



Effect of Filler-Metal Nb Content on Laves-Phase Formation and Solidification Cracking in GTA-Welded Alloy 718

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Abstract

Niobium causes precipitation strengthening of Alloy 718 but also causes segregation during weld solidification. This research investigates the relationship between the filler-wire Nb content of 4.75–5.50 wt.% with interdendritic second-phase formation and cracking in gas tungsten arc welds. Thermo-Calc Scheil-Gulliver calculations, optical microscopy, image analysis, and Vareststraint-type testing were used to compare five experimental filler wires using fixed nominal welding parameters. Increasing the calculated operational solidification interval, which is defined between liquidus and 99% solid, grows from 106 to 119 °C with the incorporation of more filler-wire Nb. The expected Laves fraction was predicted to change from 2.1 to 4.8 vol.%, while the measured interdendritic second-phase fraction increased from 2.5 ± 0.3 to 5.3 ± 0.6 area%. At 2% nominal augmented strain, the dimensionless cracking index typing went up from 0.05 ± 0.02 to 0.32 ± 0.06 . The five composition means regressions showed that cracking was strongly associated with measured or predicted phase fractions. These links are consistent with cracking assisted by segregation but do not by themselves prove the attribution. The outcomes uncovered the lower-Nb fillers as competent candidates for follow-up weldability assessments. Before a qualified filler-composition recommendation can be defined, direct dilution measurements, verified phase identification and Post weld mechanical testing.

Keywords: Alloy 718; filler-metal niobium; Laves phase; solidification cracking; GTAW; Vareststraint

1. Introduction

Alloy 718 is a nickel alloy that is strengthened by precipitation. It is used where structural components require retention of strength at elevated temperature. The strengthening response depends on its composition and thermal history, including Nb availability for γ'' precipitation. The published Alloy 718 composition interval of 4.75–5.50 wt.% refers to Nb plus Ta, rather than Nb alone [1]. According to ultrasonic testing techniques applied to PBF 718, melting initiation occurs in the vicinity of 80J/mm³. Consequently, it cannot be assumed that the final strength or ductility will be preserved by improving weldability by adjusting Nb without mechanical testing.

This trade-off is given a metallurgical base by the solidification sequence. During primary γ solidification solute redistribution can produce Nb-enriched interdendritic liquid and terminal constituents containing Laves phase. Knorovsky et al. [3] defines the solidification diagram of Alloy 718 which serves the basis for interpreting these reactions. Microscopic and calorimetric investigations of fusion zones of Alloy 718 further studied the development of Laves phase [4]. In present work, the term Laves-rich constituent refers to the interpretation of the interdendritic morphology and chemical enrichment. It does not mean that every segmented image pixel can be attributed to a crystal-structurally confirmed Laves phase.



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The composition and temperature influence the amount and distribution of interdendritic constituents. The work of Janaki Ram et al. [5] shows that pulsing the current can enhance the GTA weld microstructure. It can also lower the presence of the Laves phase and enhance the heat treatment response later on. Anbarasan et al. [6] investigated the influence of hydrogen and molybdenum on the formation of Laves phase and tensile properties of Alloy 718 GTA welds. These studies indicate that an evaluation of filler Nb should differentiate the compositional variable from variations in shielding conditions, heat input and post-weld treatment. They also warn against treating a trend of phase-fractions as the entire description of mechanical performance.

Cracking during solidification is caused by liquid distribution, feeding and the deformation of the partially solidified structure. Draxler and others created a pore-based criterion for cracking [7], a model for liquid pressure at grain boundaries [8], and simulations of Vareststraint tests for Alloy 718 [9]. The conditions under which strain localizes and liquid permeability are connected by their framework. A Scheil calculation yields composition-dependent information on solidification but does not calculate local tensile strain, liquid pressure, or even crack initiation event. It is thus necessary to perform experiments mechanisms of cracking and thermodynamic predictions.

Recent studies of arc and laser deposition expand this context. According to Artaza et al. [10] plasma-arc deposition of Alloy 718 with a focus on the effect of heat input and cooling strategies while Kindermann et al. [11] study on wire-arc deposition with a focus on heat input and build interruptions. Research has been done on quasi-continuous-wave laser processing to control Laves phase [12]. In the works of Sreekanth et al. [13], the effect of laser deposition parameters and thermal treatments on Nb-rich phases was examined. Further, it was shown that the location in a single-track deposit does impact the measured secondary-phase fraction [14]. In comparing compositions, it is important to carefully regulate the sampling region and thermal boundary conditions.

The used geometry of weld and initial state of materials also affects cracking susceptibility. Alizadeh-Sh et al. investigated the link between laser-cladding conditions and cracking based on empirical-statistical analysis [15] and the clad geometry and cracking association [16]. García-Sesma et al. explored the effects of chemistry, weld shape, and pre-weld treatments in laser welds of cast Alloy 718 [17]. Jeong et al. [18] studied the initial microstructure of Alloy 718 forgings on weld hot cracking. In a recent study, Pandey and colleagues used several different modelling approaches along with experiments to look at how the interlayer dwell time affects segregation and cracking in the deposited Alloy 718 [19]. The geometries and thermal histories of these results differ from single-pass GTA welds, providing mechanistic context.

Studies of hardness and heat-treatment deal with a complementary part of the problem. According to Orozco-Caballero et al, the variation in the local Nb concentration and the resultant nanoindentation response in cast IN718 were mapped [21]. The examination of the structure and hardness of alloy 718 TIG butt joints was carried out by Górka et al [20]. Further studies deal with solidification and homogenization [22], laves dissolution in the laser deposited material [23], and dissolution kinetics during the homogenization [24]. Research relating to the variations in heat-treatment [25] as well as the process–structure–property relationships [26] show the significance of evaluating the final treated material. The hardness or tensile measurements on the actual welds cannot be replaced by the results from these various processes.

The question being addressed here is whether the reported microstructure and cracking response varies systematically with five closely spaced filler-wire Nb levels with fixed nominal GTAW settings. Comparison of calculated solidification descriptors, measured interdendritic second-phase area fractions and a normalized surface-crack-length index. A composition-resolved comparison in this test configuration is its contribution. Because dilution is estimated and the mass balance reporting composition table is incomplete, the comparison is taken to be with the filler-wire Nb; stated is not the complete isolation of Nb and general qualification of a filler specification.

2. Experimental Procedure

2.1. Base material and filler compositions

The base material consisted of wrought Alloy 718 plates, 8 mm thick, supplied in a solution-treated and aged condition. The reported characterization used optical emission spectrometry (OES) at three locations per plate. Five experimental filler wires, F1–F5, were manufactured by vacuum induction melting followed by wire drawing to a diameter of 1.2 mm. Wire compositions were reported from OES and combustion analysis for carbon and sulfur. Table 1 reproduces the composition values supplied for this study; these values require reconciliation with the complete chemical analyses because their totals do not close to 100 wt.%.

Table 1. Reported plate and filler-wire compositions (wt.%); analytical mass balance requires verification.

Element	Base 718	F1	F2	F3	F4	F5
Ni	52.0	52.0	52.0	52.0	52.0	52.0
Cr	18.0	18.0	18.0	18.0	18.0	18.0
Fe	18.5	18.5	18.5	18.5	18.5	18.5
Nb	5.05	4.75	4.90	5.05	5.20	5.50
Mo	3.00	3.00	3.00	3.00	3.00	3.00
Ti	1.00	1.00	1.00	1.00	1.00	1.00

Element	Base 718	F1	F2	F3	F4	F5
Al	0.55	0.55	0.55	0.55	0.55	0.55
Co	0.10	0.10	0.10	0.10	0.10	0.10
Mn	0.05	0.05	0.05	0.05	0.05	0.05
Si	0.05	0.05	0.05	0.05	0.05	0.05
C	0.035	0.030	0.030	0.030	0.030	0.030
Others*	<0.1	<0.1	<0.1	<0.1	<0.1	<0.1
Listed total†	98.335	98.030	98.180	98.330	98.480	98.780

* Reported aggregate trace-element allowance; individual Ta, P, S, B and other trace concentrations were not supplied.

† Arithmetic sum excluding “Others,” not a normalized composition. The unaccounted balance exceeds the stated allowance. Base 718 is the plate; F1–F5 are experimental wires.

Base 718 identifies the plate substrate, F3 identifies a filler wire. Despite having a common Nb value of 5.05 wt.%, the reported carbon contents of 0.035 and 0.030 wt.% and differing processing histories indicate that they are not duplicate materials. In the cited F1–F5 series, carbon and other enumerated elements are constant with the exception of Nb. The difference in composition between the plate and wire carbon is a common offset unless dilution is similar between welds. It is critical that the effects of carbon cannot be ignored due to the potential impact on carbide formation and Nb partitioning. Ill-defined analyses and the measured dilution will not suffice. The base-metal dilution was assumed to be $50 \pm 5\%$ in the modelling description; no direct measurement was available. When D is the base-metal mass fraction, the composition estimate corresponds to Eq. (1), with C representing the concentration of a given element in wt.%. It is a mixing estimate instead of a weld-metal chemistry measurement.

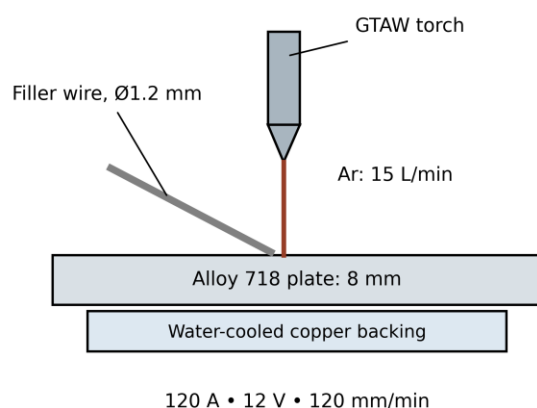
$$C_{weld} = D C_{base} + (1 - D) C_{wire} \quad (1)$$

At D Which Equals Point 50. The weld-metal Nb contents of F1–F5 is estimated at 4.900, 4.975, 5.050, 5.125 and 5.275 wt.%. Moreover, they estimated a constant carbon content of 0.0325 wt.% for all five conditions. The calculated digits indicate arithmetical precision but not analytical precision. The horizontal-axis variable experiments in this study use filler-wire Nb. It is important to verify all the normalized calculation inputs before interpreting the absolute thermodynamic predictions.

2.2. Welding procedure

The welding setup and augmented-strain bending idea are introduced in Figure 1. The dimensions of bead-on-plate specimens were $150 \times 60 \times 8$ mm. Prior to welding, their surfaces were degreased with acetone and the plates were clamped on a water-cooled copper backing bar. The welding utilized mechanized programmable platform with DCEN polarity. The diameter of the electrode was 2.4 mm with a nominal tip angle of 60° and an electrode–work distance of 2.0 mm. The torch was supplied with high-purity argon, and no backing gas was used. To minimize crater influences, the initiation and termination of the arc took place outside the assessed welding length.

(a) Welding configuration



(b) Augmented-strain concept

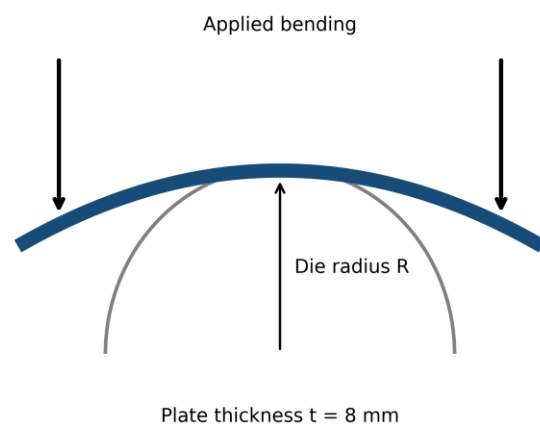


Fig. 1. Conceptual schematic of (a) the reported GTAW arrangement and (b) augmented-strain bending. Not to scale; not a photograph or a verified drawing of the test orientation.

The flat welding position proved useful to join the plates as it allowed to create single-pass weld along the plate centerline. The nominal process parameters and corresponding energy values are given in Table 2. Based on travel speed S in mm/s, the arc energy per unit length is given by Eq. (2). The net heat input will be estimated by assuming that $\eta = 0.70$. Since the efficiency was not determined calorimetrically, the net value is more an estimate than a measured thermal history.

$$E = \frac{VI}{S} = 720 \text{ J/mm} \quad (2)$$

$$Q = \eta E = 504 \text{ J/mm} \quad (3)$$

Table 2. Nominal GTAW settings and calculated energy input.

Parameter	Value
Process / polarity	GTAW / DCEN
Current, I	120 A
Voltage, V	12 V
Travel speed, S	120 mm/min = 2.0 mm/s
Electrode-work distance	2.0 mm
Electrode	W-2%ThO ₂ ; Ø2.4 mm; 60° tip
Shielding gas	Ar 99.999%; 15 L/min
Plate thickness	8 mm
Arc energy, E (Eq. 2)	720 J/mm
Assumed efficiency, η	0.70
Estimated net input, Q (Eq. 3)	504 J/mm = 0.504 kJ/mm

The current, voltage and travel speed are fixed and use a nominal energy input. They don't create the same cooling rates, weld dilution or thermal gradients. The backing configuration was maintained generic to the reported process, with no thermocouple or infrared thermal-cycle measurements available for quantitative comparison. Considering this restriction allows us to explore the solidification interval and cracking mechanism.

2.3. Thermodynamic modelling

The solidification predictions reported were obtained using Thermo-Calc and the nickel-alloy thermodynamic database with an estimated weld as stated in section 2.1. The stated calculation domain was from 1700 °C to 800 °C at a pressure of 1 bar. Scheil–Gulliver calculations assume that the liquids are perfectly mixed, and there is no further diffusion in the solid that was formed earlier. The assumptions provide an idealized segregation path, but do not resolve weld-pool fluid flow, finite-rate solute trapping or the transient temperature field. The Scheil outputs are only those that are used in the present comparison.

The temperature T_L which is the liquidus indicates when primary γ will start to solidify. The term T_{99} refers to a working endpoint that is about 99% solid and 1% liquid. The Eq. is used to find the operational solidification interval ΔT . (4). T_{99} does not represent the equilibrium solidus, and ΔT is not a directly measured brittle-temperature range.

$$\Delta T = T_L - T_{99} \quad (4)$$

The outputs of the retained model are T_L , T_{99} , ΔT , and the predicted Laves volume fraction. According to the calculation records, software/database versions, phase selections, composition normalization and temperature-step or interpolation details were not available. Accordingly, the reported retained model value gives a comparative prediction that needs site calculations projects and export verifications.

2.4. Vareststraint-type cracking tests

The specification of cracking specimens was 150 x 50 x 8 mm and a longitudinal GTAW pass was given along the center line. The procedure reported makes use of bending with the weld in its hot weak condition and cylindrical dies imposing nominally augmented surface strains of 1%, 2%, and 3%. The description provided does not pinpoint when the bending will take place in relation to arc travel or the direction of principal strain in relation to the weld. The test is described conservatively as Vareststraint-type; confirmation of its precise configuration is required from the apparatus record.

For a small t/R , the nominal outer surface strain may be expressed by Eq. R is the die radius and t is specimen thickness. The ε in the above equation has a numerical value which is a fraction; percentage strain = 100ε . We cannot just assume anything from nominal strain labels but rather check against the calibration record the actual radii, radius convention and finite-thickness or compliance correction.

$$\varepsilon \approx \frac{t}{2R} \quad (5)$$

Based on the original description of the experiment, there were a minimum of three specimens tested for each composition at each strain level. Until now, only the explicitly tabulated 2% strain data has been retained as quantitative results, which are reported as mean and standard deviation of each composition from three tests. The tables show numerical results at both the 1% and the 3% strain but did not use these values either to define a threshold strain or a quantitative strain-dependence relationship.

After testing, the weld surfaces were lightly ground and etched to improve crack visibility. Surface crack lengths were measured using calibrated images acquired by stereomicroscopy at 10–20 $\times$ magnification. The cracking index in Eq. (6) is the sum of individual surface crack lengths divided by an evaluated weld length L_{eval} of 30 mm. Both numerator and denominator use millimetres, making CI dimensionless. This index expresses accumulated visible crack length; it does not measure crack depth, fracture energy, or service life.

$$CI = \frac{\sum L_{crack}}{L_{eval}} \quad (6)$$

2.5. Metallography and quantitative image analysis

A low-speed abrasive saw was used to obtain transverse sections around weld mid-lengths. As required, sections were mounted in conductive phenolic resin, ground with 240–1200 grit papers, and polished with the indicated 3 μ m diamond suspension. Macroetching employed a liquid HCl-HNO₃ reagent. Microetching employed a CuCl₂-ethanol-HCl agent referred to as Kalling's reagent that was said to be 5–10 s long, followed by rinsing with ethanol and drying. The proportions of the reagents as well as the final polishing sequence need to be verified in the laboratory record.

Macrostructure was examined at approximately 5–10 $\times$, and optical micrographs were acquired at nominal magnifications of 50 $\times$, 100 $\times$, 200 $\times$, and 500 $\times$. Quantitative sampling used ten fields at 200 $\times$ within the central fusion zone for each composition, away from the surface and fusion boundary. This location was selected to compare similar microstructural regions, rather than to estimate the phase content of the entire weld cross-section. The ten fields are spatial measurements; they are not assumed to represent ten independently produced welds.

According to the original method, SEM in the backscattered-electron mode and the EDS at 15–20 kV and a working distance of approximately 10 mm. The provided records do not contain the related original BSE images, spectra, maps, and quantitative compositions. The reported analyses are therefore not presented here as evidence that may be independently inspected for phase identification. The segmented constituent's crystal structure cannot be established by optical morphology alone, NbC, and other regions which are etched could contribute too. Therefore, the term used for the measured quantity is referred to as the interdendritic second-phase area fraction, while Laves is referred to as the calculated phase or the literature-based interpretation.

Image processing comprised conversion to 8-bit grayscale, rolling-ball background subtraction with a reported radius of 50–100 pixels, Otsu thresholding with manual adjustment when needed, and morphological opening and closing. The original method description states that particles smaller than approximately 1 μ m² should not be included. This is an area threshold, not a busting 1 μ m particle-diameter cutoff. According to equation (6), measured fraction is defined. Image calibration, manual adjustments, and software version must be kept with the source images for reproducibility.

$$A_f = \frac{A_{interdendritic}}{A_{total}} \times 100\% \quad (7)$$

The aggregates composition mean and standard deviation were calculated across ten fields. The standard deviation indicates scatter between fields and does not quantify uncertainty between independently made welds. In the presence of appropriate stereological sampling assumptions, volume fractions can be estimated by the representative area fractions; however, a two-dimensional optical fraction containing multiple constituents does not equate to a calculated Laves-only volume fraction. Consequently, the two metrics are reported individually as area% and vol.%.

2.6. Numerical comparison

The five tabulated composition means were used to recalculate the intercept ordinary least-squares regressions. Each composition was given equal weight. The measured area fraction is related to the predicted Laves fraction, and CI to filler Nb, measured area fraction, or predicted Laves fraction. The five means there is agreement with can be described. The unavailability of the individual measurements led to the conclusion that no hypothesis tests, confidence intervals or claims of statistical independence were obtained. Regression lines appear exclusively over the observed ranges of predictor.

3. Results and Discussion

3.1. Weld geometry and surface observations

Table 3 provides a summary of the bead dimensions and surface observation findings. Figure 2 illustrates source-marked cross-sections for F1, F2, F3 and F5. Both panels with the ambiguous designation F4 are omitted from the figure; F4 continues to appear in all numerical tables and in all regressions. The omission refers to image identification, not to exclusion from the composition group. The reproduced scale labels are not independently calibrated, and the panels are to be used only for qualitative illustration (and not to derive or verify Table 3; no unwelded-base image or documented autogenous reference weld was provided) Section 2.1 describes the substrate and F3 wire as non-interchangeable experimental controls. The mean width varies from 6.1–6.3mm and the penetration depth that is average varies between 2.7–2.9mm. That gives a ratio of .44–.46 of the penetration to the width. The shapes of the beads at the given nominal settings are broadly similar. The small numerical spread cannot be interpreted as an indication of statistically equivalent geometry, as dimensional measurement numbers as well as the meaning of their confirmed uncertainty reported.

Table 3. Reported bead-on-plate geometry and surface crack observations.

Filler	Wire Nb (wt.%)	Width W (mm)	Depth P (mm)	P/W	Surface cracks
F1	4.75	6.2 ± 0.1	2.7 ± 0.1	0.44	None observed
F2	4.90	6.1 ± 0.1	2.8 ± 0.1	0.46	None observed
F3	5.05	6.2 ± 0.1	2.8 ± 0.1	0.45	None observed
F4	5.20	6.3 ± 0.1	2.9 ± 0.1	0.46	Rare; <0.3 mm
F5	5.50	6.3 ± 0.1	2.9 ± 0.1	0.46	Occasional; 0.5–1.0 mm

The ± 0.1 mm values are retained as reported. Their statistical definition and number of dimensional measurements were not specified and require confirmation. These observations are separate from the Vareststraint-type results in Table 6.

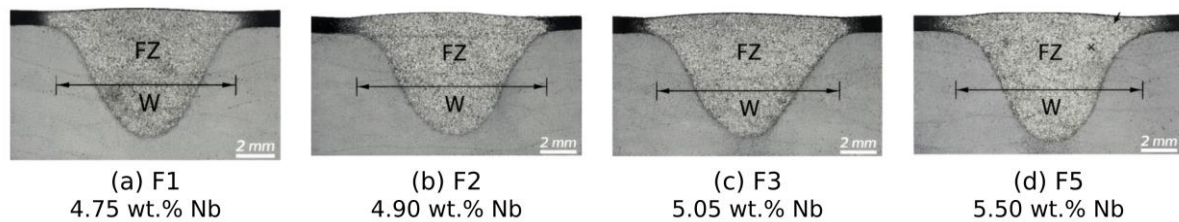


Fig. 2. Cross-sections of welds with source label for F1, F2, F3 and F5. The F4 source panels, both designated with an ambiguous label, have not been included; however, the numerical results of F4 remained in Tables 3–6. Source scale labels (2 mm) are reproduced with no independent calibration. The dimensions presented in Table 3 are neither derived from nor verified by the morphology panels, which lack an independent base-material reference.

F1–F3 bead-on-plate specimens did not show surface cracks. Reported on F4 were rare cracks with lengths shorter than 0.3 mm while F5 showed occasional cracks of about 0.5–1.0 mm. An inspection was performed, not proof that no microscopic or subsurface defects were present. The bead-on-plate specimens were used for these observations, which must be distinguished from the externally strained specimens in Section 3.5. More precisely, invisible cracks without any enforced bending are consistent with the presence of a non-zero CI during a Vareststraint-type test.

3.2. Calculated solidification behavior

Table 4 gives the reported model outputs retained for comparison: T_L , T_{99} , ΔT , and the predicted Laves fraction. Figure 3(a) plots the operational solidification interval and predicted Laves fraction against filler-wire Nb. From F1 to F5, T_L decreases by 5 °C, whereas T_{99} decreases by 18 °C. Thus, the 13 °C increase in ΔT is associated primarily with depression of the operational terminal-solidification temperature. This distinction is more informative than describing the interval increase alone.

Table 4. Reported Scheil predictions retained for the five composition conditions.

Filler	Wire Nb (wt.%)	T_L (°C)	T_{99} (°C)	ΔT (°C)	Laves (vol.%)
F1	4.75	1338	1232	106	2.1
F2	4.90	1337	1229	108	2.6
F3	5.05	1336	1226	110	3.1

Filler	Wire Nb (wt.%)	T _L (°C)	T ₉₉ (°C)	ΔT (°C)	Laves (vol.%)
F4	5.20	1335	1221	114	3.9
F5	5.50	1333	1214	119	4.8

T₉₉ is the operational endpoint at approximately 99% solid. The calculation settings and verification limitations are described in Section 2.3.

The predicted Laves fraction increases from 2.1 to 4.8 vol.%. The trends are consistent with a greater amount of Nb-enriched terminal constituent at higher filler Nb. They do not establish a longer liquid-film lifetime in seconds: converting a temperature interval to a residence time would require the local cooling history. Similarly, the final calculated phase fraction does not measure liquid-film thickness or connectivity during the critical stage of deformation.

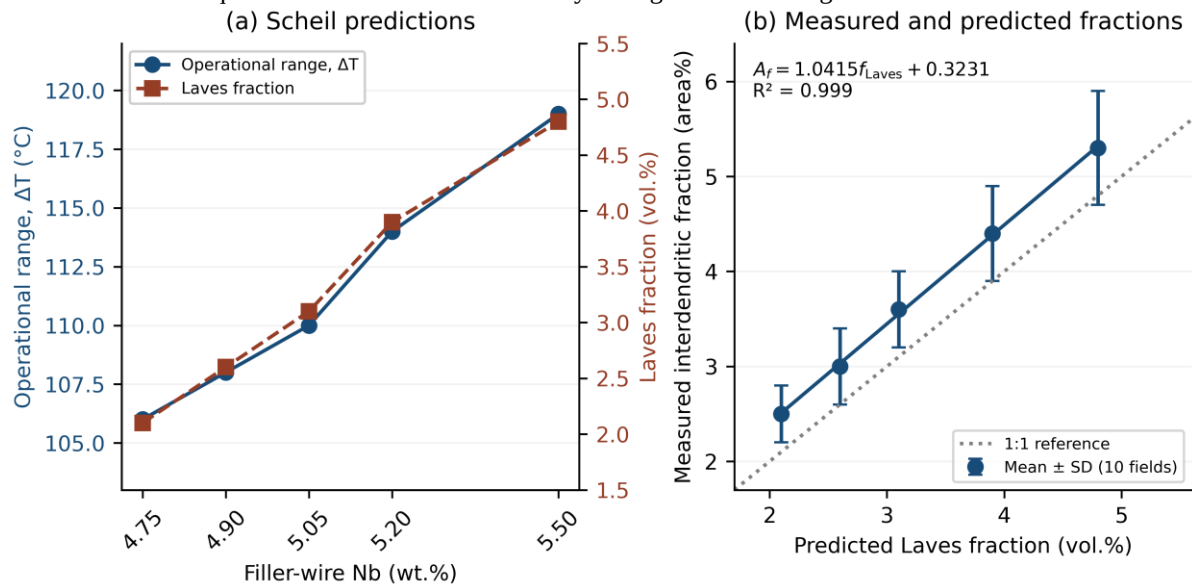


Fig. 3. (a) Reported Scheil operational solidification interval and Laves fraction versus filler-wire Nb. (b) Measured interdendritic second-phase area fraction versus predicted Laves fraction, with Eq. (8) and a 1:1 reference. Error bars are reported ± 1 SD across ten image fields; calculated phase values have no quantified uncertainty bars.

3.3. Microstructural interpretation

The original observations characterize columnar dendritic fusion-zone structures which show relatively isolated interdendritic regions present in F1 which shows an increasing apparent connectivity towards F4 and F5. The panels that are supplied are labeled in the caption of their source as etched optical micrographs. The reported evolution is qualitatively consistent with enhanced segregation, but quantitative connectivity, dendrite-arm-spacing or three-dimensional network measurements were not provided. As such, “more connected” is a morphological interpretation whose veracity requires checking against the original micrographs rather than as a separate measured network parameter.

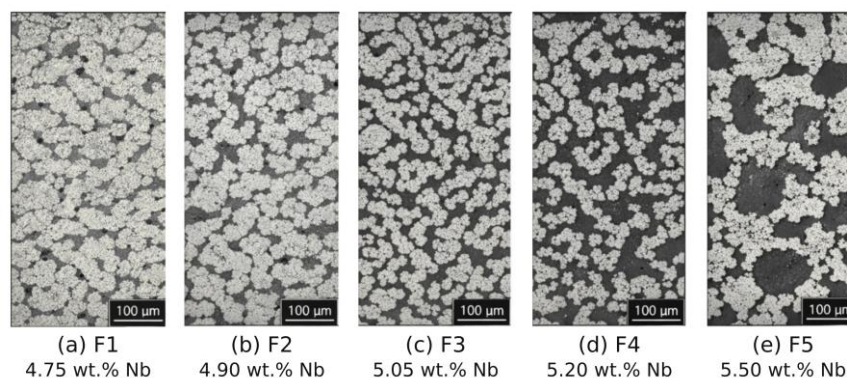


Fig. 4. Source images described as etched optical micrographs of the F1–F5 fusion zones: (a) F1, (b) F2, (c) F3, (d) F4 and (e) F5. Source scale labels: 100 μm. Dark contrast alone does not identify a phase; acquisition files and calibration remain necessary to substantiate this interpretation.

Distinguishing chemistry from structure is vital in this case. The identification of a solute-rich constituent may need elemental enrichment. However, there are no relevant EDS records available for inspection in this study package. The optical darkness alone is not proof of phase presence. The solidification behavior of Alloy 718 is reported in prior literature

[3], [4]. The presence of a constituent rich in Laves seems plausible, while the possible presence of NbC and lack of inspectable Microanalysis require careful terminology. Strengthening this assignment, electron diffraction analysis would separate the optical measurements into contributions from individual phases.

3.4. Measured interdendritic second-phase fraction

The measured fraction increases monotonically from 2.5 ± 0.3 to 5.3 ± 0.6 area% across F1–F5 (Table 5). The rise of 2.8 percentage points is roughly equivalent to a doubling of the mean segmented fraction. Overlapping standard deviations for some adjacent conditions prevent us from taking the monotonic ordering of group means as proof that every adjacent pair differ significantly. The scatter could be explained by differences in etching, threshold selection and the spatial distribution of constituents.

Table 5. Measured interdendritic fraction and predicted Laves fraction.

Filler	Wire Nb (wt.%)	Mean A_f (area%)	SD of A_f (area%)	Predicted Laves (vol.%)
F1	4.75	2.5	0.3	2.1
F2	4.90	3.0	0.4	2.6
F3	5.05	3.6	0.4	3.1
F4	5.20	4.4	0.5	3.9
F5	5.50	5.3	0.6	4.8

Over the ten fields evaluated per composition, the mean and SD were reported. Scatter from the field was confirmed, independent weld replication was not. The measured constituent may contain NbC and other etched interdendritic phases.

Measured fractions exceed the predicted Laves fractions by 0.4–0.5 percentage points. Equation (8) describes the regression plotted in Fig. 3(b), where A_f and f_{Laves} are entered as numerical percentages. The near-unity slope indicates similar relative trends within this dataset. The offset can plausibly reflect additional segmented constituents and differences between the sampling and modelling definitions; it cannot be assigned uniquely to a specific phase or used as a universal correction factor.

$$A_f = 1.0415 f_{Laves} + 0.3231 \quad (R^2 = 0.999) \quad (8)$$

The very high R^2 describes five monotonic composition means. It is not independent validation of the absolute model output, particularly because both measures vary with the same composition series. The measured fraction also samples selected central fields, while the calculation represents an assumed bulk weld composition. Spatial segregation reported in single-track Alloy 718 deposits [14] provides a reason to retain this distinction rather than equating the two quantities.

3.5. Cracking response at 2% nominal strain

Table 6 reports CI at 2% nominal augmented strain. The mean increases from 0.05 for F1 to 0.32 for F5, a 6.4-fold endpoint ratio. With $L_{eval} = 30$ mm, these means correspond arithmetically to summed surface crack lengths of 1.5 and 9.6 mm per evaluated specimen, respectively. These derived lengths provide a physical interpretation of the normalized index and do not represent a separate measurement series. Specimen-linked crack micrographs with documented strain conditions were unavailable. The CI values are based on the tabulated crack-length results.

Table 6. Reported cracking index at 2% nominal augmented strain.

Filler	Wire Nb (wt.%)	CI, dimensionless (mean $\pm$ SD; n = 3 tests)
F1	4.75	0.05 $\pm$ 0.02
F2	4.90	0.08 $\pm$ 0.03
F3	5.05	0.14 $\pm$ 0.04
F4	5.20	0.22 $\pm$ 0.05
F5	5.50	0.32 $\pm$ 0.06

CI = summed visible surface crack length / 30 mm. The original table description identifies three tests per composition at 2% strain. Numerical results at 1% and 3% strain were not supplied.

Figure 5 shows the five mean CI values with standard-deviation error bars and the regression in Eq. (9). Nb denotes filler-wire concentration in wt.%, rather than the unmeasured weld-metal concentration. All five groups have a positive mean CI at 2% strain. The results therefore do not establish a crack-free filler composition under this imposed-strain condition. They also do not establish a critical augmented strain, since the quantitative 1% and 3% data are unavailable.

$$CI = 0.3775 Nb - 1.7556 \quad (R^2 = 0.987) \quad (9)$$

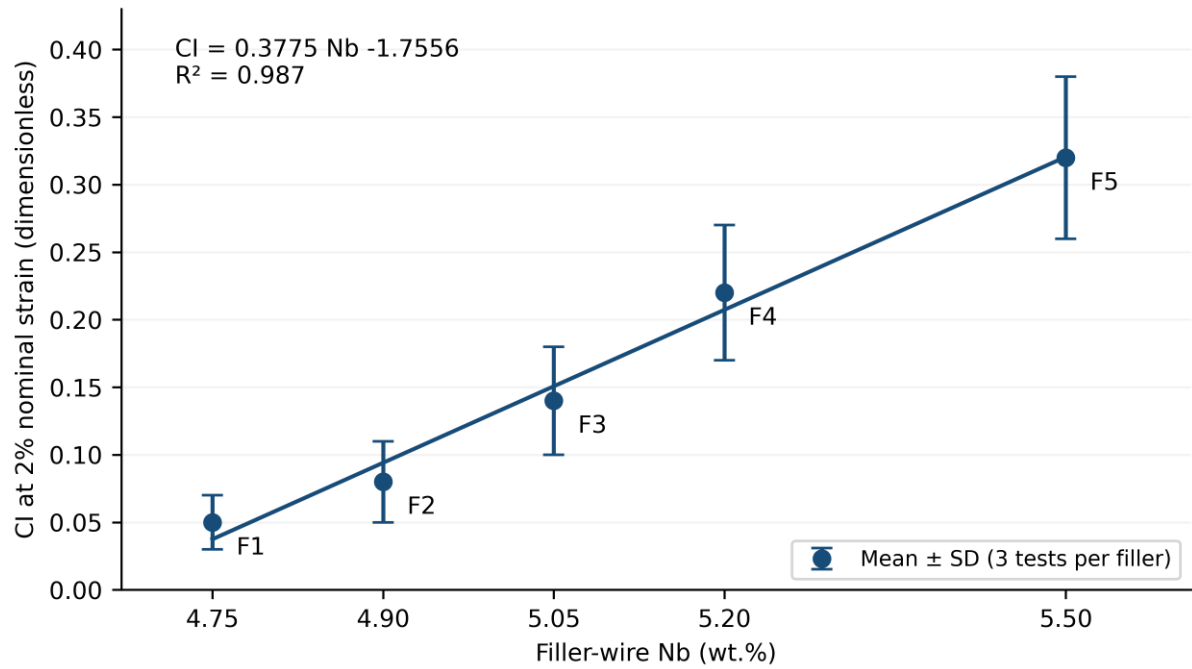


Fig. 5. Cracking index at 2% nominal augmented strain versus filler-wire Nb. Error bars are reported ± 1 SD of three tests per composition; the line is the unweighted fit of the five means in Eq. (9). F1–F5 identify the composition groups.

3.6. Associations with phase fraction

Figure 6 compares CI with the measured interdendritic fraction and the calculated Laves fraction. The corresponding least-squares relationships are given in Eqs. (10) and (11). The coefficients were recalculated from the table means without rounding intermediate calculations. Their R^2 values are 0.994 and 0.995 after rounding, respectively. The negative intercepts are mathematical features of the fitted lines and have no physical interpretation at zero phase fraction.

$$CI = 0.0980 A_f - 0.2066 \quad (R^2 = 0.994) \quad (10)$$

$$CI = 0.1022 f_{Laves} - 0.1752 \quad (R^2 = 0.995) \quad (11)$$

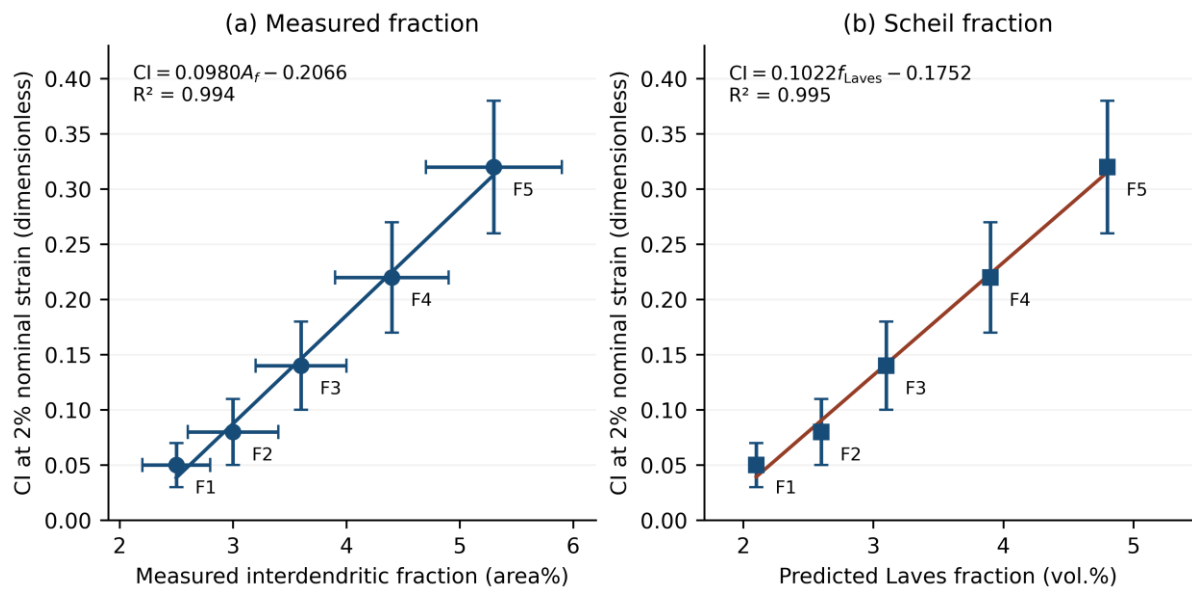


Fig. 6. CI at 2% nominal strain versus (a) measured interdendritic second-phase area fraction and (b) predicted Laves volume fraction. Vertical bars are reported ± 1 SD of three cracking tests. Horizontal bars in (a) are ± 1 SD of ten image fields; these are not paired specimen-level uncertainties. Lines show Eqs. (10) and (11) fitted to five means within the observed predictor ranges.

Filler Nb, measured area fraction and predicted phase fraction co-vary across the same five groups so the two correlations are not independent demonstrations of causality. Additionally, the sampling units of the phase measurements and the crack measurements are different. As a result, the regression indicates a relationship at the composition level but not at the specimen level for crack initiation due to local phase content. To further validate the proposed mechanism, additional independently produced welds, more phase-resolved measurements, and a larger range of measured thermal conditions would be required.

3.7. Mechanism, practical meaning, and limitations

A reasonable interpretation is that an increase in Nb changes the amount and distribution of residual interdendritic liquid during final solidification. When feeding is impeded and solid bridges are incomplete, liquid-film separation can lead to cracking if deformation occurs. The increment in calculated ΔT and phase fraction agrees with this explanation, while the measured CI trend provides its experimental association. Nevertheless, the current measurement does not resolve liquid pressure, permeability, local strain, or the time of crack initiation. The coupled models of Draxler et al. [7]–[9] identify precisely these additional variables. So, the associated mechanistic framework cannot just be described using a phase-fraction-only criterion.

Despite having a constant nominal arc energy, the thermal processing continues to interact. Novel strategies such as mutant current pulsing, arc-deposition cooling are able to modify the solidification environment as well as build interruptions and inter-layer dwell time. Similarly, other Alloy 718 configurations have cracking influenced by weld geometry and initial microstructure [16]–[18]. The present study does not measure whether its composition effect is larger or smaller than process effects. The practical value of the proposed method is to mark the contribution of filler Nb as a variable that needs to be explicitly controlled during procedure development. In addition to heat input, dilution, restraint and base-metal condition.

Fillers with a lower-Nb in this dataset exhibit a lower mean CI but this is not a qualified recommendation for all high-restraint fabrication or repair. The sampled low-Nb wires are F1 at 4.75 wt.% and F2 at 4.90 wt.%; an exact optimum or a universally safe 4.8–5.0 wt.% interval has not been established. The complete chemistry, including Ta and trace elements, must first be verified. Because Nb also contributes to strengthening, hardness, tensile properties, and the response to the intended post-weld treatment must be measured before adopting a filler on the basis of cracking alone.

Hardness comparison was outside the present solidification-cracking assessment, and no microhardness data are available for these welds. Room-temperature indentation would provide complementary property information, but it would not directly measure hot-cracking susceptibility or by itself establish retained γ strengthening. The present conclusions are therefore restricted to the reported cracking and microstructural descriptors. Published TIG-joint hardness measurements [20] and Nb-resolved nanoindentation results [21] justify this additional characterization, but they cannot supply missing values for F1–F5. Similarly, the homogenization and Laves-dissolution literature [22]–[24] and heat-treatment studies [25], [26] inform future experimental design without establishing that the current welds will recover an acceptable property balance after treatment. A smaller as-welded Laves-rich fraction and satisfactory post-weld performance are related questions requiring separate measurements.

The main limitations are of the five-level composition series, one assumed welding condition, estimated dilution, incomplete compositional mass balance, and unavailable calculation settings. Optical segmentation does not yield crystallographically pure Laves phase. Furthermore, its standard deviation shows a spatial rather than between-weld variation. Verification of the timing and the calibration of the imposed bending is required while quantitative cracking data are limited 2% nominal strain. Due to these limitations, the findings are restricted to merely a descriptive comparison of the conditions being reported and cannot be extrapolated to multi-pass joints, other restraint states, or to their service performance.

The establishment of complete plate, wire, and weld-metal analyses along with directly measured dilution and thermal histories should be future work. Optical images, BSE and EDS data calibrated and paired with phase-resolved characterization and repeatable segmentation. Weld reproductions should distinguish between-weld variability from different field variations. Hardness and tensile tests both before and after the proposed heat treatment will test the practical trade-off. Ultimately, the documentation of the bend timing and strain calibration complete with 1–3% strain results would make the cracking-susceptibility assessment more informative.

4. Conclusions

The reported GTAW dataset shows a systematic association between filler-wire Nb content, calculated terminal-solidification descriptors, measured interdendritic second-phase fraction, and cracking at 2% nominal augmented strain. Within the stated experimental and modelling limitations:

1. Increasing filler-wire Nb from 4.75 to 5.50 wt.% increased the calculated operational interval $T_L - T_{99}$ from 106 to 119 °C. The change was dominated by a reduction in T_{99} , rather than a large shift in the liquidus (Table 4).
2. The predicted Laves fraction increased from 2.1 to 4.8 vol.%, while the measured interdendritic fraction increased from 2.5 ± 0.3 to 5.3 ± 0.6 area%. These are distinct quantities and do not establish the same absolute phase content (Table 5).
3. Mean CI at 2% nominal strain increased from 0.05 ± 0.02 to 0.32 ± 0.06 , corresponding to a 6.4-fold endpoint increase. None of the five-composition means was zero under this test condition (Table 6).

4. Regressions against measured and predicted phase fractions gave $R^2 = 0.994$ and 0.995 , respectively. These fits describe five composition means and support an association; they do not independently establish a causal or transferable prediction model (Fig. 6).
5. The lower-Nb wires warrant further investigation as candidates for reduced cracking under comparable conditions. A specification-level recommendation and retention of post-weld strengthening require verified chemistry, experimental records, and mechanical testing.

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