

Effect of Ni–Mn–N alloying on ferrite suppression and weld solidification mode in 316L-series austenitic stainless steels

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Abstract

316L-series austenitic stainless steels are important candidate materials for structural and blanket-related components in nuclear fusion reactors, while small variations in Ni, Mn, and N contents can markedly affect phase constitution, magnetic and electrochemical behavior, and weld-metal solidification. In this study, the microstructures of 316L and 316LN stainless

steels were characterized by EBSD and XRD and correlated with their magnetic and electrochemical responses. Equilibrium phase diagrams and Scheil solidification paths were further calculated for 316L(N)/CLF-1 dissimilar weld metals to clarify the coupled effect of Ni-Mn-N on solidification behavior. 316LN exhibited an almost single-phase austenitic microstructure, whereas 316L retained a small amount of ferrite, giving relative permeabilities of 1.004 and 1.165 at 300 K, respectively. The more homogeneous austenitic microstructure of 316LN improved passive-film stability, with a polarization resistance approximately 2.3 times that of 316L. Low Ni-Mn-N weld metal solidified predominantly through the $L \rightarrow L + \delta$ path, with limited $L + \delta \rightarrow \gamma$ peritectic reaction and incomplete subsequent $\delta \rightarrow \gamma$ transformation, resulting in a ferrite + martensite microstructure. Higher Ni-Mn-N contents expanded the γ stability field and promoted more extensive peritectic and $\delta \rightarrow \gamma$ transformations. Meanwhile, the calculated M_s decreased by approximately 30 °C, enhancing austenite stability and producing columnar austenite with interdendritic lath martensite. These results demonstrate the sensitivity of phase stability and weld solidification to coupled Ni-Mn-N variations and provide a thermodynamic basis for composition and solidification-mode control in 316L(N)/CLF-1 dissimilar joints.

Keywords: 316L-series austenitic stainless steels; Ni-Mn-N alloying; Ferrite suppression; Weld solidification mode; Nuclear fusion application

1. Introduction

Austenitic stainless steel is one of the most widely used structural materials in nuclear engineering [1-3]. Low-carbon 316L stainless steel has long been used for critical load-bearing components, such as the pressure vessels, main pipelines and steam generator tube sheets in pressurized water reactors, due to its excellent corrosion resistance, toughness and high-temperature mechanical properties [4, 5]. In nuclear fusion reactors, 316L stainless steel has also been selected as a candidate material for critical components such as vacuum vessel shells and support structures [6-8]. Furthermore, 316LN stainless steel, which is made by increasing the nitrogen content of the 316L stainless steel offers good corrosion resistance and significantly enhanced yield strength, as well as improved resistance to radiation swelling and stress corrosion cracking [9-11]. However, fusion reactors have stringent low-radioactivity requirements and the International Thermonuclear Experimental

Reactor (ITER) therefore imposes strict limits on long-lived radioactive impurity elements such as Co, Nb and Ta that exceed those found in commercial 316L(N) compositions, forming the ITER-grade material system that balances low radioactivity, corrosion resistance, weldability, and high-temperature mechanical properties. It is suitable for pipelines subjected to higher thermal loads and shielding components exposed to high radiation doses [12-14].

The service performance of austenitic stainless steels is strongly affected by the contents of alloying elements such as N, Mn, and Ni [15-20]. Nitrogen is widely regarded as one of the most effective austenite-stabilizing elements in 316L-series austenitic stainless steels. Increased nitrogen levels provide solid-solution strengthening, refine austenite grains, and improve yield strength and the strength-ductility balance. In addition, N can significantly reduce the volume fractions of ferrite and martensite, thereby lowering the ferromagnetic response. It also leads to clear advantages over low-N 316L stainless steel in electrochemical metrics such as the pitting potential and passive-film stability, enabling a favorable combination of strength and localized corrosion resistance [21, 22]. Manganese is commonly used as an austenite stabilizer to partially substitute for Ni. By adjusting the Mn/N ratio, the hot-workability and formability can be improved and the yield strength can be increased. Mn also enhances the solubility of N in austenite, thereby providing additional compositional latitude for high-nitrogen alloying. However, increasing Mn content alters the composition of corrosion products and oxide films, which can introduce more complex effects on corrosion resistance and high-temperature oxidation behavior [16, 23]. Nickel is the most important substitutional austenite-stabilizing element in the 316L series. Numerous studies have shown that increasing Ni content helps suppress the formation of ferrite and martensite and promotes austenite formation [24, 25]. Nevertheless, excessive Ni can inhibit strain-induced martensitic transformation and the associated transformation-induced plasticity effect, thereby reducing the work-hardening capability of the material [26].

Although existing research provides valuable insights into how Ni, Mn and N influence microstructure and properties, most studies focus on the regulation of individual elements. Systematic, quantitative descriptions of the synergistic effects of Ni, Mn and N within the same material system remains lacking, particularly under welding conditions. Under welding conditions in particular, most studies have focused solely on the Cr_{eq}/Ni_{eq} ratio or

on individual elements, discussing either ferrite content or crack sensitivity [27-31]. However, there is still a lack of clear analysis of how the collective content of nitrogen, manganese, and nickel influences the solidification patterns of weld metal. Therefore, a systematic investigation into the synergistic effects of Ni, Mn and N on microstructure and properties is essential.

This study investigates the use of 316L and 316LN austenitic stainless steels in nuclear fusion reactors. The chemical composition and microstructure of the two are compared, with EBSD and SEM utilized to clarify the distribution characteristics and phase fraction differences of ferrite (δ), austenite (γ) and martensite in both steels. Combined with magnetic and electrochemical testing, a correlation is established between the base material composition, microstructure, and magnetic and electrochemical properties. Furthermore, thermodynamic calculations were performed for dissimilar welded joints between 316L(N) stainless steel and CLF-1 steel based on the actual weld Ni-Mn-N content, in order to derive equilibrium phase diagrams and non-equilibrium solidification paths at different Ni-Mn-N concentrations. This clarifies the effect of variations in Ni, Mn and N on weld solidification patterns. The study systematically reveals the intrinsic relationship between microstructure, magnetic and electrochemical behavior, and solidification patterns in 316 series austenitic stainless steel under varying Ni-Mn-N compositions. The study provides a theoretical basis for optimizing alloy composition and enabling microstructural design in dissimilar joints between 316 series austenitic stainless steel and CLF-1 steel for use in nuclear fusion reactors.

2. Materials and methods

2.1. Material

Plates of 316L austenitic stainless steel, 316LN austenitic stainless steel and CLF-1 steel with a thickness of 5 mm were used in this study. The measured chemical compositions of 316L, 316LN and CLF-1 are listed in Table 1. The microstructure of 316L consists of austenite with a small amount of ferrite. The microstructure of 316LN austenitic stainless steel is fully austenitic, which is suitable for non-magnetic and cryogenic applications. The base metal of CLF-1 exhibits a tempered martensitic microstructure.

Table 1 Chemical composition of the 316L, 316LN austenitic stainless steel and CLF-1 steel (wt.%).

Materials	C	Cr	W	Mo	Mn	Si	V	Ta	N	Ni	Fe
316L	0.020	15.92	/	2.08	0.97	0.27	0.05	/	0.03	10.04	Bal.
316LN	0.022	16.07	/	2.13	1.7	0.32	0.025	/	0.14	12.40	Bal.
CLF-1	0.11	7.97	0.98	/	0.51	/	0.26	0.14	0.006	0.22	Bal.

2.2. Physical property test

The magnetic properties of the materials were characterized by temperature-dependent magnetization measurements using a superconducting quantum interference device (SQUID, Quantum Design). Specimens were sectioned from the weld cross-sections, mechanically ground and polished to 0.05 μm , then sequentially cleaned with deionized water and ethanol, dried, and mounted in non-magnetic straws or quartz tubes fixed at the center of the detection coils. The measurements were carried out under a direct current magnetic field of $H = 1000$ Oe over the temperature range of 2-300 K, with a heating rate of 3 K/min. The instrument output the magnetization $\mu(T)$ as a function of temperature. Based on the specimen mass and nominal chemical composition, $\mu(T)$ was converted to the volume-based relative magnetization and the corresponding magnetic parameters, which were used for comparative analysis of different base materials and welded regions.

2.3. Chemical property test

Electrochemical measurements were conducted using a Corrtest350 electrochemical workstation in a conventional three-electrode cell. 316L stainless steel and 316LN stainless steel were used as working electrodes, a saturated Ag/AgCl electrode served as the reference electrode, and a platinum electrode was used as the counter electrode. A 3.5 wt.% NaCl aqueous solution was employed as the electrolyte, and all tests were performed at room temperature. After grinding and polishing, the specimens were cleaned with deionized water and ethanol and then dried. The open-circuit potential (OCP) was monitored for 1200 s with a sampling frequency of 1 Hz. Subsequently, electrochemical impedance spectroscopy (EIS) was carried out at the stabilized OCP. The impedance spectra were fitted to an equivalent circuit using ZSimpWin software to obtain the charge-transfer resistance and the parameters of the constant phase element (CPE). Potentiodynamic polarization tests were then performed, and the corrosion potential (E_{corr}) and corrosion current density (i_{corr}) were

determined by Tafel extrapolation.

2.4. Welding property test

Laser welding experiments were carried out using a high-power dual-mode continuous-wave fiber laser (IPG YLS-30000) with a maximum output power of 30 kW, a laser wavelength of 1070 nm, and a focused spot diameter of approximately 600 μm . The welding head was mounted on the end-effector of a six-axis KUKA industrial robot to achieve precise control of the welding path and travel speed. To protect the optical components of the laser welding system from potential contamination by metal vapor and spatters, the robot was tilted backward by 10° during the welding process.

2.5. Thermodynamic calculations

Thermodynamic calculations were performed using Thermo-Calc 2026b with the TCFE12 database. The experimentally measured weld-metal compositions were used as the reference compositions, with Fe taken as balance. Equilibrium phase-fraction calculations were conducted at a pressure of 1 atm over the temperature range of 350-1700 $^\circ\text{C}$. Non-equilibrium solidification was simulated using the Scheil model, assuming complete mixing in the liquid, negligible diffusion in the solid phases, and local equilibrium at the solid/liquid interface.

The investigated Ni, Mn, and N composition ranges were selected with reference to the experimentally measured compositions of the 316L/CLF-1 and 316LN/CLF-1 weld metals. For the parametric calculations, Ni, Mn, or N was varied individually while the other major alloying elements were maintained at the reference composition, allowing the individual effect of each element on phase stability and solidification behavior to be evaluated. The experimentally measured weld-metal compositions were subsequently used to assess the coupled influence of Ni-Mn-N under realistic weld-metal conditions.

3. Results

3.1. Microstructure of 316L/316LN stainless steel

Fig. 1(a) shows that fine recrystallized equiaxed grains with well-defined grain boundaries exist in the 316L stainless steel, with discontinuous, elongated band-like structures embedded in the matrix. The corresponding IPF diagram reveals that the equiaxed grains are finely-grained austenite with multiple orientations. According to the phase distribution diagram (Fig. 1(c)), the discontinuous band-like structures distributed in the matrix are

ferrite. In contrast, Fig. 1(d) shows that the equiaxed grains in 316LN stainless steel are significantly coarser with straight, continuous grain boundaries. The IPF diagram (Fig. 1(e)) reveals visible annealed twin layers in 316LN stainless steel. The phase distribution diagram (Fig. 1(f)) indicates a near-single-phase austenite matrix with only sporadic ferrite at the pixel level. The XRD patterns (Fig. 1(g)) reveal that both 316L and 316LN stainless steels are predominantly austenitic, with main diffraction peaks corresponding to the (111), (002), and (220) crystal planes, respectively. However, the XRD pattern of 316L stainless steel also exhibits a ferrite diffraction peak at 44.84° corresponding to the (110) crystal plane, consistent with the EBSD results.

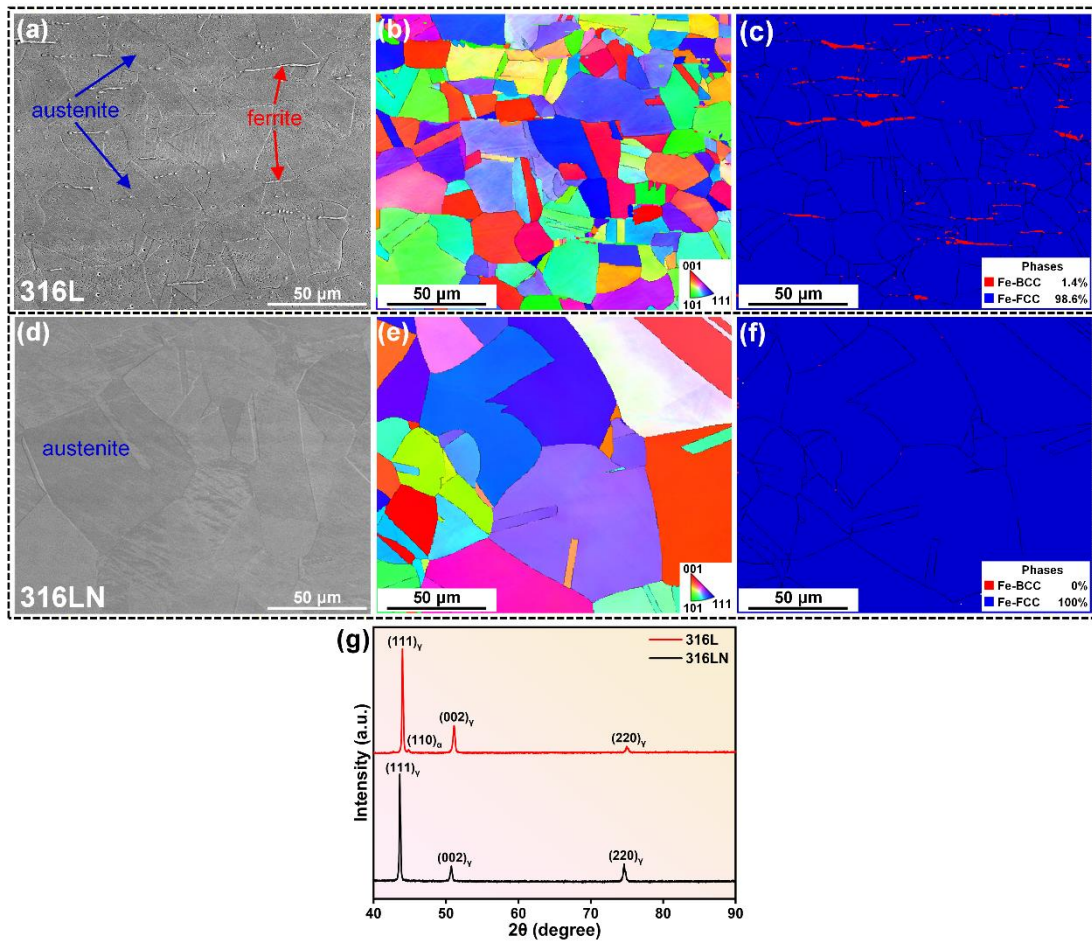


Fig. 1 Differences in microstructure and composition of 316L and 316LN stainless steel: (a) 316L SEM images, (b) 316L IPF maps, (c) 316L phase distribution, (d) 316LN SEM images, (e) 316LN IPF maps, (f) 316LN phase distribution, (g) 316L and 316LN XRD patterns.

3.2. Physical property of 316L/316LN stainless steel

In the 316L series of stainless steels, ferrite exhibits distinct ferromagnetism at room temperature, while austenite displays paramagnetic. Therefore, measuring magnetic parameters is a sensitive, indirect method of revealing and comparing subtle microstructural

differences between 316L and 316LN stainless steels.

Fig. 2 shows the relative permeability and magnetization of 316L and 316LN stainless steels. The relative permeability of both stainless steel materials peaks at 40 to 50 K and then decreases monotonically as the temperature increases. However, within the experimental temperature range, the relative permeability of 316L stainless steel is consistently higher than that of 316LN stainless steel. The relative permeability of 316L stainless steel ranges from approximately 1.165 to 1.182, while that of 316LN stainless steel remains within the narrower range of 1.002 to 1.017. In terms of temperature dependence, both the amplitude and slope of the curve for 316L stainless steel are significantly greater than those of the 316LN stainless steel. Furthermore, the relative permeability is a dimensionless parameter defined as the ratio of the magnetic permeability of a material to that of free space [32]. For austenitic stainless steels, relative permeability is sensitive to the presence and fraction of ferromagnetic phases and can therefore be used to characterize subtle differences in phase constitution [33]. As shown in Table 2, at 300 K (room temperature), the relative permeability of 316LN stainless steel is 1.00363, which is very close to 1 and indicates paramagnetic behavior. In contrast, the relative permeability of 316L stainless steel is 1.16491, which significantly deviates from 1, demonstrating a strong magnetic response and exhibiting weak ferromagnetic properties. This difference in permeability indicates that the ferrite content in 316L stainless steel is significantly higher than that in 316LN stainless steel. Furthermore, further analysis of the magnetization intensity can more clearly reveal the fundamental differences in microstructure between the two, particularly in terms of ferrite content.

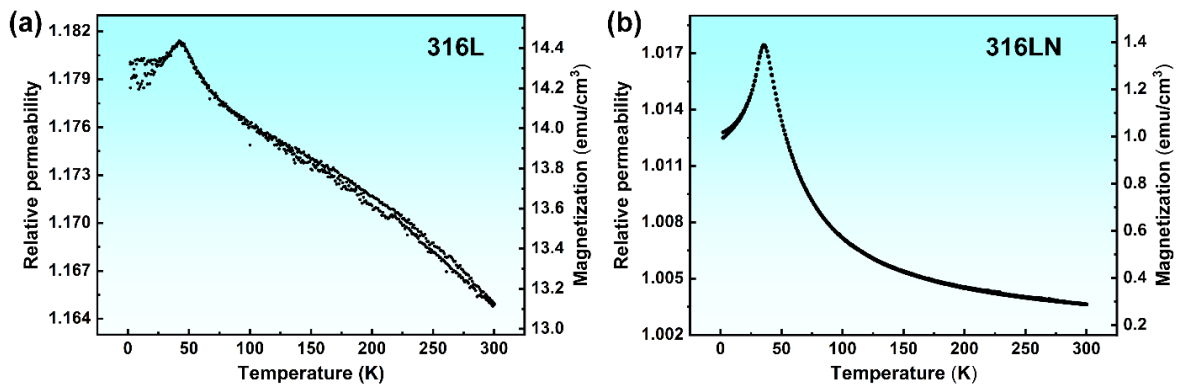


Fig. 2 Relative permeability and magnetization of 316L series stainless steels: (a) 316L stainless steel, (b) 316LN stainless steel.

Table 2 Relative permeability and magnetization of 316L and 316LN stainless steel at 300K.

Sample	Temperature (K)	Relative permeability	Magnetization (emu/cm ³)
316L	300	1.16491	13.12337
316LN	300	1.00363	0.28918

The equivalent magnetization (M) under a constant magnetic field strength (H) of 1000 Oe is shown on the right axis of Fig. 2, calculated according to Eq. (1).

$$M = (\mu_r - 1)H \quad (1)$$

Where μ_r denotes relative magnetic permeability and H represents magnetic field strength. At 300 K (room temperature), calculations indicate that the magnetization of 316L stainless steel is approximately 13.12337 emu/cm³, whereas the magnetization of 316LN stainless steel is only 0.28918 emu/cm³. Under identical external magnetic field conditions, this indicates that the magnetization of 316L stainless steel is approximately 45 times greater than that of 316LN stainless steel. There is a significant difference in magnetic behavior between the two stainless steels, which demonstrates that the substantial ferrite content in 316L stainless steel markedly enhances its magnetization.

3.3. Chemical property of 316L/316LN stainless steel

Fig. 3 illustrates the electrochemical characteristics of 316L and 316LN stainless steels. Under identical test conditions, electrochemical impedance spectroscopy (EIS) reveals differences between the two materials. The Nyquist plot (Fig. 3(a)) shows that the fitted curve for 316LN stainless steel exhibits a larger capacitive arc, whereas the corresponding curve for 316L demonstrates a significantly smaller capacitive arc. The Bode plot (Fig. 3(b)) indicates that 316LN exhibits higher $|Z|$ values across the entire frequency range, with more pronounced phase angle peaks and high-phase angle plateaus, whereas 316L demonstrates relatively lower $|Z|$ values and phase angles. Additionally, parameters obtained from equivalent circuit fitting are presented in Table 3. For 316LN stainless steel: solution resistance $R_s=6.79 \Omega \cdot \text{cm}^2$; high-frequency constant phase element (CPE) parameters: $Q_{\text{coat}}=4.44 \times 10^{-5} \Omega^{-1} \cdot \text{s}^n \cdot \text{cm}^{-2}$, $n_{\text{coat}}=0.9215$; high-frequency resistance $R_{\text{coat}}=223170 \Omega \cdot \text{cm}^2$; Low-frequency CPE parameters: $Q_{\text{dl}}=1.03 \times 10^{-4} \Omega^{-1} \cdot \text{s}^n \cdot \text{cm}^{-2}$, charge transfer resistance $R_{\text{ct}}=48092 \Omega \cdot \text{cm}^2$, polarization resistance $R_p=271262 \Omega \cdot \text{cm}^2$. For comparison, 316L stainless steel has $R_s=8.79 \Omega \cdot \text{cm}^2$, $Q_{\text{coat}}=5.37 \times 10^{-5} \Omega^{-1} \cdot \text{s}^n \cdot \text{cm}^{-2}$, $n_{\text{coat}}=0.9058$, $R_{\text{coat}}=97510$

$\Omega \cdot \text{cm}^2$; $Q_{dl}=7.18 \times 10^{-5} \Omega^{-1} \cdot \text{s}^n \cdot \text{cm}^{-2}$, $R_{ct}=22140 \Omega \cdot \text{cm}^2$, $R_p=119290 \Omega \cdot \text{cm}^2$. It is evident that the polarization resistance of 316LN stainless steel is significantly higher than that of 316L stainless steel.

As shown in Fig. 3(c) and Table 4, the dynamic polarization curves provide corrosion kinetic parameters. For 316LN stainless steel, the corrosion current density (i_{corr}) is $3.19 \times 10^{-8} \text{ A/cm}^2$, the corrosion potential (E_{corr}) is $-0.227 \text{ V vs. Ag/AgCl (1M KCl)}$, and the anodic (b_a) and cathodic (b_c) Tafel slopes are 64.19 and -32.59 mV/dec , respectively. For 316L stainless steel, the values are $i_{corr} = 4.28 \times 10^{-8} \text{ A/cm}^2$, $E_{corr} = -0.229 \text{ V vs. Ag/AgCl (1M KCl)}$, $b_a = 62.79 \text{ mV/dec}$, and $b_c = -24.04 \text{ mV/dec}$.

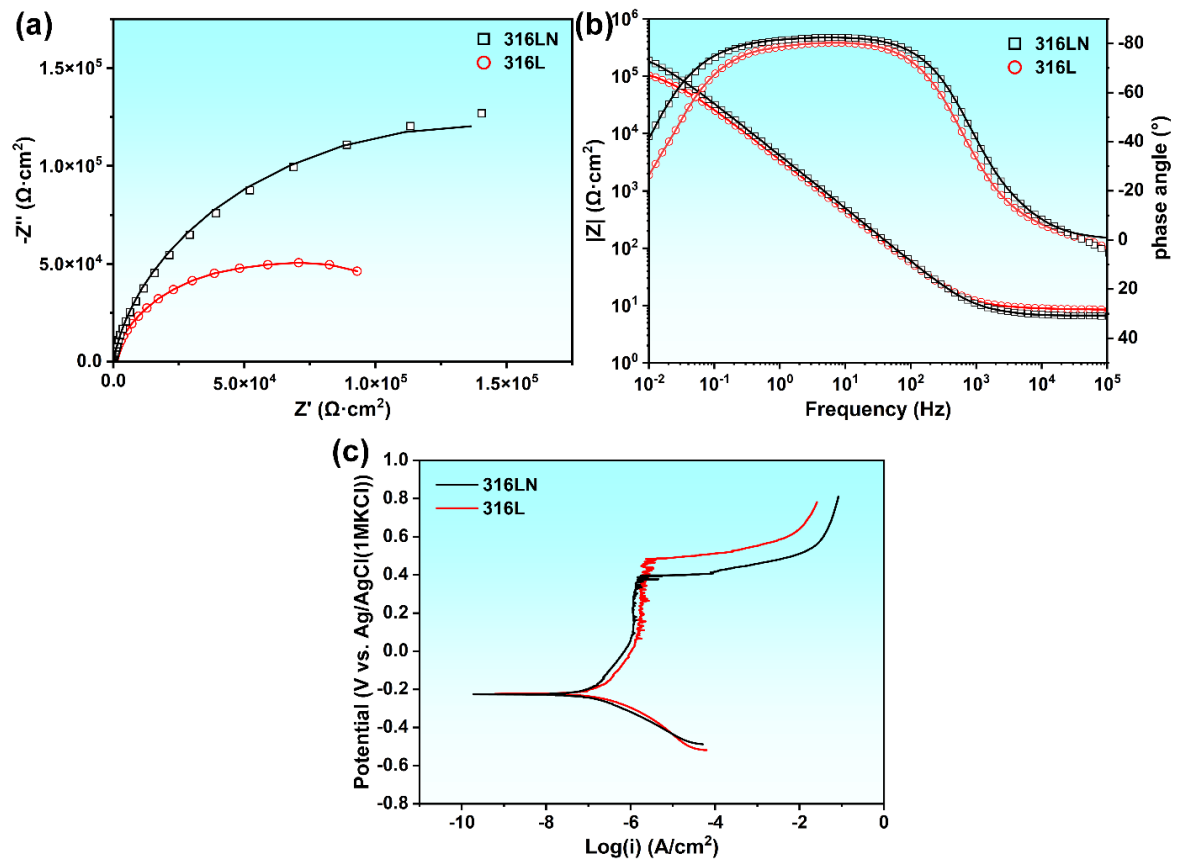


Fig. 3 Electrochemical characterization of 316L and 316LN stainless steel: (a) Nyquist plots, (b) Bode plot, (c) potentiodynamic polarization curves.

Table 3 Parameters from EIS measurements.

Sample	R_s ($\Omega \cdot \text{cm}^2$)	Q_{coat} ($\Omega^{-1} \text{ s}^{-n} \text{ cm}^{-2}$)	n_{coat}	R_{coat} ($\Omega \cdot \text{cm}^2$)	Q_{dl} ($\Omega^{-1} \text{ s}^{-n} \text{ cm}^{-2}$)	R_{ct} ($\Omega \cdot \text{cm}^2$)	R_p ($\Omega \cdot \text{cm}^2$)
316L	8.79	5.37×10^{-5}	0.9058	97150	7.18×10^{-5}	22140	119290
316LN	6.79	4.44×10^{-5}	0.9215	223170	1.03×10^{-4}	48092	271262

Table 4 Fitting parameters by Tafel polarization curves of samples.

Sample	i_{corr} (A/cm ²)	E_{corr} (V vs. Ag/AgCl (1M KCl))	b_a (mV/dec)	b_c (mV/dec)
316L	4.28×10^{-8}	-0.229	62.79	-24.04
316LN	3.19×10^{-8}	-0.227	64.19	-32.59

Combining EIS and polarization parameters, 316LN stainless steel exhibits higher impedance and polarization resistance along with lower corrosion current density, whereas 316L stainless steel shows relatively lower values.

3.4. Microstructure of 316L(N)/CLF-1 dissimilar steel welds

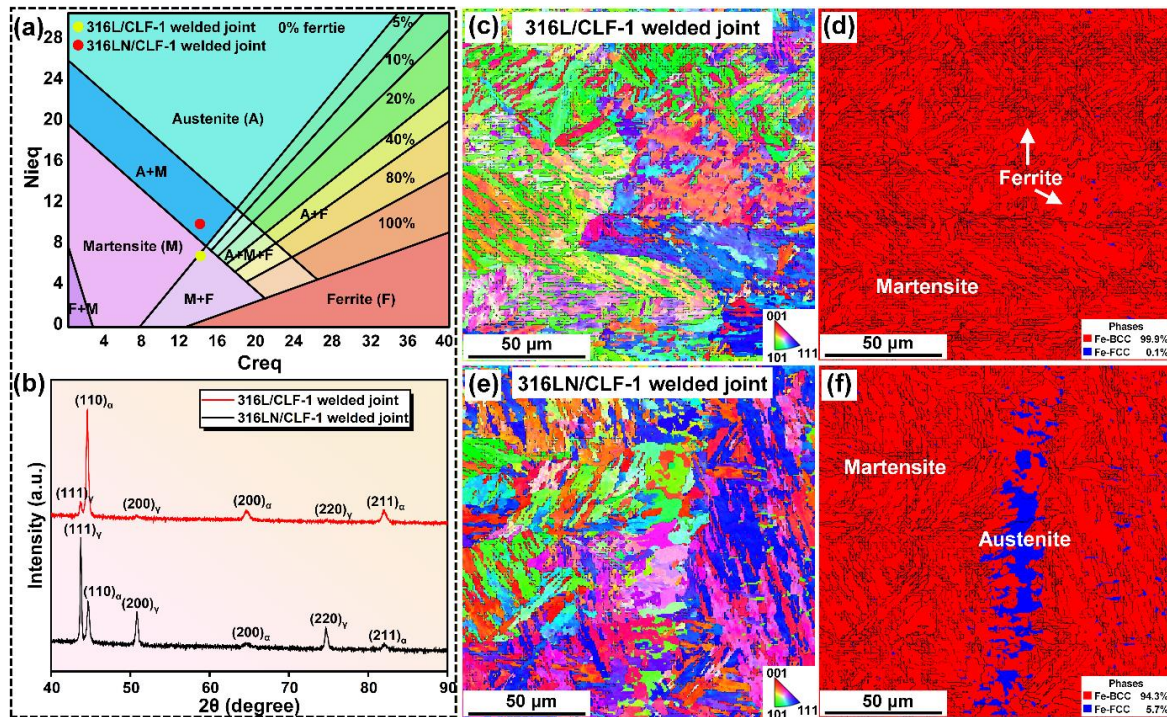


Fig. 4 Analysis results of the 316L(N)/CLF-1 dissimilar steel welds: (a) Schaeffler diagram, (b) XRD pattern, and (c)-(f) EBSD results.

Table 5 Welding element composition of 316L(N)/CLF-1 dissimilar steel welds

Materials	C	Cr	W	Mo	Mn	Si	V	Ta	N	Ni	Fe
316L/CLF-1 welded joint	0.055	12.36	0.44	1.20	0.71	0.21	0.12	/	0.035	6.14	Bal.
316LN/CLF-1 welded joint	0.054	12.39	0.49	1.09	1.04	0.24	0.14	/	0.075	7.06	Bal.

Fig. 4 shows the Schaeffler diagrams, XRD patterns, and EBSD results for the 316L(N)/CLF-1 dissimilar steel welds. Using the Schaeffler equation, the microstructural composition of the weld metal was predicted based on the actual measured weld composition in Table 5. Since the Schaeffler equation does not account for Ta (the primary alloying element in Reduced activation ferritic/martensitic steel), Ni in the equation was

replaced with Ta, which has the same coefficient. Substituting the elemental composition data from Table 5 into Eq. (2) and Eq. (3), we calculated the Ni_{eq} in the 316LN/CLF-1 weld to be 9.9% and the Cr_{eq} to be 13.8%. For the 316L/CLF-1 weld, the Ni_{eq} and Cr_{eq} were 6.8% and 13.9%, respectively.

$$Cr_{eq} = Cr + Mo + 1.5Si + 0.5Ta \quad (2)$$

$$Ni_{eq} = Ni + 30C + 0.5Mn \quad (3)$$

As shown in the Schaeffler diagram (Fig. 4(a)), the microstructure of the 316L/CLF-1 dissimilar steel weld comprises austenitic and martensitic phases, whereas the microstructure of the 316LN/CLF-1 dissimilar steel weld consists of ferritic and martensitic phases. Based on the XRD results of the longitudinal section of the weld, both zones of the two joints are composed of α -Fe and γ -Fe, but the proportions of these phases differ significantly. Specifically, the primary phase in the 316L/CLF-1 dissimilar steel weld is α -Fe, whereas in the 316LN/CLF-1 dissimilar steel weld is γ -Fe. Combining the IPF and phase diagrams at the weld zone shows that the microstructure of the 316L/CLF-1 dissimilar steel weld consists of lath martensite and blocky ferrite, and that of the 316LN/CLF-1 dissimilar steel weld comprises columnar austenite and lath martensite. These results align with predictions from the Schaeffler diagram and XRD pattern analysis.

4. Discussion

4.1 Effects of Ni, Mn, and N content on ferrite content, magnetic properties, and electrochemical performance of 316-series stainless steel

Variations in the Ni, Mn, and N contents exert a pronounced influence on the ferrite fraction, magnetic response, and electrochemical behavior of 316-series stainless steels. Compared with 316L stainless steel, the higher Ni and Mn levels in 316LN stainless steel markedly increase the nickel equivalent, thereby promoting austenite stability and narrowing the thermal stability range of ferrite. In addition, because the solubility of nitrogen is higher in austenite than in ferrite, N preferentially partitions to and becomes enriched in the austenitic phase during solidification. Ni and Mn also reduce the martensite start temperature (M_s) [34], suppressing the austenite-to-martensite transformation. Consequently, 316LN stainless steel exhibits an almost single-phase austenitic microstructure at room temperature, with essentially no ferrite present. By contrast, 316L stainless steel has a higher Cr_{eq}/Ni_{eq}

ratio and a lower N content, which enhance ferrite stability and lead to a small amount of retained ferrite [30, 35]. Since ferrite is ferromagnetic whereas austenite is paramagnetic, even limited ferrite significantly increases the magnetization response and causes the relative magnetic permeability to deviate noticeably from unity, consistent with the magnetization measurements as well as the EBSD and XRD results obtained in this study. Regarding electrochemical corrosion behavior, the higher ferrite volume fraction in 316L stainless steel, together with the compositional heterogeneity between ferrite and austenite, gives rise to a larger number of ferrite/austenite micro-galvanic couples at the surface. These sites are prone to localized passivity breakdown and preferential anodic dissolution, which manifests in the polarization tests as a higher corrosion current density and a lower polarization resistance (Fig. 3(c) and Table 4). In contrast, owing to the higher Ni, Mn, and N contents, 316LN stainless steel exhibits a microstructure that is closer to single-phase austenite at room temperature, thereby markedly reducing the interfacial potential differences and the associated electrochemical heterogeneity on the macroscopic scale. Meanwhile, Ni and N promote the formation of a more compositionally uniform, defect-lean Cr/N-enriched passive film, and Mn further stabilizes this film by increasing N solubility. As a result, interfacial charge transfer is more strongly inhibited and the overall corrosion kinetics are significantly passivated [21, 36]. Accordingly, compared with 316L stainless steel, which contains a higher ferrite fraction and therefore shows more pronounced localized anodic activity, 316LN stainless steel develops a more stable passive film and exhibits a lower overall corrosion rate under the given electrochemical conditions, providing mechanistic support for its superior corrosion resistance.

4.2 Effect of Ni, Mn, and N content on the solidification behavior of 316L(N)/CLF-1 dissimilar steel weld metal

Analysis of the equilibrium phase diagrams and the non-equilibrium solidification paths at different Ni, Mn, and N levels (detailed equilibrium phase diagrams for varying Ni, Mn, N levels are provided in Supplementary Figs. S1-S4) indicates that the three alloying elements exert distinctly different effects on the weld metal of the 316L(N)/CLF-1 dissimilar steels joint. Ni, which is a typical substitutional austenite stabilizer, substantially expands the austenite stability region at intermediate and high temperatures, increases the equilibrium fraction of austenite, and promotes the participation of the peritectic reaction ($L + \delta \rightarrow \gamma$).

As a result, austenite becomes the dominant phase during solidification, driving the solidification mode transition from FA to AF. By contrast, Mn, within the investigated composition range, also acts as an austenite stabilizer, but its effect is primarily manifested in the intermediate-temperature $\delta + \gamma$ two-phase field, where it moderately increases the γ fraction and reduces the final δ content. The influence of Mn on the liquidus temperature and the solidification trajectory is relatively mild, and Mn alone is insufficient to alter the identity of the primary phase. In comparison, N exhibits a very strong austenite-stabilizing effect even at extremely low additions [37]. Within the present range, N markedly increases the peak γ fraction in the intermediate-temperature $\delta + \gamma$ two-phase field and significantly decreases M_s [38], while the solidification path remains $L \rightarrow L + \delta$. The γ phase is therefore formed via solid-state transformation during cooling, and the addition of N does not change the solidification mode; instead, it controls the microstructural evolution by modifying the subsequent solid-state transformation and the room-temperature microstructure. These results also indicate that the coupled effect of Ni, Mn, and N cannot be regarded simply as an arithmetic superposition of their individual austenite-stabilizing effects. Ni markedly modifies the high-temperature γ stability field and solidification trajectory, Mn moderately adjusts the δ/γ phase balance within the investigated composition range, whereas N strongly stabilizes γ and decreases M_s . When Ni, Mn, and N vary simultaneously, their effects become coupled through the multicomponent phase equilibria, changing the relative thermodynamic stability of the liquid, δ , and γ phases and shifting the corresponding phase boundaries. Consequently, both the extent of γ formation during solidification and the subsequent $\delta \rightarrow \gamma$ and $\gamma \rightarrow$ martensite transformations are modified.

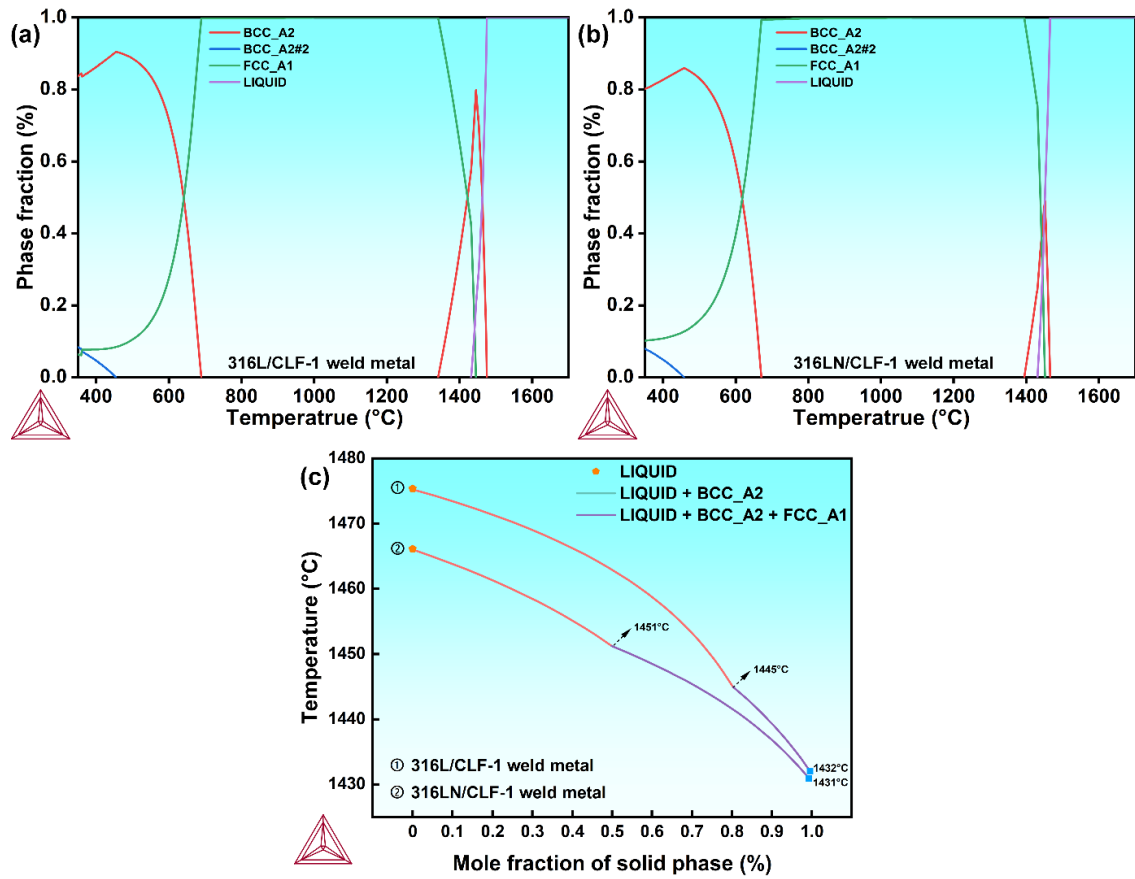


Fig. 5 Equilibrium phase diagrams and non-equilibrium solidification paths of the weld metal in 316L(N)/CLF-1 dissimilar steels joints, calculated based on the weld composition: (a) equilibrium phase diagram of the 316L/CLF-1 dissimilar steels weld, (b) equilibrium phase diagram of the 316LN/CLF-1 dissimilar steels weld, and (c) non-equilibrium solidification path of the 316L(N)/CLF-1 dissimilar steels weld metal.

As shown in Fig. 5, the weld metal with relatively low Ni-Mn-N content exhibits a broader ferrite stability temperature range in the equilibrium phase diagram, whereas the $\delta + \gamma$ two-phase field is confined to a very narrow temperature interval in the intermediate-temperature regime. Consistently, the corresponding non-equilibrium solidification path remains almost entirely within the $L + \delta$ field and only briefly intersects the $L + \delta + \gamma$ three-phase field over a very limited temperature range near the end of solidification. Therefore, solidification under this composition is dominated by primary ferrite, and only a small amount of austenite forms at the final stage of solidification via the peritectic reaction $L + \delta \rightarrow \gamma$ [39]. During subsequent cooling, the extremely high cooling rate associated with laser welding prevents the $\delta \rightarrow \gamma$ solid-state transformation from approaching equilibrium, thereby retaining a substantial fraction of primary ferrite to room temperature. Moreover, the lower Ni-Mn-N levels correspond to a higher M_s , which causes the limited austenite formed in this composition range to transform fully into martensite upon cooling below M_s . Consequently,

the room-temperature microstructure consists of a massive ferritic matrix with scarce grain boundaries and lath martensite, as shown in Fig. 4(c) and Fig. 4(d). This microstructural feature is in good agreement with the calculated equilibrium phase diagram and the non-equilibrium solidification path.

The calculated equilibrium phase diagram of the weld metal with higher Ni-Mn-N contents indicates a pronounced expansion of the γ stability field and a clear contraction of the δ stability field. Consistently, the corresponding non-equilibrium solidification path traverses the $L + \delta + \gamma$ three-phase region over a relatively wide solid-fraction range, which enables extensive progress of the peritectic reaction ($L + \delta \rightarrow \gamma$) during solidification. Quantitatively, the high-Ni-Mn-N weld metal enters the $L + \delta + \gamma$ three-phase region at approximately 1451 °C and a solid fraction of ~ 0.50 , whereas the low-Ni-Mn-N weld metal reaches this region at approximately 1445 °C and a solid fraction of ~ 0.80 . This difference indicates that increasing the Ni-Mn-N content advances the participation of γ in the solidification process and extends the solidification interval involving the $L + \delta \rightarrow \gamma$ peritectic reaction. Consequently, the solidification process becomes γ -dominated. This behavior is evident in the columnar austenitic dendritic morphology observed in SEM images, as shown in Fig. 4(e) and Fig. 4(f). Near the end of solidification, the small amount of residual ferrite is further consumed during subsequent cooling through the $\delta \rightarrow \gamma$ solid-state transformation, such that the high-temperature microstructure prior to martensitic transformation becomes nearly fully austenitized and the ferrite volume fraction decreases to a very low level. The calculated M_s value supports this conclusion: relative to the low Ni-Mn-N weld metal, the M_s value of the high Ni-Mn-N weld metal is reduced by approximately 30 °C, indicating stronger austenite stability. However, the M_s values of both weld metals remain well above room temperature. Therefore, the austenite formed during solidification and the subsequent solid-state transformations undergoes only partial $\gamma \rightarrow M$ transformation upon cooling. The local depletion of Ni, Mn, and N between austenitic dendrites raises the local M_s , leading to an almost complete transformation of austenite into lath martensite. Conversely, in dendrite core regions, the enrichment of Ni, Mn, and N significantly lowers the local M_s , thereby retaining a higher fraction of austenite at room temperature. This compositional partitioning results in an interwoven microstructure comprising continuously distributed columnar austenitic dendrites and dense lath martensite.

The marked difference between the $\delta + M$ microstructure in the low Ni-Mn-N weld metal and the $\gamma + M$ microstructure in the high Ni-Mn-N weld metal stems from the fact that the two compositions are located in different primary-phase stability regions of the multicomponent phase diagram. Even minor variations in the Ni-Mn-N content can shift the solidification path from an FA mode characterized by primary ferrite, in which the primary ferrite is only partially transformed into austenite during post-solidification solid-state reactions, to an AF mode dominated by primary austenitic dendrites, in which ferrite is largely consumed through the peritectic reaction and subsequent solid-state transformation. Such compositional sensitivity has also been observed in 316L stainless steels, where relatively small variations in alloy composition resulted in distinct solidification modes and corresponding differences in the amount and morphology of retained δ -ferrite [40]. However, ferrite suppression should not be regarded as unconditionally beneficial for weld performance. Although reducing the retained ferrite fraction promotes a more austenitic microstructure, excessive ferrite suppression may increase solidification-cracking susceptibility, whereas retained ferrite can undergo subsequent decomposition during thermal exposure, potentially affecting long-term microstructural stability [41]. Therefore, optimization of the Ni-Mn-N composition should balance ferrite control, weldability, and long-term phase stability rather than simply pursue the complete elimination of ferrite.

5. Conclusion

This study systematically elucidates the roles of Ni, Mn, and N contents in governing the microstructure, magnetic response, and electrochemical behavior of 316-series austenitic stainless steels, and further establishes the mechanistic framework by which these alloying elements regulate the solidification mode of weld metal in 316L(N)/CLF-1 dissimilar steel joints. The principal conclusions are as follows:

(1) Differences in Ni, Mn, and N contents significantly affect the ferrite fraction in 316L-series stainless steels. The higher Ni-Mn-N level in 316LN stainless steel stabilizes the austenitic matrix, resulting in an almost single-phase austenitic microstructure. In contrast, the lower Ni-Mn-N level in 316L stainless steel promotes the retention of a small amount of lath-like ferrite. Consistent with these microstructural differences, 316LN stainless steel, 316L stainless steel exhibits a stronger magnetic response and a higher overall corrosion rate than 316LN stainless steel.

(2) The equilibrium phase diagrams and non-equilibrium solidification paths demonstrate that Ni, Mn, and N strongly influence the solidification behavior of the 316L(N)/CLF-1 weld metal. The high-Ni–Mn–N weld enters the $L + \delta + \gamma$ three-phase region at approximately 1451 °C and a solid fraction of ~0.50, whereas the low-Ni–Mn–N weld reaches this region at approximately 1445 °C and ~0.80, indicating earlier and more extensive participation of the peritectic reaction in the former. Meanwhile, the calculated M_s decreases by approximately 30 °C with increasing Ni–Mn–N content, reflecting enhanced austenite stability and suppression of the subsequent $\gamma \rightarrow$ martensite transformation. Consequently, the room-temperature microstructure changes from ferrite + martensite to columnar austenite + interdendritic lath martensite.

(3) Integrating EBSD, magnetic, and electrochemical characterizations with equilibrium phase diagram and solidification path calculations established a mapping of composition, microstructure, and properties among Ni, Mn, and N contents. This mapping provides a quantitative basis for alloy-composition optimization and welding design of 316L(N)/CLF-1 dissimilar steel welded joints under nuclear fusion service conditions.

CRedit authorship contribution statement

Hangbiao Mi: Writing – original draft, Investigation, Data curation, Formal analysis.

Lingyun Feng: Writing – review & editing, Methodology, Formal analysis. **Ruiwen Dai:**

Investigation, Data curation, Formal analysis. **Guoqing Xin:** Data curation. **Binyan He:**

Writing – review & editing, Methodology, Conceptualization, Data curation. **Wang**

Jincheng: Writing – review & editing, Methodology. **Wei Guo:** Writing – review & editing,

Supervision, Funding acquisition, 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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