSZ) length, decreasing the time to feed shrinkage, and generating centerline grain segregation. On the other side, it hinders solidification crack formation by shifting the compression cell to the mushy zone, reducing the time exposed to strain, and refining weld metal grains.

The present study focuses on the nickel-based superalloy known as Inconel 792 (IN792), which is distinguished by its excellent strength, creep resistance, and corrosion resistance at elevated temperatures. Consequently, it is frequently used in the production of industrial

turbines and aircraft gas turbine (jet) engines [13,14]. In dealing with these components, damage primarily arises from hot gas erosion and other mechanisms associated with steep thermal gradients and external stresses, making repairing procedures a common recommendation for refurbishing these components. Welding could serve as a repair method, were it not for its high susceptibility to cracking.

In a recent study, Barbieri et al. [15] investigated the laser weldability of directionally solidified (DS) IN792. They achieved sound (defect-free) welds by employing pulsed laser welding at a frequency of 60 Hz, a peak power of 2080 W, a welding speed of 25 mm/s, and a preheating temperature of only 200 °C. In another study, Angella et al. [16] examined the electron beam weldability of IN792, focusing on preheating temperature and welding speed while keeping the other parameters constant (power = 1 kW, acceleration voltage = 50 kV, beam current = 20 mA). They discovered that crack-free joints can be achieved with a preheating temperature of 300 °C and a welding speed of 41.7 mm/s. Preheating the components was identified as a determining factor in the work proposed by Suharno et al. [14] where Gas Tungsten Arc Welding (GTAW) of IN792 parts was investigated. Preheating temperatures varied from 500 °C to 1010 °C to contrast the phenomenon of strain age cracking, which is correlated to the embrittlement of the alloy within specific temperature ranges.

Fig. 1 summarizes laser welding bead-on plate (BoP) trials conducted by varying both laser power and welding speed on both IN792 plates with different thicknesses and different microstructures (directionally solidified, DS, and equiaxed, E, grain structure). The data are taken from Ref. [15,17] and from other experimental campaigns conducted by some of the authors of the present work. Process parameters summarized in Fig. 1 guaranteed a full penetration through both 2 mm and 3 mm sheet thicknesses, which means that the process window for a fully penetrated weld was quite wide. It is worth noting that despite the lower preheating temperatures than those suggested in [14], DS IN792 alloy resulted in sound (crack-free) specimens. An issue not yet solved in literature is, therefore, the higher laser weldability of directionally solidified IN792 alloy compared to that with an equiaxed structure, despite the same nominal chemical composition. In the following sections, IN792 welded joints with DS and E microstructure welded with the same process parameters are thoroughly compared to investigate the reasons behind the different crack susceptibility of the two microstructures. Moreover, to better understand the influence of the laser welding process on the hot cracking susceptibility of the IN792 alloy, a numerical model was

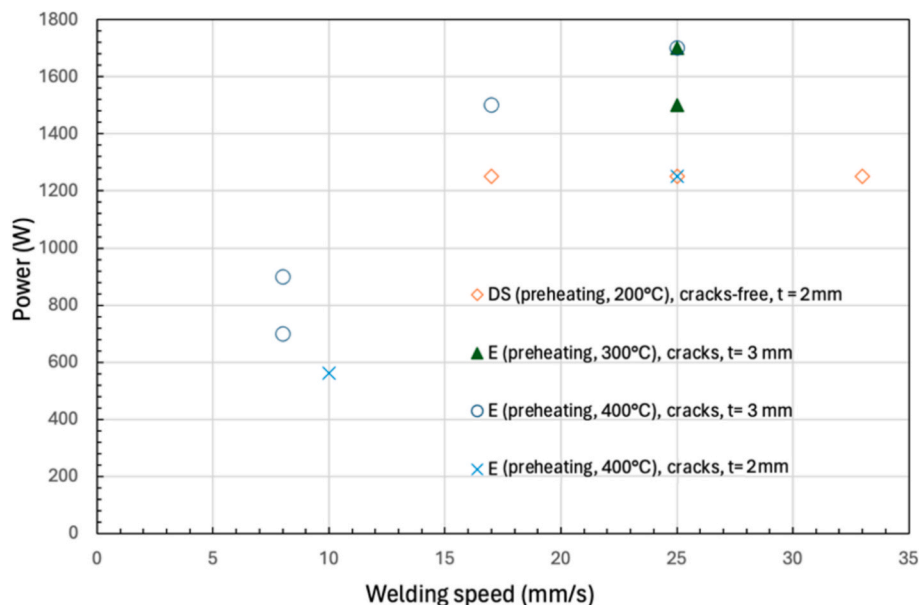


Fig. 1. Results of preliminary BoP tests conducted by the research group and partially presented in [15,17].

developed and used to interpret the experimental results for both equiaxed and directionally solidified microstructures.

2. Experimental and numerical methods

2.1. Material and experimental setup

The analyzed material is the nickel-based alloy IN792, which was processed to obtain two different microstructures: the first consists of equiaxed grains (E IN792), while the second consists of elongated grains obtained via directional solidification (DS IN792). The materials were supplied in the form of cylindrical ingots ($\Phi = 25$ mm, $L = 60$ mm), and plates with a thickness of 2 mm were obtained by spark-erosion cutting along the ingot axis. For the DS alloy, the [001] growth direction of the columnar grains coincides with the cylindrical axis. Laser welding was conducted with the solid-state Yb fiber laser IPG YLS-4000 (IPG Laser GmbH, Burbach, Germany). In the case of DS alloy, welding trials were carried out on the plates using a continuous laser beam traveling along the longitudinal direction *i.e.* parallel to the solidification direction. A proper clamping system equipped with a heating table in Titanium PZ28-3 T by Harry Gestigkeit GMBH (2000 W) was designed to carry out the experiments by setting a preheating temperature. To reduce the amount of porosity, the laser power was modulated, holding the maximum power $P_{\max} = 2000$ W for 10 ms and the minimum power $P_{\min} = 0$ W for 6 ms to obtain an average power $P_{\text{avg}} = 1250$ W. The welding speed (v) was set to 25 mm/s and the preheating temperature PHT was 300 °C. The choice of using 300 °C as preheating temperature is related to the results of other investigations on the weldability of IN792 with high energy density techniques, such as [15,16]. Argon was used as a shielding gas flowing on the top and bottom surfaces of the plates. The same process parameters, summarized in Table 1, were used for welding the DS and E plates.

The welded plates were cut using a rotating cutting machine to obtain a cross-section (*i.e.* section perpendicular to the welding direction) and a longitudinal section for each microstructure condition, for a total of four sections that were embedded in hot conductive resin. A schematic of the sections evaluated can be found in Fig. 4. The specimens for microstructure investigations were then prepared using a standard metallographic procedure involving grinding with abrasive papers with a grit size up to P2400, followed by polishing with a 6- μm diamond paste and finally polishing using colloidal silica suspension (0.04 μm abrasive size).

Mirror polished surfaces were etched with the Marble's reagent (10 g $\text{CuSO}_4 + 50$ mL $\text{H}_2\text{O} + 50$ mL HCl) to reveal the macrostructure of the samples. Afterward, the specimens were polished again and exposed to an electrochemical etching (12 ml H_3PO_4 , 40 ml HNO_3 , and 45 ml H_2SO_4 , with 6 V DC for 10 s [18]) to detect secondary phases and precipitates. The microstructural observations were conducted using the optical microscope (OM) Leica® CTR6 and the scanning electron microscope (SEM) FEI® Quanta 250. Elemental segregations and chemical composition of the secondary phases were assessed through the energy dispersive spectroscopy (EDS, Thermofisher® UltraDry X-Ray detector) analysis, while the grain orientation was investigated by using the electron backscattered diffraction (EBSD, EDAX® Hikari camera). The indenter FutureTech® microhardness tester FM810 was used to obtain Vickers's microhardness profiles on the polished cross-section with a load of 500 g.

Table 1
Parameters adopted to weld the plates.

Power mode	P_{avg}	v	PHT	Thickness
Pulsed	1250 W	25 mm/s	300 °C	2 mm

2.2. Numerical model

The computational welding mechanics (CWM) approach was used, where volumetric heat sources are commonly used to describe the laser penetration into the alloy. In this work, a combination of two conical shape heat sources, with Gaussian power density distribution, is used to reproduce the molten pool shape observed in experiments. According to the schematic shown in Fig. 2, the analytical expressions of the two volumetric heat sources are:

$$Q_1(x, y, z) = \frac{6\beta_1 \bullet P}{\pi H_1 (r_e^2 + r_m r_m + r_m^2)} \exp\left(-2 \frac{r^2}{r_0^2}\right) r_0(z) \\ = r_e - (r_e - r_m) \bullet \frac{(z_e - z)}{(z_e - z_m)} z_e < z < z_m \quad (1)$$

$$Q_2(x, y, z) = \frac{6\beta_2 \bullet P}{\pi H_2 (r_m^2 + r_m r_i + r_i^2)} \exp\left(-2 \frac{r^2}{r_0^2}\right) r_0(z) \\ = r_m - (r_m - r_i) \bullet \frac{(z_m - z)}{(z_m - z_i)} z_m < z < z_i \quad (2)$$

where r_e , r_m , and r_i are the radius at the top, middle, and bottom of the plate, respectively; $r = \sqrt{\xi(t)^2 + \psi(t)^2}$, with ξ and ψ coordinates of the local systems moving in the xy plane along the welding path at a constant laser source speed v_0 ; z_e , z_m , z_i are the upper, medium, and lower planes of the source, respectively. Note that the welding path is described by the position as a function of time of the local reference system described by the coordinates $(x_0(t), y_0(t))$. For a straight welding path along the y direction, $x_0(t) = \text{cost}$ and $y_0(t) = Y_0 + v_0 t$ (with Y_0 initial position of the source). The same power P is defined in both the sources given by Eqs. (1) and (2). However, the total absorbed energy (βP , where β is the laser-beam absorptivity) is split into two parts by

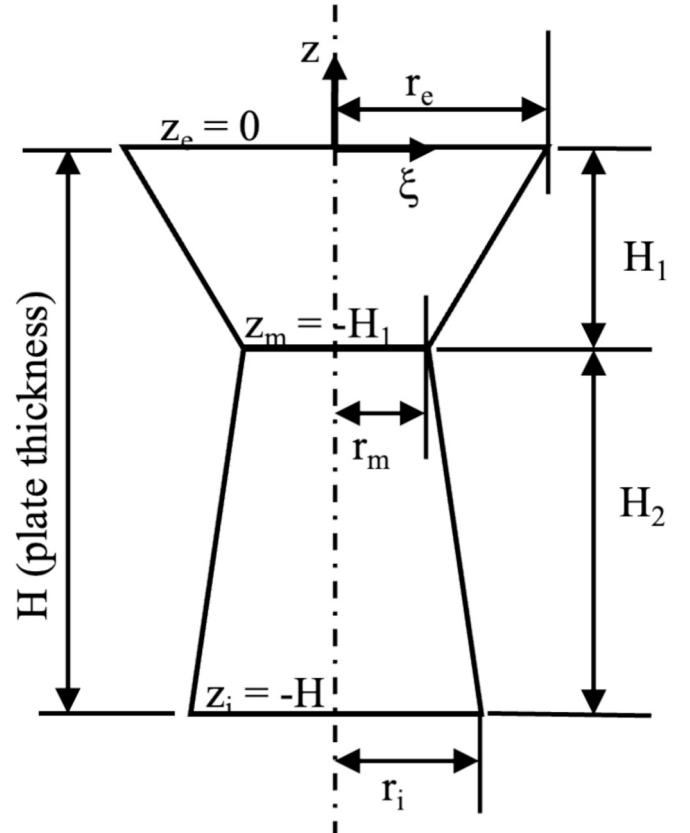


Fig. 2. Volumetric heat source superposition and local reference system moving along the y direction at a speed of v_0 .

using different absorption coefficients for the two conical heat sources:

$$\beta P = (\beta_1 + \beta_2) \bullet P \quad (3)$$

After convergence analysis, the best mesh configuration used in the numerical model is shown in Fig. 3. For geometrical and loading symmetry reasons, only one-half of the joint was modeled using 16,100 8-node brick finite elements. To capture the high thermal gradient induced by the high-power density welding process, a mesh with a graduated density was adopted.

The elements were divided into groups (called *Bead* and *Parent Metal*) on which different material properties were applied to capture the effects of the isotropic and/or anisotropic behavior of the material on thermal and mechanical results.

The adiabatic condition was applied to the symmetry plane, while the convective heat transfer coefficient (h_c) and the emissivity of the material were set to 25 [W/(m²K)] [19] and 0.7, respectively. Physical and thermal material properties were taken as a function of temperature from the ThermoClac® database and from reports of experimental tests previously carried out on DS IN792 and reported in the Appendix. To consider the latent heat produced during the liquid-to-solid transformation (and vice versa), the apparent heat capacity approach was used [20].

In the frame of CWM, the temperature history calculated at each node of the numerical model is used as input load for the mechanical computation (uncoupled thermo-mechanical analysis). In this simplification, the heat generation due to plastic deformation is neglected, being much lower than the heat induced by the laser heat source. The kinematic strain hardening model was chosen, and the clamping condition was imposed as isostatic. Thermomechanical properties as a function of temperature for the E IN792 were taken from the report of Nickel Institute [21] and the work by Du et al. [22]. The mechanical properties as a function of the temperature of the directionally solidified IN792 were taken from the reports in the Appendix. It is important to mention that no distinction was made between the material properties of the Bead and Parent Metal of IN792 in the equiaxed form, while a different longitudinal (y direction) thermal conductivity, Young's modulus, and yield stress were applied to the Parent Metal compared to the Bead of the DS IN792. This choice comes from the fact that the equiaxed microstructure is naturally isotropic, while the DS alloy presents an anisotropic component that cannot be neglected. Ideally, the fusion and recrystallization of the elements in the bead promote a grain distribution with no preferential orientation, thus isotropic, that is not

dependent on the microstructure of the parent material. Despite this assumption, the microstructure analysis conducted in the fusion zone of DS IN792 that is reported in the following section revealed a fair share of grains in the fusion zone of the DS joint that follows the orientation of the grains in the parent material. Hence, it can be argued that the material properties p of the bead can be calculated using the 'rule of mixtures' as follows:

$$p = fp_E + (1 - f)p_{DS} \quad (4)$$

where p_E and p_{DS} are the material properties of the alloy in the equiaxed and directionally solidified form, respectively, and f is the fraction of equiaxed grains with random orientation detected in the fusion zone of DS IN792. In the present work, it was found that $f = 0.4$ (approximately). Finally, to consider the effect of the fusion zone, a function that sets to zero the material properties of those nodes that overcome the liquidus temperature (1350 °C) was implemented into the model. Sysweld® numerical code was used to build and solve the model.

3. Results

3.1. Experimental results: Sound and unsound specimens

Fig. 4 presents representative images of the sections investigated. The cross-sections of the welds are displayed before (Fig. 4a and 4c) and after etching (Fig. 4b and 4d). The images reveal that the FZs are characterized by finer grains compared to the parent material, growing in the direction of the thermal gradient, i.e., from the boundaries of the FZ to the center of the welds. Numerous carbides in the parent material are visible, especially in Fig. 4a and 4c, which were taken by acquiring backscattered electrons (BSEs), to reveal compositional differences in a sample surface. At this magnification, the carbides appear like fine particles dispersed in the matrix, and they can be distinguished by their shiny appearance. Some of them are indicated with yellow arrows. The hourglass shape of the FZ is highlighted in Fig. 4b and 4d, which show the etched sections. According to [23], the hourglass shape is typical of laser welding conducted in keyhole mode with open roots, while Sekhar et al. [24] attributed the hourglass shape to the blooming effect of the laser due to the high power involved. Moreover, the hourglass shape can be determined by the surface tension of the melt pool [25]. Indeed, hourglass weld pools in welded IN792 were also reported in [15]. Concerning the evaluation of the welds' soundness, many cracks were detected. Some of these are indicated with red arrows and numbered, as

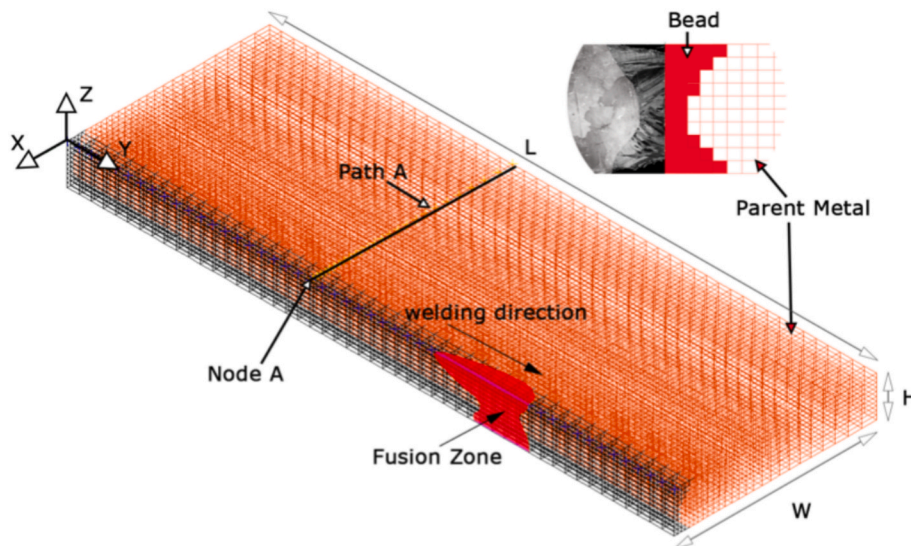


Fig. 3. Numerical model (mesh) with the reference system, geometrical parameters ($L = 30$ mm, $W = 10$ mm, $H = 2$ mm), and groups of elements (say, Bead and Parent Metal).

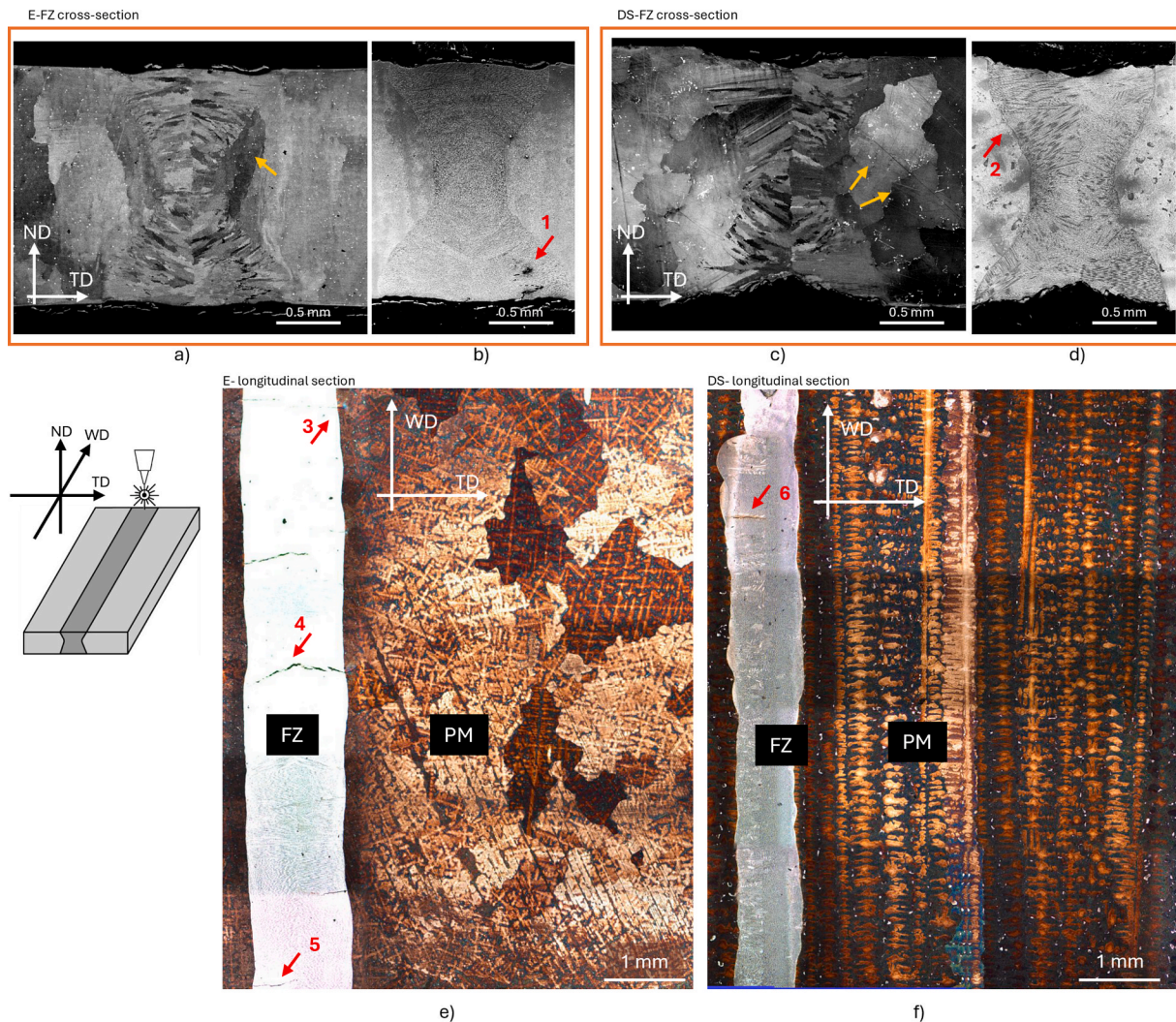


Fig. 4. Microstructures of the E and DS welds. a, c) BSE-SEM micrographs of E and DS IN792 polished cross-sections; b, d) SE-SEM micrographs of E and DS cross-sections etched with the Marbles' reagent; e, f) OM micrographs of E and DS longitudinal sections etched with Marble's reagent. The red arrows with numbers indicate some of the cracks that will be investigated in the following sections. The yellow arrows indicate some carbides. To clarify the orientation of the sections, a schematic with the corresponding reference system have been added. WD stands for Welding Direction, ND for Normal Direction, and TD for Transversal Direction.

they will be discussed in the upcoming sections. The E IN792 weld has good penetration with only a slight undercut defect on the top surface. A large crack (crack 1) is visible in the lower part of the FZ. The DS IN792 weld appears underfilled on both the top and bottom surfaces. A thin crack (crack 2) can be seen in the HAZ.

Optical micrographs of the etched longitudinal sections are shown in Fig. 4e and 4f. As expected, the microstructure of the equiaxed IN792 sample shows dendritic grains with no preferential growth directions. On the contrary, the grains in the DS material are columnar and extend parallel to the welding direction across the whole plate. As previously mentioned, the E weld presents a conspicuous number of cracks, while the DS can be considered a sound joint, although one crack was detected (crack 6). The weldability of the materials was simply evaluated by measuring the total length of cracks over a 10 cm coupon of weld bead: the total crack length in the E weld is 6.3 cm, while 0.51 cm in the DS one.

3.2. Anisotropy and grain orientation

The Inverse pole figure (IPF) maps obtained from the EBSD analysis are reported in Fig. 5. The different microstructures in the parent material of E and DS IN792 are shown in Fig. 5a and 5b. As explained in

[26], the mechanical behavior of a DS component is anisotropic, which means that properties such as tensile properties and thermal properties differ depending on the direction considered. The differentiation involves the direction parallel to the grain growth, named longitudinal, and the directions perpendicular to the grain growth, which will be named transversal. Ideally, the transversal properties can be considered isotropic [26] since the orientation of the columnar grains aligned along the [001] direction is not constrained. This is corroborated by the IPF map in Fig. 5b, where the elongated grains are depicted with different colors, which means different orientations, while in Fig. 5f, which shows the cross-section of the DS IN792, all the grains outside the fusion zone are represented with the red color, corresponding to the same [001] grain orientation. The IPFs presented at the bottom of Fig. 5 also prove that in the E weld, there isn't a strongly favored grain orientation, while in the DS weld, high intensity is calculated corresponding to the [001] orientation, especially on the parent material.

The microstructure of the fusion zone can be observed from the IPF maps of the longitudinal sections (Fig. 5c and 5d) and cross-sections (Fig. 5e and 5f). These areas are mainly composed of small grains elongated along the thermal gradient direction and growing epitaxially from the partially melted grains of the parent material although massive grains are present in the beads in both the E and DS welds; indeed, it is

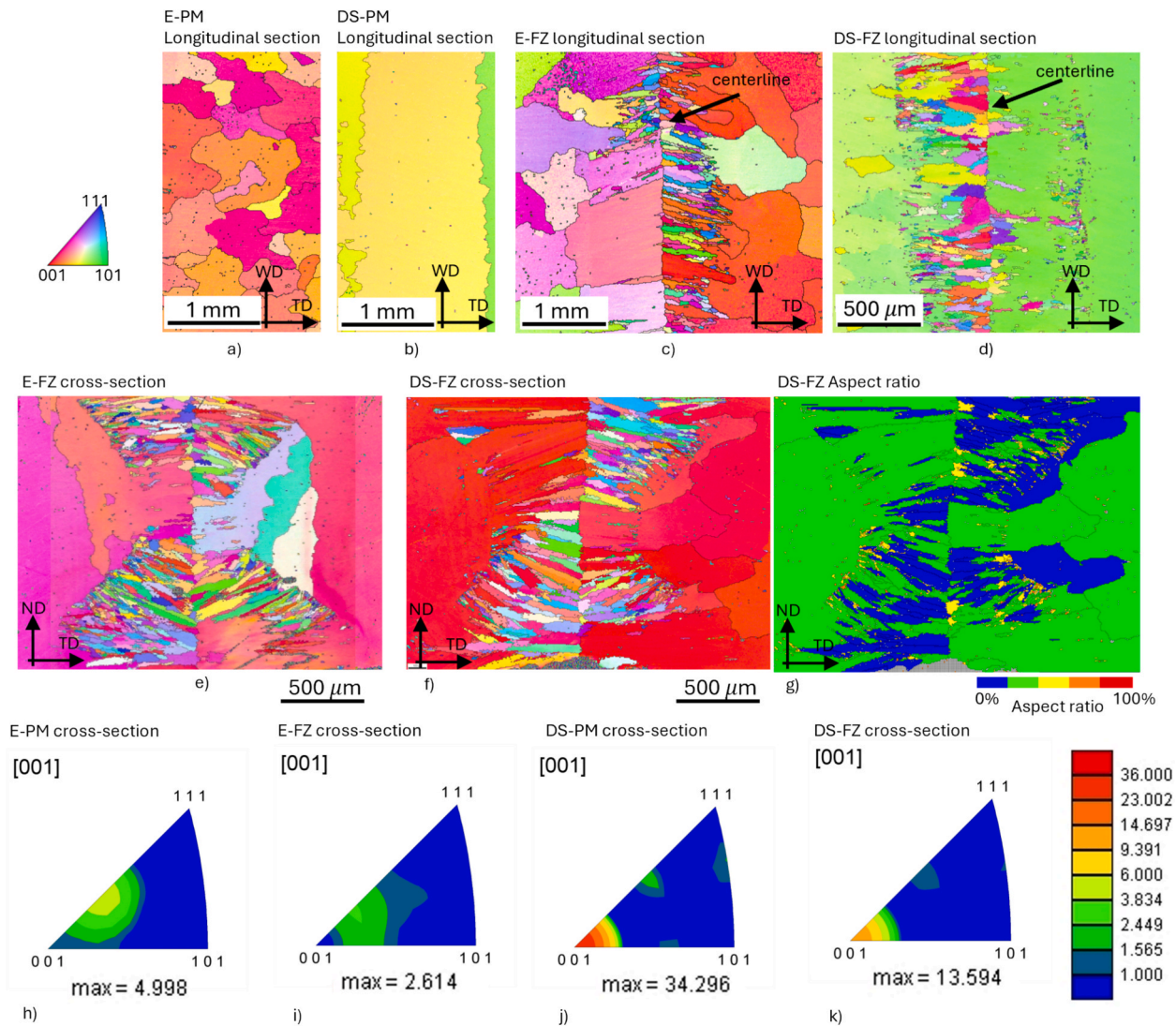


Fig. 5. IPF maps obtained via EBSD analysis. a) E longitudinal section- PM; b) DS longitudinal section- PM; c) E longitudinal section- FZ; d) DS longitudinal section- FZ; e) E cross-section; f) DS cross-section; g) DS cross-section- grain shape aspect ratio map. IPFs in the PM and FZ of the E weld (h, i) and DS (j, k).

difficult to recognize the boundaries of the fusion zone. To help the reader, the centerlines are indicated. Note for example, the region mapped in Fig. 5d shows that only the left side of the bead experienced recrystallization, while the right side is mostly composed of a unique grain (depicted in green). It must be noted that the number of grains sectioned depends on the depth of the polishing procedure. The interface between FZ and HAZ is underlined in some areas by small equiaxed grains. The grains aspect ratio map of the DS cross-section is shown in Fig. 5g to highlight the presence of these grains, marked in yellow. The same effect was noticed on the E cross-section, which, for the sake of brevity, was not reported. This effect is probably related to the so-called 'chill zone', which refers to the shell of fine equiaxed grains in a cast ingot corresponding to the surfaces in contact with the mold, characterized by an elevated thermal gradient. Moreover, the fusion zones observed from the cross-section (Fig. 5e and 5f) suggest that in comparison to E, the DS fusion zone is rich in 'red grains', i.e. grains oriented in [001]. The software Image J was used to quantify this difference, revealing that 58 % of the fusion zone in the DS weld is occupied by 'red grains', compared with 20 % of the E fusion zone. This difference is believed to be important in the definition of the material properties required by the numerical model proposed in the previous section. Several works available in the literature attempted to quantify the stiffness of a single crystal of nickel-based superalloy along [001], [111],

and [101] directions, revealing that Young's modulus along the [001] direction is lower in comparison to the others [26–33]. Although additional issues come into play in a polycrystalline alloy, the large share of [001] grains in the DS fusion zone cannot be overlooked, and expedients such as the mixture rule can be useful to quantify the material properties of the fusion zones, as described in section 2.2.

3.3. Microstructural and chemical analysis of the parent material

The chemical composition of the material was assessed through EDS analysis performed on wide areas sufficiently far from the FZ and the HAZ to avoid local concentrations of segregated elements. The results, reported in Table 2, show no significant differences in the chemical composition of the E and DS alloy, suggesting that crack susceptibility of the E weld cannot be ascribed to chemical issues. For the sake of completeness, the nominal composition of IN792 alloy is also reported in the table. It is worth noting that the discrepancy with the estimated Carbon contents is due to the EDS technique which is not accurate in measuring low atomic number elements in small amounts [34].

Precipitated-hardened Ni-based superalloys, such as IN792, owe their excellent mechanical properties at both room and high temperatures to the precipitation of the secondary phase γ' in the FCC matrix γ . The γ' phase is an ordered intermetallic ($\text{Ni}_3\text{Al}(\text{Ti})$) compound with an

Table 2
Chemical composition in wt% of IN792.

wt%	E	DS	Cast [21]
C	1.66 ± 0.19	1.67 ± 0.10	0.12
Al	3.05 ± 0.06	3.28 ± 0.06	3.1
Si	0.25 ± 0.08	0.26 ± 0.10	—
Ti	3.96 ± 0.08	3.97 ± 0.13	4.5
Cr	12.13 ± 0.24	12.13 ± 0.13	12.4
Co	8.81 ± 0.24	8.78 ± 0.24	9
Ni	59.01 ± 0.79	58.87 ± 0.54	61
Mo	1.78 ± 0.07	1.79 ± 0.08	1.9
Ta	4.90 ± 0.7	4.80 ± 0.69	3.9
W	4.47 ± 0.39	4.38 ± 0.34	3.8
Others	—	—	0.12

FCC structure (L12) that is usually coherent with the γ matrix [35]. After the electroetching of the cross-sections, it was possible to study the characteristics of γ' as well as other phases such as carbides and eutectic regions.

Fig. 6 collects representative micrographs of the microstructure observed in the FZ, HAZ, and PM of the equiaxed cross-sections. The same features can be found in the DS.

Fig. 6a shows the FZ where traces of dendrites are easily identified, as well as very fine shiny carbides. Thin cracks, as the one depicted here, were often found in the fusion zone, running along the dendrites. The transition between FZ and HAZ can be observed in Fig. 6b. The interface is characterized by the total absence of carbides and dendrite traces. Possibly, this region corresponds to the equiaxed fine grains that were observed in Fig. 5g at the boundaries of the FZ. In the HAZ, white particles appear. These particles are γ' particles with an undefined shape since the vicinity of the heat source has partially dissolved them. In the FZ, no signs of γ' were observed. This is attributed to the rapid cooling induced by the laser welding technique that does not allow sufficient diffusion of the elements to form precipitates. Nonetheless, in other

studies, very fine spherical γ' was observed in the FZ as well [15]. The aspect of γ' can finally be observed in Fig. 6c, showing the typical appearance of the parent material. The size and density of γ' change between interdendritic and dendritic core regions because the interdendritic region is richer in the forming elements of γ' (Al and Ti), promoting the coarsening of the precipitate [36]. The grain boundaries are indeed often decorated by large γ' particles, as depicted in the picture. Fig. 6c also shows a Ti/Ta carbide and a 'crown' shaped γ - γ' eutectic region. Eutectic regions are often found with similar shapes [37] and the origin is traced back to the solidification process, as a result of the segregation of elements such as Al and Ti in the interdendritic region during dendrites growth [38]. Indeed, eutectic regions were often found in proximity to grain boundaries.

The variation in the shape and size of γ' is better investigated in the second row of Fig. 6, showing pictures of areas at increasing distances from the FZ. In Fig. 6d, next to the FZ, γ' is partially dissolved due to the temperature rise. According to [36], the material surrounding γ' has the γ - γ' eutectic composition. An elevated heating rate, expected during laser welding, does not allow a complete dissolution of the precipitate phase, promoting the insurgency of areas with eutectic composition. The presence of eutectic in proximity to the FZ extends the temperature solidification range of the material. Hence, the uncompleted dissolution of γ' is one of the main reasons behind the mechanism of boundary liquation cracking. Incipient dissolution is also visible in Fig. 6e, which was taken approximately 90 μm from the FZ boundary of the E cross-section. Finally, far from the FZ (approximately 150 μm), the γ' shape is sharp presenting either the shape of cubes or octocubes. The shape of γ' depends on the thermal history of the component, primarily on the growth rate R and on the chemical composition of the material. According to [39], the octocubes, sometimes called butterfly shape [40] or ogdoadically-diced cubes [42], are a result of the coarsening of cubic γ' since the cubes tend to divide to maintain low interfacial energy [41]. Further coarsening from this shape would imply a shift from the

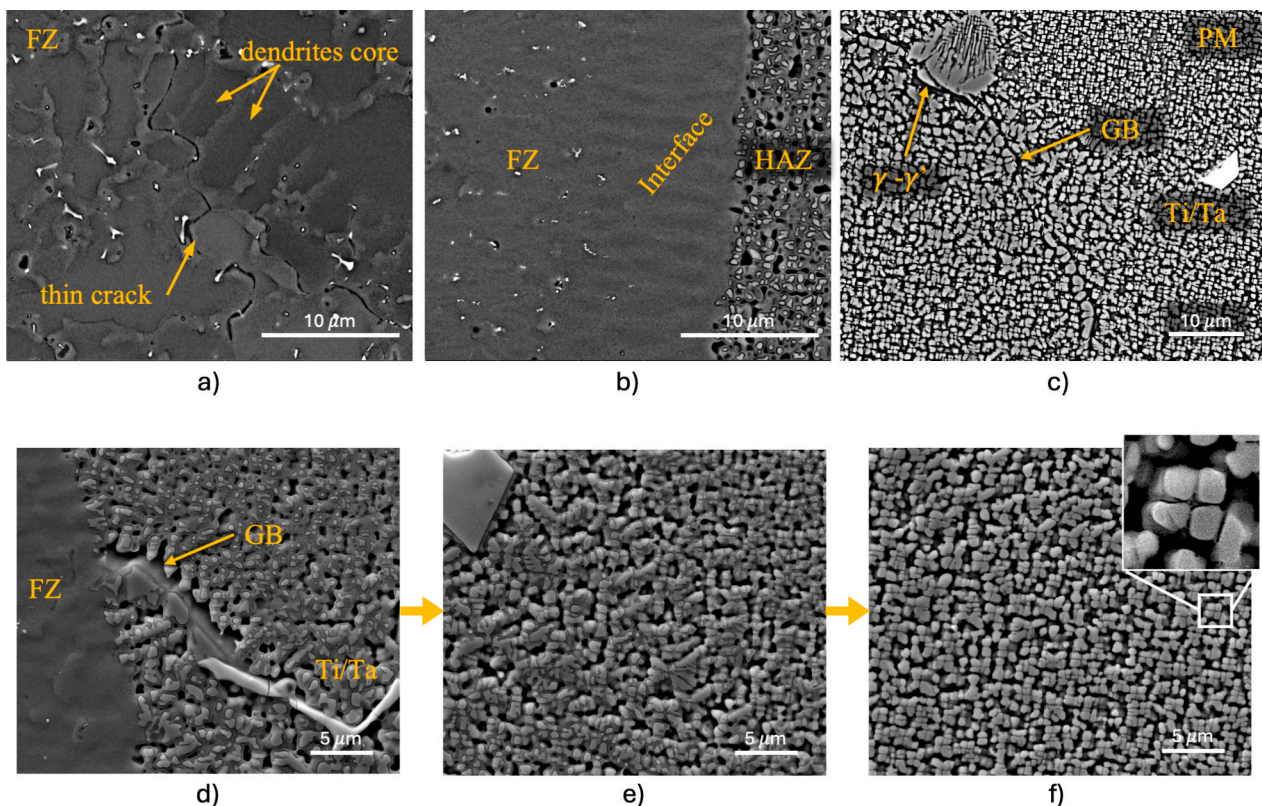


Fig. 6. SEM pictures of some microstructure details captured in the electro-etched cross-section of the equiaxed weld. a) fusion zone FZ; b) interface between FZ and heat-affected zone (HAZ); c) parent material (PM); d-f) evolution of γ' shape with increasing distance from the fusion line.

interfacial growing mechanisms to the diffusion mechanisms, promoting a dendritic growth that changes the nature of the precipitate from coherent to incoherent with the matrix, thus losing the benefits associated with γ' in terms of strengthening [39,42]. Note that no dendritic γ' was observed in the samples investigated in this work.

3.4. Carbides analysis

According to the literature, IN792 alloy is likely to form MC-type carbides rich in Ti and Ta [43], whose morphology depends on the chemical composition of the parent material and the solidification process [44]. Observing Fig. 7a and b, it can be noticed that the carbides in the E plates have a globular ‘blocky’ shape, while in the DS plates, they present an irregular slender shape commonly referred to as China script. Since there are no differences in the chemical composition (see Table II), the different morphology observed is only to be ascribed to the solidification process. DS superalloys typically present China script carbides that are associated with the high growth rate that directional solidification entails. Blocky carbides are common in equiaxed as-cast components since they grow in near-equilibrium conditions [43]. It has been proven that increasing the growth rate R in a thermal process leads to the formation of more carbides and provokes the shift from blocky to script shape [38,44]. The chemical composition of the carbides is the same in both the E and DS plates; therefore, Fig. 7c reports for simplicity only the EDS maps obtained for the blocky carbides found along a grain boundary in the equiaxed structure. The particles lack Ni, Cr, and Co, the three main alloying elements of IN792, while they are rich in Ti and Ta. It is interesting to notice that Ti accumulates in the interdendritic areas, where the carbides are more numerous. Indeed, the availability of carbide-forming elements is also a primary factor in the

carbide size [38,44] thus, segregations at grain boundaries create favorable locations for carbide growth.

In the FZs, the melting of the material produces the almost complete dissolution of the carbides. Nonetheless, some small blocky carbides can be seen in both the E and DS fusion zones (Fig. 7d and g), these particles are thought to be the core of pre-existing carbides that survived the fusion [45]. The fusion zone of the E sample is slightly richer in carbides’ fragments, which can be explained considering that blocky carbides are massive in comparison to China script and, thus less prone to dissolve. A higher magnification picture (Fig. 7f) shows that very fine strip-like carbides are also diffused in the FZ of both materials. Indeed, the fine microstructure in the FZ results in less interdendritic space available for the carbides to grow, thus forcing a transition to a finer rod-like morphology. Interestingly, the EDS revealed that the microsegregation of Ti in the interdendritic areas in the FZ is enhanced (Fig. 7g).

3.5. Microhardness measurements

Microhardness profiles were performed on the polished cross-sections approximately in the middle and near the top section of the hourglass fusion zones (Fig. 8). Considering only the indentations on the parent material, far from the FZ and the HAZ, the average HV0.5 calculated for the E and DS plates is 427 ± 8 and 405 ± 16 HV, which is in good accordance with what was calculated in [15,16]. The difference in the hardness is not significant, although it could be explained by the abundance of grain boundaries in the equiaxed weld. The FZs of both E and DS welds present peaks of hardness, reaching a maximum of 500 HV0.5, which again can be attributed to the refinement of the grains in the FZ in comparison to the parent material. In [15], it is reported that the presence of very fine spherical γ' in the FZ is responsible for the

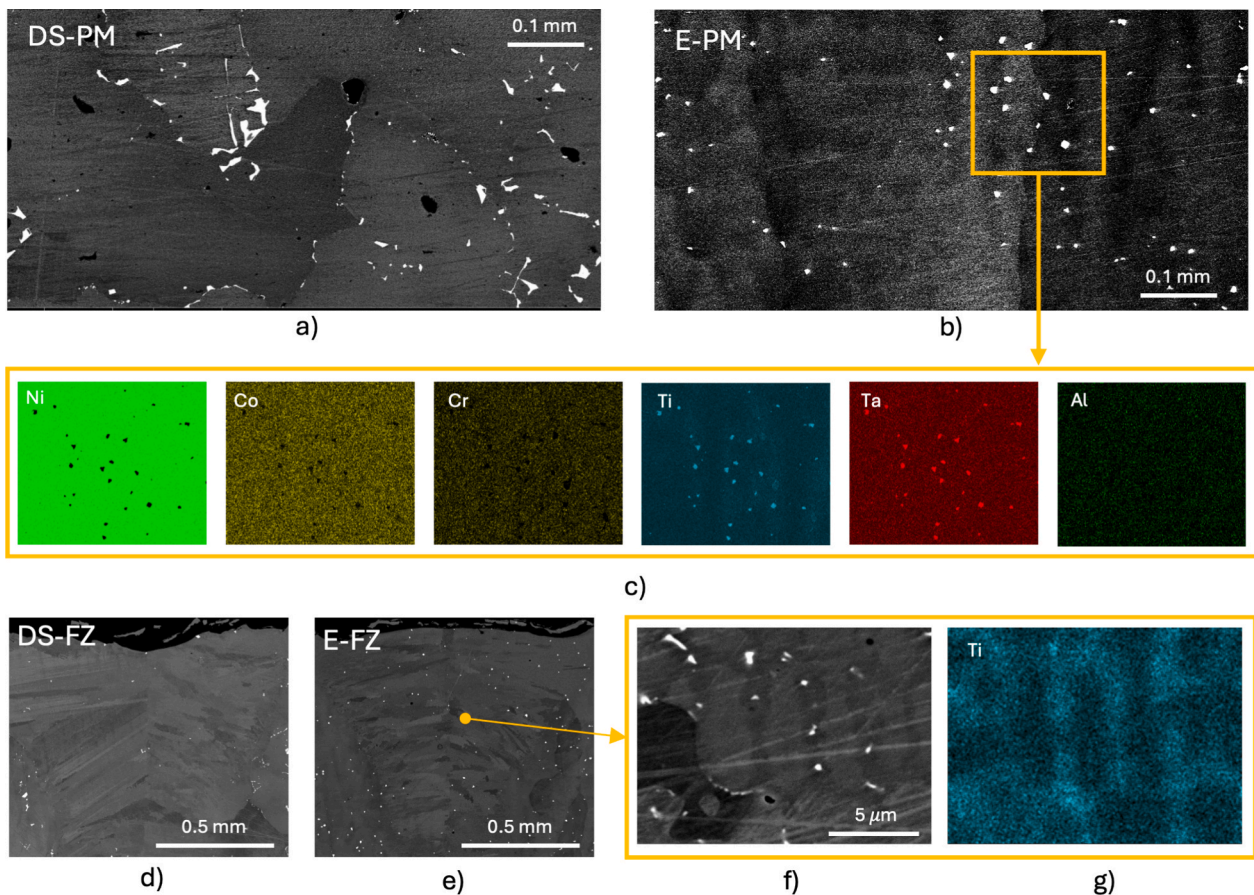


Fig. 7. Investigation of the carbide's morphology. Typical appearance of the parent material of the DS plates (a) and E plates (b); c) EDS element maps of a detail of Fig. 7b; magnifications of the fusion zones of the DS weld (d) and E (e); f) detail of the FZ of the E weld and g) corresponding EDS map of element Ti.

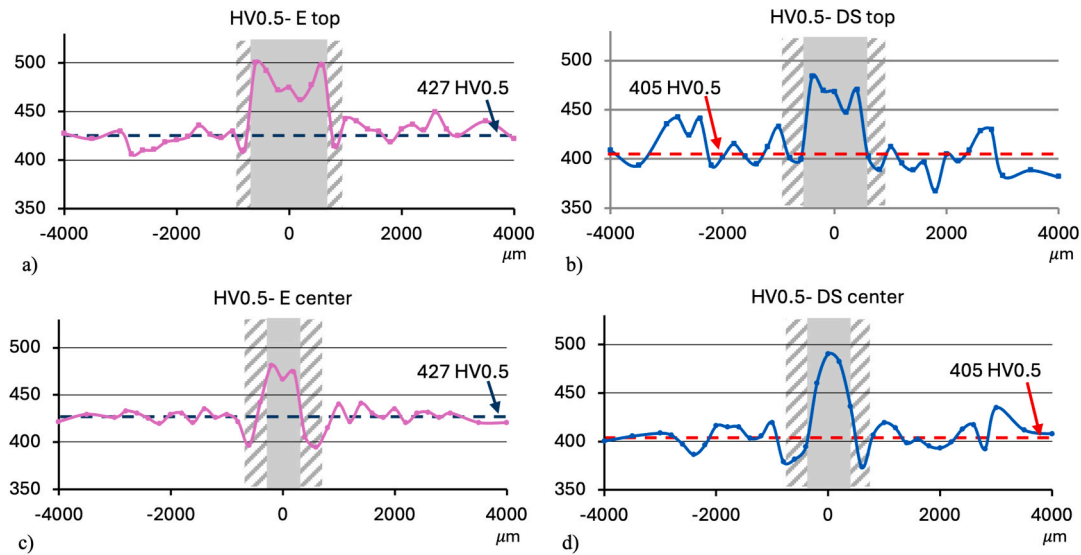


Fig. 8. Microhardness profiles taken close to the top surface (a, b) and middle section (c, d) of the E (pink) and DS (blue) cross-sections. Average values of the parent material hardness are indicated with dashed lines. The fusion zone extension is represented with a grey region, while the HAZ has a dashed pattern.

enhanced hardness. As already mentioned, no such a γ' distribution was found in the FZ, although this might be related to the limitations of the equipment used.

Since a constant 200 μm step between indentations was adopted, the profile taken in the middle sections poorly represents the FZ, which is at its minimum width in this area (c.a. 0.5 mm), whereas, from the top sections, it can be observed that the microhardness peaks are located on the boundaries of the FZ with the HAZ, and not in the center. This effect might be related to the fast nucleation of fine spherical grains at the interface of the FZ, as discussed in section 3.2.

In the plots below, the fusion zone is highlighted with a grey area whose thickness was measured exactly where the profile was taken. The width of the HAZ (represented with a dashed pattern) could not be measured from the microscope pictures. It was drawn on the plots to include the local minimum of the profiles that are thought to represent the HAZ softening, a typical issue of fusion welding related to the dissolution of the carbides and γ' . According to [46], the softening can also be ascribed to the presence of liquation cracking. The result is that

the HAZ width ranges from 250 to 400 μm , and it is larger in the middle of the hourglass fusion zone.

3.6. Analysis of the cracks

To investigate the origins of the cracks observed in the IN792 welds, the present section presents a focalized examination of some representative examples.

In the E longitudinal section, the transversal cracks are the most evident defect, even visible in the as-weld conditions. The left part of Fig. 9 is dedicated to the crack indicated with number 4 in Fig. 4, characterized by a V-shape with the apex pointing towards the welding direction. The internal surfaces of the crack are magnified in Fig. 9b, showing rough surfaces that might indicate that the rupture occurred at an advanced stage of the solidification process. The presence of a liquid film is likely to be excluded. The IPF map shown in Fig. 9c proves that the crack is intergranular. The virtual superimposition of the IPF map with the SEM picture of the etched weld shows that the crack is extended

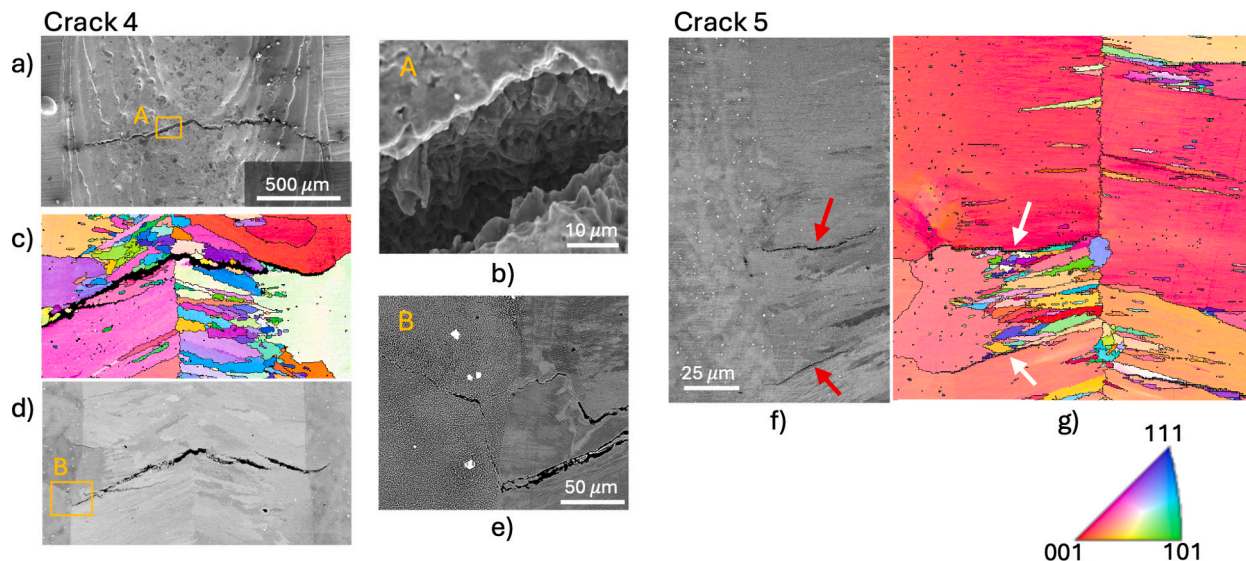


Fig. 9. Investigation on crack 4 and crack 5. Crack 4: a) as-weld picture; b) magnification of detail A; c) IPF map; d) polished and etched section; e) magnification of detail B. Crack 5: f) polished and etched surface; g) IPF map.

to the HAZ grain boundaries, as highlighted in Fig. 9d and e. Many cracks, such as crack 5, do not present the V-shape. Higher magnification of the region around crack 5 (Fig. 9f) reveals the presence of a second crack immediately below. The IPF map (Fig. 9g) shows that these two cracks are associated with the boundaries of the same grain that was partially melted during the welding process. The molten material solidified, nucleating a multitude of smaller elongated grains with random orientation. The relative misorientation of the grains included between the two cracks reaches peaks of 60° , while the rest of the map shows larger grains with a relative misorientation below 30° . The cracks stop at the central grain boundary or earlier, suggesting a propagation direction from the HAZ to the FZ.

Crack 3, shown in Fig. 10, is similar to crack 5. A careful investigation of the portion of crack located in the HAZ (detail A) shows that the edges are dark and present a different texture in comparison to the surrounding material. In the latter, the prints of γ' corroded by the Marble's reagent are visible. The crack edges investigated with the EDS analysis (Fig. 10c) show that the crack is depleted in Al and Ti, except for some particles of carbides in the fracture area. Above the crack, the composition of γ' rich in Al and Ti and depleted in Co and Cr is confirmed by the speckled appearance of the distribution of these elements; a crown-shaped eutectic region, also rich in Al and Ti, is located below the crack.

Fig. 11 is dedicated to two cracks found in the DS weld. Observing simultaneously the SEM pictures and the IPF map, it can be noticed that the crack 6 extension is limited to the FZ, and it does not cross the central grain boundaries. Moreover, the IPF map shows a longitudinal grain boundary indicated with an orange arrow that can be mistakenly taken for the boundary of the FZ. In reality, the SEM picture of the etched sample shows that this grain is completely contained in the FZ, and the boundary is cracked. Analogous thin cracks were often found in the FZ of the etched samples. To further investigate the region, the kernel average misorientation (KAM) map is reported in Fig. 11d. KAM is a measure of the local misorientation of every point in a grain in comparison to the neighbor points. The local misorientation can be associated with the strain and, thus, the local deformation. The figure shows an increase in the KAM value along the boundaries of both crack 6, with a peak on the left side.

Crack 2, depicted on the left side of Fig. 11, was observed in the DS

cross-section and developed in the HAZ, following the interface between HAZ and FZ. This crack is covered by a film of dark material.

A possible explanation for the presence of a dark film of material on the edges of some of the cracks detected in the HAZ is the mechanism called 'backfilling'. The grain boundaries liquated, opening a way for the molten material in the FZ to migrate into the HAZ. To confirm this theory, the following micrographs in Fig. 12 show magnifications of the area between FZ and HAZ in both the E and DS cross-sections. Many thick grain boundaries running from the HAZ to the FZ are visible. Micrograph at higher magnification shows the appearance of these boundaries as a thick film of a material with no γ' , and EDS point analysis revealed the richness of Cr and Co, which confirms that the composition of this region is identical to the one of the FZ. Indeed, it is useful to compare the main elements detected in point 1, in Fig. 12a1, and point 2 in Fig. 12a2, collected via EDS analysis and presented in Table 3. Ojo et al. [47] reported that the thickness of the layer of melted material migrated into grain boundaries is a key factor in preventing further cracking since a thick liquid film preserves the continuity of the material. Nonetheless, the high magnification picture (Fig. 12b2) shows the incipient separation of small particles precipitating on the internal edges, whose composition is difficult to assess. According to the literature, they can be either small γ' particles or eutectic spots [45,48].

3.7. Numerical results

Following the CWM approach, the geometrical parameters of the heat source are calculated by numerical run trials aimed at matching the numerical and experimental weld pool shapes. Some parameters of the laser source, such as P and v, are those used in the welding tests. The geometrical parameters (r_e , r_m , r_b , H_1 , and H_2) and the thermal efficiency (β_1 and β_2) are instead calculated by using both a cross-section of the bead and thermal histories measured with thermocouples at some distance from the welding line. Since thermocouple measurements were not carried out in this work, the results obtained by Barbieri et al. [15] on laser welding of DS IN792 were used for the heat source parameters calibration. A good agreement between experimental and numerical results, obtained by using the parameter values collected in Table 4, can be observed in Fig. 13.

Fig. 14 shows the effect of isotropic (E), and anisotropic (DS) thermal

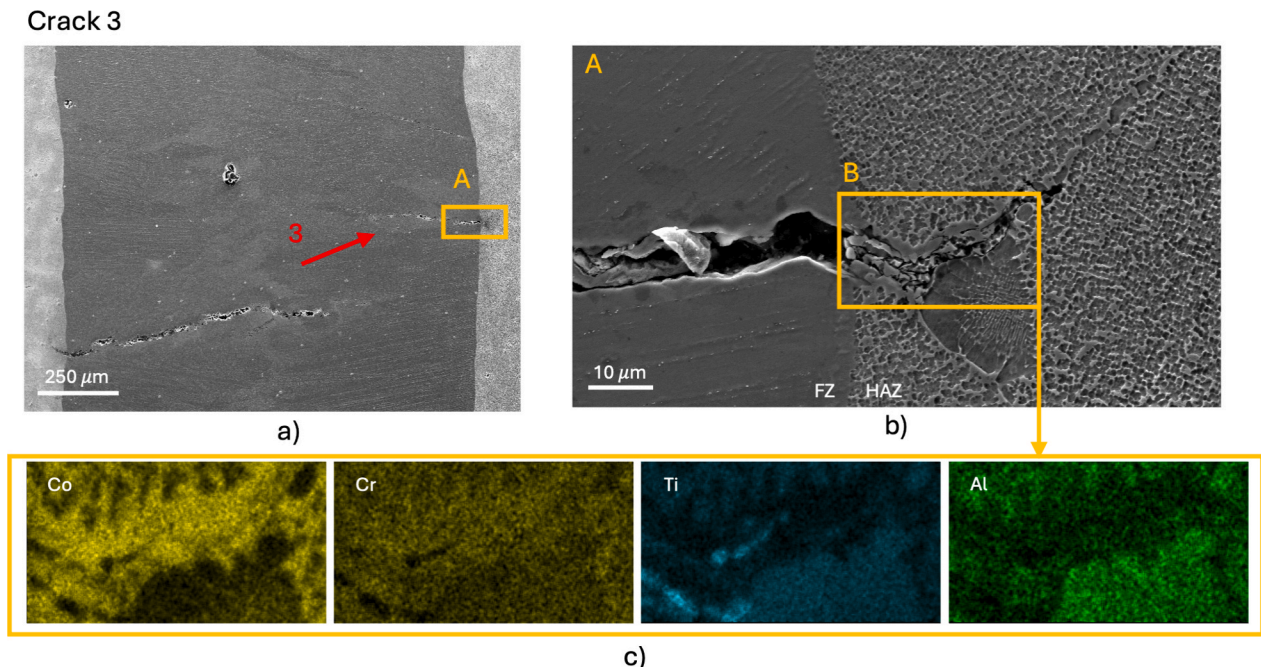


Fig. 10. Investigation on crack 3. a) etched and polished surface; b) magnification of detail A; c) EDS element maps of detail B.

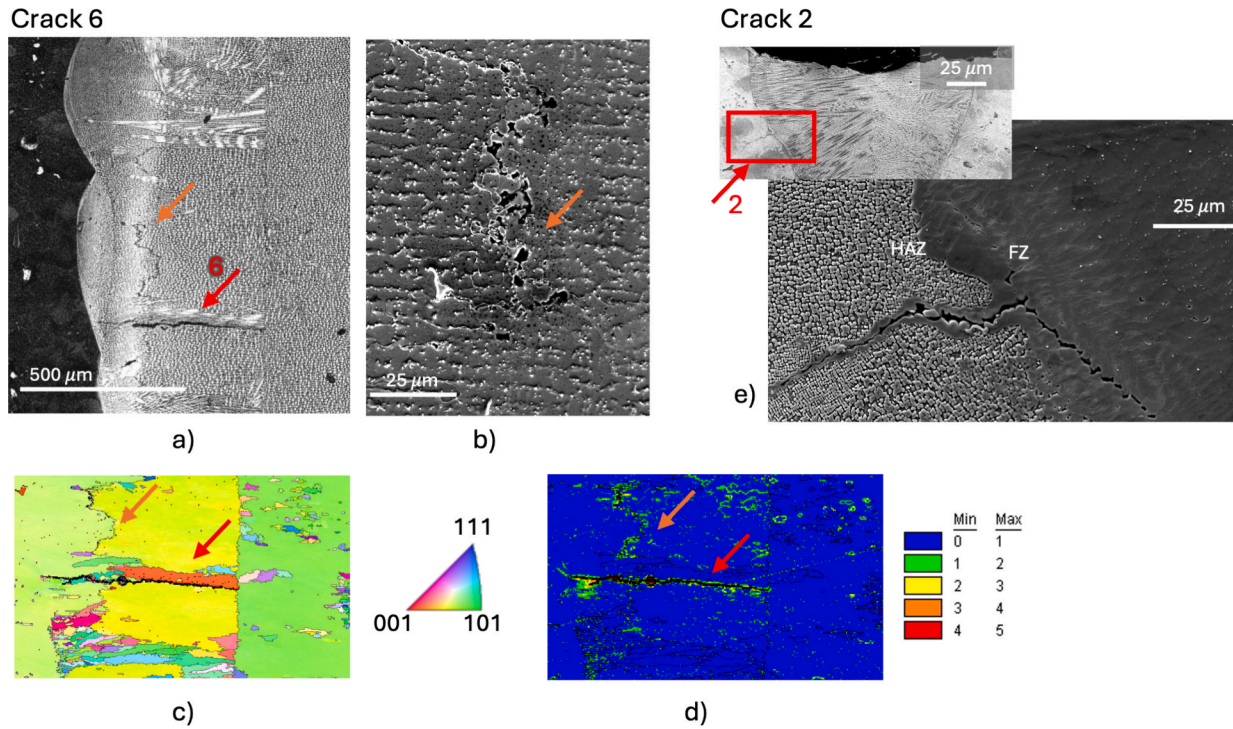


Fig. 11. Investigation on crack 6 and crack 2. Crack 6: a) SEM picture of the etched section and b) magnification of the thin crack indicated with orange arrow; c) IPF map; d) KAM map. Crack 2: e) overview of the cross-section upper parts and magnification of crack 2. The section was in electrochemically-etched conditions.

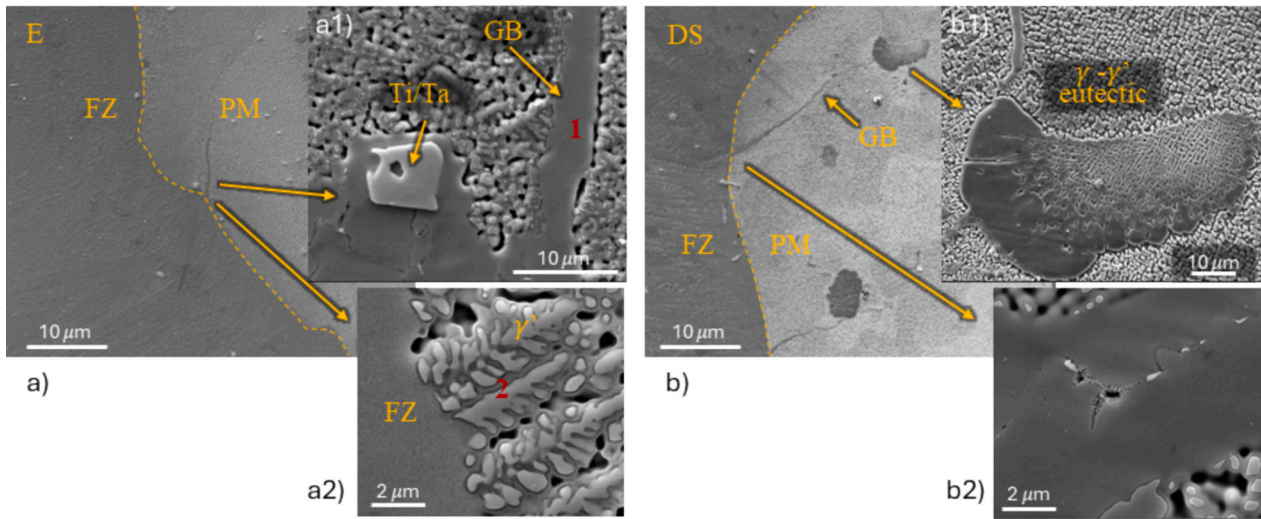


Fig. 12. Backfilling mechanism in the E (a, a1, a2) and DS (b, b1, b2) cross-sections. a) overview of a border region between the FZ and the HAZ in the E weld. A thick grain boundary is emerging from the FZ into the HAZ; a1) magnification of the grain boundary next to the fusion line; a2) dissolution of γ' in the FZ; b) grain boundaries and eutectic regions in the HAZ of the DS cross-section; b1) magnified crown-shaped eutectic; b2) magnification of a grain boundary incipient separation on the interface between FZ and HAZ.

Table 3

Main elements of the chemical composition of points 1 and 2 in Fig. 12.

wt%	Al	Ti	Cr	Co	Ta
pt. 1	3.29	4.21	11.29	8.11	5.43
pt. 2	4.31	6.44	5.70	6.57	5.44

Table 4

Laser welding source parameters.

Source	P (W)	v (mm/s)	r_e (mm)	r_m (mm)	r_l (mm)	$H_{1;2}$	$\beta_{1;2}$
1	1250	25	1.8	0.25	—	0.8	0.4
2	1250	25	—	0.25	0.44	1.2	0.3

conductivity applied to the parent metal on the temperature contours on the top of the welded plates. The arrows highlight a slight difference in the semisolid zone extension, the so-called ‘mushy zone’, where possible solidification cracks occur because of the limited possibility of the liquid

metal to flow and compensate for the shrinkage [49]. The table reporting the thermal conductivity values that were used in the model (reported in the Appendix) shows that along the welding direction, the thermal conductivity of DS alloy is slightly enhanced at high

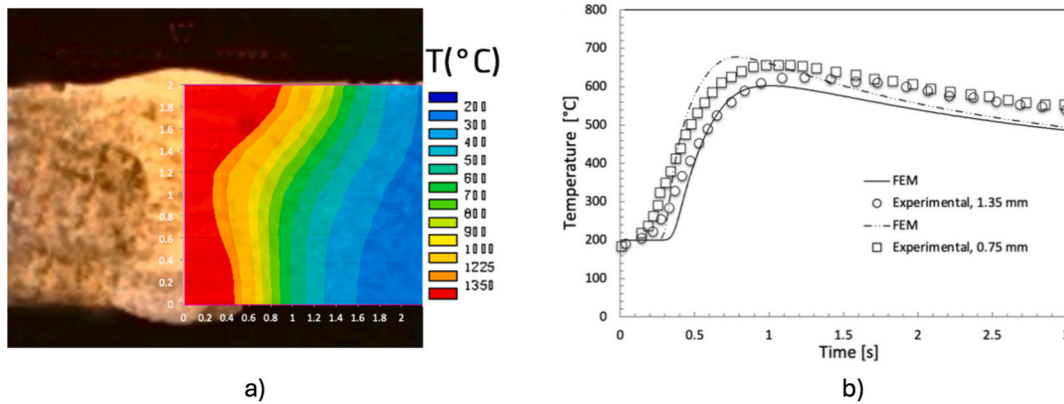


Fig. 13. Matching between experimental and numerical results: a) weld pool shape (red zone), b) temperature history at 0.75 mm and 1.35 mm from the seam border. Experimental results are taken from the paper by Barbieri et al. [15] (DS IN792, P = 1250 W, v = 25 mm/s, preheating temperature = 200 °C, H = 2 mm).

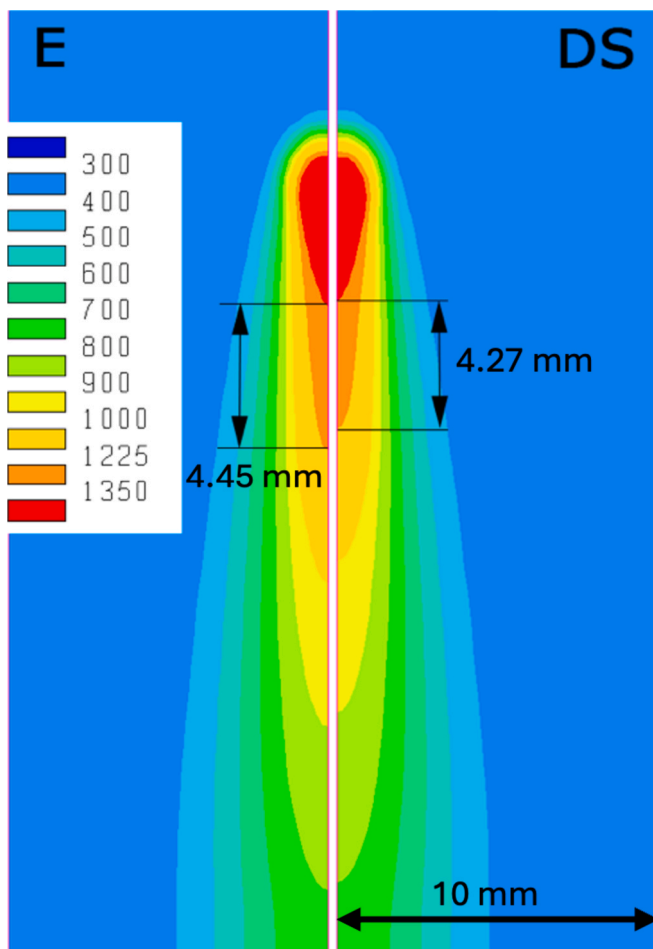


Fig. 14. Temperature contours (°C) on the top of the plate: comparison between E and DS alloys.

temperatures in comparison to the E alloy, which means that the grain boundaries act as thermal barriers. However, this difference is very modest, and it seems insufficient to justify the different hot cracking susceptibility.

Using the same approach described by Hu and Richardson in [8], the plastic strain in the y-direction, believed to be the main cause of transverse crack formation, was studied. Fig. 15a shows a comparison between temperature and plastic strain evolution calculated at a generic node (node A in Fig. 3), using E and DS IN792 properties, respectively,

during and after solidification. It is observed that for both the analyzed alloys, no plastic strain occurs during the very fast solidification, suggesting that hot cracks could not be solidification cracks but instead sub-solidus cracks. Fig. 15b, elaborated from [22], reports the trend of the elongation to rupture of IN792 alloy at different temperatures suggesting that the material presents minimum ductility in a range of temperatures comparable to the one in which the plastic strain assumes a value different from zero in Fig. 15a. Fig. 15c illustrates the plastic strain variation along path A, shown in Fig. 3, at $t = 2.5$ s. The maximum value of the plastic deformation is not reached in the centerline of the weld, suggesting that this is not the most likely location of origin of hot cracks. Moreover, Fig. 15d shows that the longitudinal stress peak is at the FZ/HAZ interface and is much higher in the case of the E IN792 alloy. This highlights the non-negligible influence of material anisotropy on the hot-cracking susceptibility of the alloy investigated.

4. Discussion

Crack susceptibility of Ni-based superalloys depends on many different factors related to stresses and alloy-dependent features, such as grain structure and formation of low-melting components [50,51]. Moreover, the choice of appropriate process parameters is paramount to limit the tendency of an alloy to crack. The same equipment and process parameters have been used to weld the IN792 plates in the E and DS conditions. In addition, Table II shows that there is no difference in terms of chemical composition.

According to several empirical approaches, the weldability of a superalloy can be predicted from the chemical composition. For example, in several studies [24,52]), new enhanced versions of the diagram originally reported in [53] are presented to graphically show that an alloy is prone to cracking when the following relation holds: $2 \text{ Al wt\%} + \text{Ti wt\%} > 6 \text{ wt\%}$, which implicates, according to Table II, that IN792 is not suitable for welding in either of the conditions proposed. It must be noted, though, that the relation was defined for TIG welding. The contribution in [54] investigated the effect of the Ti/Ta ratio on the castability of IN792. The paper is not focused on welding, however, the casting defect called hot tearing is very similar to the mechanism of welding liquation cracking, and it is useful to underline how these two elements were found to have a critical role in the integrity of IN792. The authors suggested avoiding a Ti/Ta ratio between 0.7 and 1. The value calculated from Table II is 0.8, forecasting a predisposition for grain boundary liquation in both microstructures. This is corroborated by the fact that HAZ fissuring was mainly detected in the E welds but crack 2 (see Fig. 11e) proves that the issue affects the DS welds as well. Thus, it can be concluded that IN792 is hardly 'weldable', but the directional solidification is responsible for the mitigation of the issues related to the chemical elements involved and prevents the hot cracking phenomenon

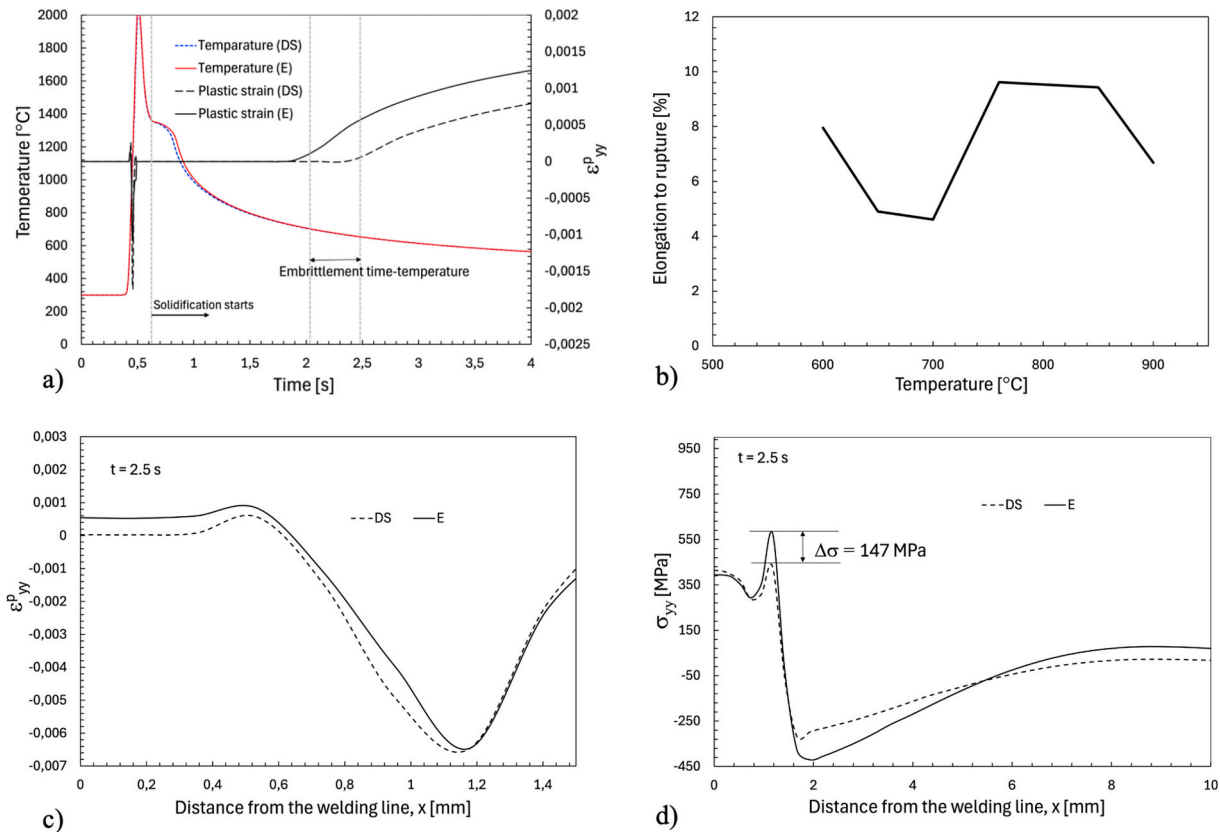


Fig. 15. Thermo-mechanical numerical results: a) temperature and plastic strain evolution during time at node A (Fig. 3); b) elongation at rupture of the alloy in the equiaxed state as a function of temperature [22]; c) plastic strain distribution along path A (Fig. 3) at time 2.5 s; d) longitudinal thermal stress along the path A at time $t = 2.5$ s.

from being a serious issue.

The main advantage related to the process of directional solidification is the removal of a large share of grain boundaries, which can be dangerous, especially in harsh environments with high stresses and temperatures. For these reasons, DS alloys are known to perform better in terms of ductility, creep resistance, and fatigue resistance than conventionally cast components [26–28]. Observing the cracks in the longitudinal section of the E weld, it can be noticed that they are associated with the grain boundaries in the HAZ. The absence of grain boundaries in the parent material of the DS alloy might be the formal reason behind the rare appearance of both liquation and solidification cracks in this alloy. A comparison between Fig. 9f and 11a is useful to highlight this difference: the figures show intergranular cracks in the FZ of DS (crack 6) and E sections (crack 5). None of them cross the centerline of the weld. Instead, crack 5 in the E weld crosses the boundaries of the fusion zone since its left end is in the HAZ, while crack 6, in the DS weld, is limited to the sole fusion zone. According to the IPF maps, it seems clear that crack 6 stops in the fusion zone because there are no other grain boundaries to follow.

The crack propagation path is still unclear. Ke et al. [48] reported that the transverse cracks started from the centerline of the weld, diverting with a certain angle in both directions and creating a sort of V-shape. Instead, according to Zhi et al. [55], cracks extended in both the FZ and the HAZ are thought to start in the HAZ. The description of [48] fits well the aspect of crack 4: Fig. 9 shows that the crack appeared in an area of the bead characterized by many grains nucleated from both sides of the fusion zone that merged in the weld centerline. This location is thus a delicate point that might experience a lack of continuity, thus provoking a crack that followed the grain boundaries into the HAZ as well. It is possible that cracks as cracks 4 and 5 (Fig. 9) followed another mechanism: the centerline of the weld seems more suitable as a crack

stopper than a crack initiation point, indicating that the HAZ fissuring is the origin of the transverse cracks. Probably, more than one mechanism contributed simultaneously. The liquated grain boundaries in the HAZ do not offer any retention for a crack to open. On the contrary, they might create initiation points on the interface between the fusion zone and HAZ.

Other effects might play a role in the fracture mechanisms. For example, David et al. [50] attributed the origin of weld cracking in nickel-base superalloys to the local stress-strain conditions, stray crystal formation (promoted by constitutional supercooling), and the presence of low-melting eutectic films. ‘Stray grains’ are grains with a significant misorientation from the parent material whose formation is associated with dendritic fragmentation and with a low cooling rate that favors the elongation of the dendrites to the point of fragmenting. Although both IN792 E and DS are polycrystalline alloys, a sort of similar effect can be noticed. This is particularly clear from the IPF map in Fig. 11: crack 6 follows a grain marked with orange, which interrupts a wider grain, marked with green. Through the EBSD analysis, the misorientation between these two grains respectively below and above the crack was calculated as 26° . Similarly, crack 5 and the crack below are associated with an area where multiple equiaxed grains while the surrounding is occupied by larger grains. Thus, the presence of the cracks might have been favored by the strong misorientation between the grains involved in comparison to the surrounding region.

The numerical model proposed helps in understanding the deformation occurring in the welds. Fig. 15a shows the almost identical temperature vs time curves in node A (Fig. 3) of the bead for both DS and E models. The start of the solidification process is set after approximately 0.62 s when the temperature reaches a value below 1350 °C (IN792 melting point). Shortly after, at 1.3 s, the spatial distribution of σ_{yy} shows a switch from compression to tension. Hu and Richardson

indicated this moment as the actual start of the solidification process. Interestingly, no plastic strain along the welding direction is detected until 2 s for the E alloy and 2.5 s for the DS alloy. This might suggest that after 2 and 2.5 s, the solidification was over, although it is not possible to ascertain at which exact moment the melting pool completely solidified. The temperatures corresponding to the appearance of plastic strain are between 700 and 650 °C. Du et al. [22] found that in this temperature range, IN792 is characterized by minimum elongation at fracture, as reported in Fig. 15b. This effect is called intermediate temperature embrittlement, which is attributed to grain boundary segregation [56–58]. This suggests that although the plastic strain observed is limited, the alloys are in a critical temperature range, in which their ability to compensate for such a strain is at its minimum. Indeed, Hu and Richardson [8] stated that cracking occurs when the deformation or strain in the weld is above the maximum ductility level that the alloy can sustain. Following these assumptions, the transversal cracks observed are to be considered sub-solidus cracks, since they appeared when the solidification process was concluded. In this regard, it is worth mentioning that David et al. [50], dealing with laser welding of single crystal alloy IN738, during in situ weld crack monitoring investigations, observed that hot cracks formed about 2.3 s from the beginning of the process, while the weld pool solidification was completed after only 1.19 s. Moreover, the observation of the internal surfaces of the cracks, as in Fig. 9b, seems to exclude the presence of a liquid film when the separation of the grain boundaries occurred. Fig. 9b indeed appears quite different from what is reported in [8] regarding solidification cracking of investigated aluminum alloys: the authors observed traces of ‘liquid state ruptures’ in the internal surfaces of their cracks, thereby concluding that the cracks appeared in a semi-solid state.

For a deeper understanding of the phenomenon, stress and strain spatial distribution at 2.5 s are shown in Fig. 15c and d. At this moment a non-zero plastic strain has built up on the centerline of the E weld while it is negligible in the DS alloy. The strain increases and reaches a peak at a distance from the welding centerline of 0.53 mm. The E alloy presents the highest peak. As previously mentioned, this deformation, although modest, is difficult to accommodate considering the temperature range, and it could have induced the opening of the transversal cracks in the fusion zone. The plot of σ_{yy} depicted in Fig. 15d shows that a peak in the stress is stable at c.a. 1.17 mm from the center of the bead, which is a distance that represents approximately the HAZ. According to [8] the peak is a consequence of the thermal inversion that the cross-section of the weld experiences: in the fusion zone and HAZ, the temperature drops quickly, and the material tends to contract. This induces a positive tensile stress since the rest of the material, at a lower temperature, constrains the FZ. Note that the temperature in the parent material increases by conduction. Thus, the peak in the tensile stress might be responsible for the opening of the liquated grain boundaries in the HAZ. Note that there is a difference of about 150 MPa between the stress experienced by the E and the DS alloy.

5. Conclusions

Laser beam welding was performed on IN792 plates with equiaxed and directionally solidified microstructures adopting the same process parameters. The process produced sound welds with the DS plates, while the welds obtained with the E plates presented multiple transversal cracks affecting the fusion zone and the HAZ. The results of several metallurgical analyses and the outcome of a numerical model were used to understand the reasons behind the different behavior of the two

microstructures and the origin of the cracks. The following conclusions are drawn:

- 1- The chemical compositions of IN792 E and DS are identical and according to various empirical correlations found in the literature, not suitable for welding. Moreover, the same process parameters were used to weld the plates, thus no differences shall exist in terms of crack susceptibility if the microstructure is not considered.
- 2- The transversal cracks observed in the FZ of the E weld are hot cracks that occur in a sub-solidus temperature range, as confirmed by the solid structures observed in the internal surfaces of the cracks. The cracks in the FZ probably originated from the HAZ liquated boundaries, which are also cracked.
- 3- The absence of grain boundaries perpendicular to the fusion line in the HAZ is thought to be the formal reason behind the integrity of the DS joint.
- 4- The numerical model proves that the grain boundary action in terms of thermal conductivity is not sufficient to determine a significant difference in the thermal history of the welds.
- 5- The simulations revealed that, in the E weld, a plastic strain builds up at an earlier stage in comparison to the DS. This happens in a temperature range in which IN792 is known to experience a ductility drop. Moreover, the longitudinal stress spatial distribution shows a peak on the interface between FZ and HAZ which is higher in the case of the E welds in comparison to the DS welds.

The grain boundaries' presence and the increased longitudinal stress acting on them are therefore the causes of the cracks detected in the equiaxed joints. Thus, the conclusion that the ability to mitigate the welding crack susceptibility must be added to the list of advantages the directionally solidified alloys offer in comparison to regular polycrystalline equiaxed alloys. Reducing the thermal stresses that act on the welds is a key point in the limitation of hot cracking susceptibility. Several strategies can be adopted, among them increasing the preheating temperature or including a second heat source ahead of the main one.

CRediT authorship contribution statement

Saveria Spiller: Writing – original draft, Visualization, Validation, Investigation, Data curation. **Alessandra Varone:** Writing – review & editing. **Giuliano Angella:** Writing – review & editing. **Francesco Cognini:** Writing – review & editing, Investigation. **Alberto Fabrizi:** Writing – review & editing, Investigation. **Franco Bonollo:** Supervision. **Paolo Ferro:** Writing – review & editing, Visualization, Supervision, Methodology, Formal analysis, Conceptualization.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Appendix 1

The physical and mechanical properties of equiaxed IN792 (isotropic) and directionally solidified IN792 along the grain growth direction are presented in Fig. A1. The properties were used to build the numerical model.

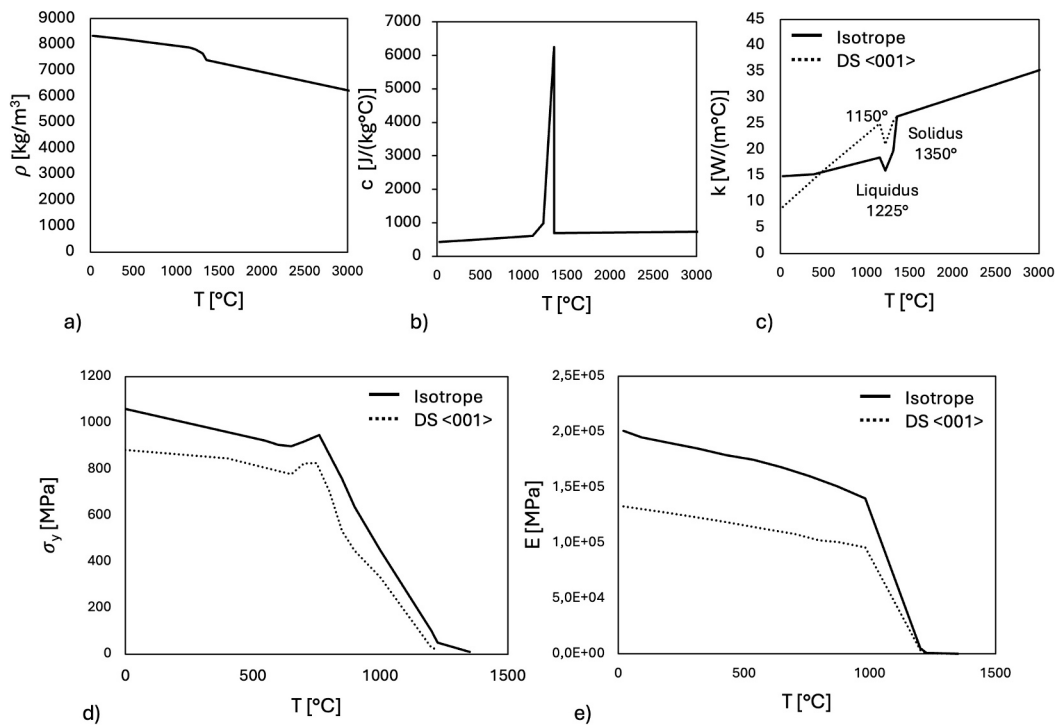


Fig. A1. Physical and mechanical properties of IN792 E and DS (along the grain growth direction).

Fig. A2 shows the plot of the stress distribution on the joints obtained by the numerical model. The stress σ_{yy} (along the welding direction) was recorded along path A (as depicted in Fig. 3), from the weld centerline towards the parent material. The stress was recorded at different moments from t1 to t13.

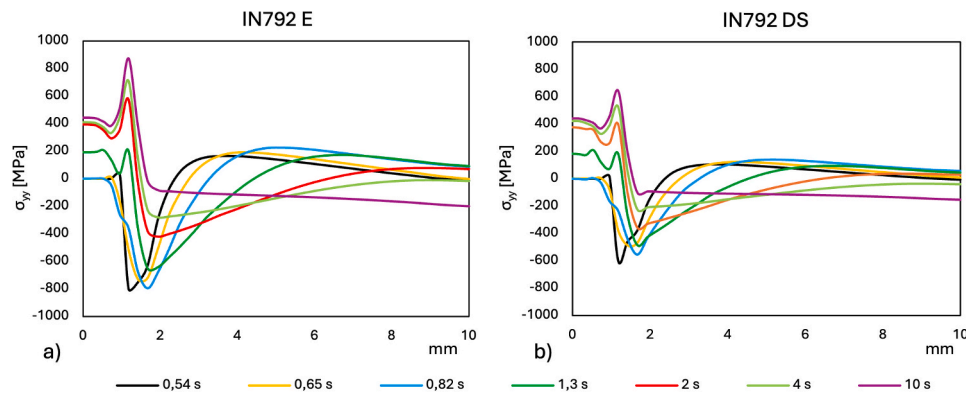


Fig. A2. Stress distribution on path A at different moments for IN792 E and DS obtained with the numerical model.

Data availability

Data will be made available upon reasonable request.

References

- [1] J.M. Wilson, C. Piya, Y.C. Shin, F. Zhao, K. Ramani, Remanufacturing of turbine blades by laser direct deposition with its energy and environmental impact analysis, *J. Clean. Prod.* 80 (Oct. 2014) 170–178, <https://doi.org/10.1016/j.jclepro.2014.05.084>.
- [2] C. Rinaldi et al., "Automatic refurbishment of gas turbine components by CO sub 2 robo-laser," in *Joining and Repair of Gas Turbine Components*, Indianapolis, USA: "Materials Solutions 1997 on joining and repair of gas turbine components, Sep. 1997, pp. 109–117.
- [3] D. W. and F. G. J. and P. A. J. and S. J. and V. R. Gandy, *Development of a laser-based/high strength weld filler process to extend repair limits on IN-738 gas turbine blades*. 4th International EPRI conference welding and repair technology for power plants, 2000.
- [4] R. H. AL-Nafeay, A. O. AL-Roubaiy, and H. Omidvar, "Overview of Joining and Repairing Techniques of Ni-Based Superalloy for Industrial Gas Turbine Applications," *IOP Conf Ser Mater Sci Eng.* vol. 1094, no. 1, p. 012141, Feb. 2021, doi: 10.1088/1757-899x/1094/1/012141.
- [5] S. A. David, J. A. Siefert, J. N. Dupont, and J. P. Shingledecker, "Weldability and weld performance of candidate nickel base superalloys for advanced ultrasupercritical fossil power plants part I: Fundamentals," Jul. 28, 2015, *Maney Publishing*. doi: 10.1179/1362171815Y.0000000035.
- [6] F. Yan, et al., Influence of heat input on HAZ liquation cracking in laser welded GH909 alloy, *Opt. Laser Technol.* 92 (Jul. 2017) 44–51, <https://doi.org/10.1016/j.optlastec.2017.01.016>.
- [7] L.O. Osoba, R.K. Sidhu, O.A. Ojo, On preventing HAZ cracking in laser welded DS Rene 80 superalloy, *Mater. Sci. Technol.* 27 (5) (May 2011) 897–902, <https://doi.org/10.1179/026708309X12560332736593>.
- [8] B. Hu, I.M. Richardson, Mechanism and possible solution for transverse solidification cracking in laser welding of high strength aluminium alloys, *Mater. Sci. Eng. A* 429 (1–2) (Aug. 2006) 287–294, <https://doi.org/10.1016/j.msea.2006.05.040>.

- [9] M. Norouzi, M. Amne Elahi, and P. Plapper, "A review: Suppression of the solidification cracks in the laser welding process by controlling the grain structure and chemical compositions," Jun. 01, 2023, *Elsevier B.V.* doi: 10.1016/j.jaip.2023.100139.
- [10] N. Coniglio, C.E. Cross, Effect of weld travel speed on solidification cracking behavior. Part 1: weld metal characteristics, *Int. J. Adv. Manuf. Technol.* 107 (11–12) (Apr. 2020) 5011–5023, <https://doi.org/10.1007/s00170-020-05231-y>.
- [11] N. Coniglio, C.E. Cross, Effect of weld travel speed on solidification cracking behavior. Part 2: testing conditions and metrics, *Int. J. Adv. Manuf. Technol.* 107 (11–12) (Apr. 2020) 5025–5038, <https://doi.org/10.1007/s00170-020-05232-x>.
- [12] N. Coniglio and C. E. Cross, "Effect of Weld Travel Speed on Solidification Cracking Behavior. Part 3: Modeling." doi: 10.1007/s00170-020-05233-w.
- [13] N. Hashemi, Z.S. Seyedraoufi, Y. Shajari, The effect of Partial solution Heat Treatment Parameters on Microstructure Evaluation of IN792 Superalloy, *Metall. Microstruct. Anal.* 10 (5) (Oct. 2021) 627–633, <https://doi.org/10.1007/s13632-021-00777-z>.
- [14] S. Suharno, A. Sugianto, R. E. Buntario, D. Widagdo, Y. Estriyanto, and B. Harjanto, "Advantage of SWET Technique on Joining Inconel 792 Material," *Type: Double Blind Peer Reviewed International Research Journal Publisher: Global Journals Inc.*, vol. 12, 2012, [Online]. Available: <https://www.researchgate.net/publication/343745711>.
- [15] G. Barbieri et al., "Process Optimization in Laser Welding of IN792 DS Superalloy," *Metals (Basel)*, vol. 14, no. 1, Jan. 2024, doi: 10.3390/met14010124.
- [16] G. Angella, G. Barbieri, R. Donnini, R. Montanari, M. Richetta, A. Varone, Electron beam welding of IN792 DS: Effects of pass speed and PWHT on microstructure and hardness, *Materials* 10 (9) (2017) Sep, <https://doi.org/10.3390/ma10091033>.
- [17] G. Barbieri et al., "Segregation and crack formation in laser welded IN792 equiaxed superalloy," *in press*.
- [18] Y. Li, X. Ye, J. Li, Y. Zhang, Y. Koizumi, A. Chiba, Influence of cobalt addition on microstructure and hot workability of IN713C superalloy, *Mater. Des.* 122 (May 2017) 340–346, <https://doi.org/10.1016/j.matdes.2017.03.014>.
- [19] A. Solomon, G. Aravithan, R. Deepak, M. Kishore Kumar, M. Anantha Padmanaban, Simulation and modeling of the effect of welding process parameters of Inconel 625, *Int J Eng Res Technol* 6 (7) (2018) 1–6.
- [20] M. Zeneli, A. Nikolopoulos, S. Karellas, and N. Nikolopoulos, "Numerical methods for solid-liquid phase-change problems," *Ultra-high temperature thermal energy storage, transfer and conversion*, pp. 165–199, 2021.
- [21] "High-temperature high-strength nickel-base alloys. A practical guide to the use of nickel-containing alloys n°393," Nickel Institute. Accessed: Mar. 04, 2025. [Online]. Available: https://nickelinstitute.org/media/8d93486143182f5/nickel_incopub393_updated-june-2021.pdf.
- [22] B. Du, L. Sheng, Z. Hu, C. Cui, J. Yang, and X. Sun, "Investigation on the microstructure and tensile behavior of a Ni-based IN792 superalloy," *Advances in Mechanical Engineering*, vol. 10, no. 2, Feb. 2018, doi: 10.1177/1687814017752167.
- [23] M.Y. Krasnoperov, R.R.G.M. Pieters, I.M. Richardson, Weld pool geometry during keyhole laser welding of thin steel sheets, *Sci. Technol. Weld. Join.* 9 (6) (Dec. 2004) 501–506, <https://doi.org/10.1179/136217104225021733>.
- [24] N.C. Sekhar, R.C. Reed, Power beam welding of thick section nickel base superalloys, *Sci. Technol. Weld. Join.* 7 (2) (Apr. 2002) 77–87, <https://doi.org/10.1179/136217102225002628>.
- [25] B.S. Koo, P. Thasananaphan, H.F. Nied, Numerical simulation of the formation of hourglass welds during laser welding, *J. Mater. Process. Technol.* 263 (Jan. 2019) 176–185, <https://doi.org/10.1016/j.jmatprotec.2017.12.015>.
- [26] J.H. Lafen, Analysis of an idealized Directionally Solidified FCC Material, *J. Eng. Mater. Technol.* 105 (1983) Oct.
- [27] I.L. Svetlov, N.N. Sukhanov, A.I. Krivko, I.N. Roshchina, I.M. Khatsinskaya, A. I. Samoilov, Temperature-orientation dependence of characteristics of short-term strength, Young's modulus, and linear expansion coefficient of single crystal of ZrSiF alloy, *Strength Mater.* 19 (1) (1987) 58–64.
- [28] A. Xu, M. Saleh, J. Davis, L. Edwards, D. Bhattacharyya, In-situ micro-tensile investigation of strain rate response along <100> and <110> directions in single crystal nickel, *Int. J. Plast* 106 (Jul. 2018) 129–144, <https://doi.org/10.1016/j.ijplas.2018.03.005>.
- [29] T. Khan, P. Caron, Y.G. Nakagawa, Mechanical Behavior and Processing of DS and Single Crystal Superalloys, *JOM* 38 (7) (1986) 16–19.
- [30] A. Abdul-Aziz, "Estimation of the engineering elastic constants of a directionally solidified superalloy for finite element structural analysis," 1991. [Online]. Available: <https://www.researchgate.net/publication/24299678>.
- [31] W. Hermann, H.G. Sockel, J. Han, A. Bertram, "Elastic Properties and Determination of Elastic Constants of Nickel-Base Superalloys by a Free-Free Beam Technique," *the Minerals, Metals and Materials Society* (Aug. 2012) 229–238, https://doi.org/10.7449/1996/superalloys_1996_229_238.
- [32] Q. Yin, Y. Lian, Z. Wen, H. Pei, J. Wang, Z. Yue, Atomic simulation of the effect of orientation on tensile/compressive properties in nickel-based single crystal superalloys, *J. Alloys Compd.* 893 (Feb. 2022), <https://doi.org/10.1016/j.jallcom.2021.162210>.
- [33] D.M. Shah, D.N. Duhl, The effect of orientation, temperature and gamma prime size on the yield strength of a single crystal nickel base superalloy, *Superalloys* (1984) 105–114.
- [34] D.E. Newbury, N.W.M. Ritchie, Is scanning electron microscopy/energy dispersive X-ray spectrometry (SEM/EDS) quantitative? *Scanning* 35 (3) (May 2013) 141–168, <https://doi.org/10.1002/sca.21041>.
- [35] M.M. Cueto-Rodriguez, et al., Numerical and Experimental analyses of the effect of Heat Treatments on the phase Stability of Inconel 792, *Adv. Mater. Sci. Eng.* 2018 (2018), <https://doi.org/10.1155/2018/4535732>.
- [36] O.A. Ojo, N.L. Richards, M.C. Chaturvedi, Contribution of constitutional liquation of gamma prime precipitate to weld HAZ cracking of cast Inconel 738 superalloy, *Scr. Mater.* 50 (5) (Mar. 2004) 641–646, <https://doi.org/10.1016/j.scriptamat.2003.11.025>.
- [37] O.A. Ojo, N.L. Richards, M.C. Chaturvedi, Study of the Fusion Zone and Heat-Affected Zone Microstructures in Tungsten Inert Gas-Welded INCONEL 738LC Superalloy, *Metall. Mater. Trans. A* 37 (2006) 421–433.
- [38] W. Zhang, L. Liu, Solidification microstructure of directionally solidified superalloy under high temperature gradient, *Rare Met.* 31 (6) (Dec. 2012) 541–546, <https://doi.org/10.1007/s12598-012-0554-z>.
- [39] T. Grosdidier, A. Hazotte, A. Simon, Precipitation and dissolution processes in γ/γ' single crystal nickel-based superalloys, *Mater. Sci. Eng. A* 256 (1–2) (1998) 183–196.
- [40] A. Ghasemi, M. Pouranvari, Thermal processing strategies enabling boride dissolution and gamma prime precipitation in dissimilar nickel-based superalloys transient liquid phase bond, *Mater. Des.* 182 (Nov. 2019), <https://doi.org/10.1016/j.matdes.2019.108008>.
- [41] S. J. Yeom, D. Y. Yoon, and M. F. Henry, "The Morphological Changes of γ' Precipitates in a Ni-8Al (Wt Pct) Alloy during Their Coarsening,"
- [42] C. Hays, "Size and shape effects for gamma prime in alloy 738," in *Journal of Materials Engineering and Performance*, Springer Science and Business Media, LLC (2008) 254–259, <https://doi.org/10.1007/s11665-007-9135-y>.
- [43] X. Wang, et al., Effect of withdrawal rate on precipitation characteristics of MC-type carbides in a nickel-based directionally solidified superalloy with high Re content, *Vacuum* 183 (Jan. 2021), <https://doi.org/10.1016/j.vacuum.2020.109800>.
- [44] X.W. Li, L. Wang, J.S. Dong, L.H. Lou, Effect of Solidification Condition and Carbon Content on the Morphology of MC Carbide in Directionally Solidified Nickel-base Superalloys, *J. Mater. Sci. Technol.* 30 (12) (Dec. 2014) 1296–1300, <https://doi.org/10.1016/j.jmst.2014.06.010>.
- [45] K. Han, H. Wang, F. Peng, B. Zhang, L. Shen, Investigation of microstructure and mechanical performance in IN738LC joint by vacuum electron beam welding, *Vacuum* 162 (Apr. 2019) 214–227, <https://doi.org/10.1016/j.vacuum.2018.12.047>.
- [46] S. Li, K. Li, Z. Cai, J. Pan, Behavior of M23C6 phase in Inconel 617B superalloy during welding, *J. Mater. Process. Technol.* 258 (Aug. 2018) 38–46, <https://doi.org/10.1016/j.jmatprotec.2018.03.009>.
- [47] O.A. Ojo, M.C. Chaturvedi, Liquation microfissuring in the weld heat-affected zone of an overaged precipitation-hardened nickel-base superalloy, *Metall. Mater. Trans. A Phys. Metall. Mater. Sci.* 38 (2) (Feb. 2007) 356–369, <https://doi.org/10.1007/s11661-006-9025-1>.
- [48] K. Han, H. Wang, L. Shen, B. Zhang, Analysis of cracks in the electron beam welded joint of K465 nickel-base superalloy, *Vacuum* 157 (Nov. 2018) 21–30, <https://doi.org/10.1016/j.vacuum.2018.08.011>.
- [49] N. Wang, S. Mokadem, M. Rappaz, W. Kurz, Solidification cracking of superalloy single- and bi-crystals, *Acta Mater.* 52 (11) (Jun. 2004) 3173–3182, <https://doi.org/10.1016/j.actamat.2004.03.047>.
- [50] S.A. David, S.S. Babu, J.M. Vitek, Weldability and microstructure development on Nickel-base superalloys. *Conference on Numerical Analysis of Weldability*, 1997.
- [51] C. Fink, D. Keil, M. Zinke, Evaluation of hot cracking susceptibility of Nickel-based alloys by the PVR test, *Weld. World* 56 (2012) 37–43.
- [52] L. Haußmann, et al., Crack-Free Welding of a Co-Base Superalloy with High γ' Precipitate Fraction, *Adv. Eng. Mater.* 24 (12) (2022) Dec, <https://doi.org/10.1002/adem.202200609>.
- [53] M. Prager, S. Shira, "welding Resourcing Council Bulletin" (1968).
- [54] J. Zhang, Effect of Ti and Ta on hot cracking susceptibility of directionally solidified Ni-based superalloy IN792, *Scr. Mater.* 48 (6) (2003) 677–681, [https://doi.org/10.1016/S1359-6462\(02\)00559-6](https://doi.org/10.1016/S1359-6462(02)00559-6).
- [55] Z. Tian, et al., Dynamic laser welding hot cracking behavior and mechanism of new structural material Ni–28W–6Cr alloy for molten salt reactor, *J. Mater. Res. Technol.* 25 (Jul. 2023) 3022–3036, <https://doi.org/10.1016/j.jmrt.2023.06.170>.
- [56] L. and S. G. and M. Y. and C. R. and S. R. Zheng, "Mechanism of intermediate temperature embrittlement of Ni and Ni-based superalloys," *Critical Reviews in Solid State and Materials Sciences*, vol. 37, no. 3, pp. 181–214, 2012, doi: 10.1080/10408436.2011.613492.
- [57] L. Zheng, R. Chellali, R. Schlesiger, D. Baither, G. Schmitz, Intermediate temperature embrittlement in high-purity Ni and binary Ni(Bi) alloy, *Scr. Mater.* 65 (5) (Sep. 2011) 428–431, <https://doi.org/10.1016/j.scriptamat.2011.05.024>.
- [58] W. Kai, X. Tingdong, W. Yanqing, D. JinHui, Intermediate-temperature embrittlement induced by non-equilibrium grain-boundary segregation of sulfur in Ni–Cr–Fe alloy, *Philos. Mag. Lett.* 89 (11) (2009) 725–733, <https://doi.org/10.1080/09500830903277180>.