

Optimizing δ -ferrite structure to enhance high-temperature elongation in ER308L stainless steel deposited metal

Research Article

Cong Jiang*, Yanchang Qi, ZiXin Xu, Guangchang Yang, Chengyong Ma

CISRI Welding Institute, Central Iron and Steel Research Institute, Beijing 10081, China

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Abstract: As the size and weight of crucial equipment in third-generation nuclear power plants increase, the demand for high-temperature plasticity in welds has become more critical. This study examines the effect of varying δ -ferrite content on the high-temperature tensile properties of 308L austenitic stainless steel deposited metal at 350°C. The results reveal that as the δ -ferrite content decreased from 11.6 to 7.4%, the ferrite morphology shifted from continuous lathy and network structures to a discontinuous skeletal form. Correspondingly, 350°C high-temperature elongation increased from 26 to 32%, while tensile strength remained stable across all specimens, exhibiting good strength and plasticity at room temperature. The discontinuous skeletal δ -ferrite (7.4%) suppresses interfacial crack propagation during high-temperature tensile testing, thereby improving the high-temperature ductility.

Keywords: High-temperature elongation • 308L stainless steel • microstructure • welding

Nomenclature

ASS	austenitic stainless steel
Creq	chromium equivalent
EDS	energy dispersive spectrometer
F	ferrite
FA	ferritic-austenitic
ICP-AES	inductively coupled plasma atomic emission spectroscopy
L	liquid phase
Nieq	nickel equivalent
SEM	scanning electron microscope
UTS	ultimate tensile strength
YS	yield strength

1. Introduction

Austenitic stainless steel (ASS) are critical materials in nuclear reactors and power stations due to their excellent high-

temperature mechanical properties, corrosion resistance, and weldability [1,2]. Among these, 308L stainless steel is commonly used as a corrosion-resistant overlay for low-alloy steel shells and dissimilar metal joints in equipment such as nuclear reactor pressure vessels [3,4]. These welded joints near the reactor core must withstand temperatures up to 350°C while operating under high pressure and complex cyclic stresses [5,6]. Typically, the elongation of deposited metal in nuclear applications ranges between 24 and 26% at 350°C.

With the development of third-generation nuclear technology, nuclear units' capacity and service life are being extended [7,8], resulting in larger and heavier reactor pressure vessels. That, in turn, raises the demand for improved high-temperature plasticity of welds at 350°C, as it directly affects the long-term stability and crack resistance of welded joints.

The microstructure of 308L stainless steel consists of an austenitic matrix with δ -ferrite [9]. The δ -ferrite content in stainless steel weldments is typically controlled between 5 and 12% for nuclear power applications. Excessive ferrite content can form brittle σ -phases [10,11], while too little ferrite increases the risk of solidification cracking [12]. Numerous studies have shown that the amount and

* E-mail: jc13576695162@163.com

morphology of δ -ferrite significantly affect the room-temperature mechanical properties of 308L welds. For instance, post-weld heat treatment can reduce the ferrite content in ASS welded joints, enhancing their mechanical properties [13]. Adjustments to the welding process also allow for control over δ -ferrite content [14,15]. Layer-by-layer control of welding parameters during laser additive manufacturing could result in a stepped ferrite content distribution, improving room temperature strength compared to uniform parameter settings [16]. However, shallow ferrite content can be detrimental. The 308L welded joints with a ferrite content of 6% exhibited better room-temperature plasticity than those with only 1% [17].

Another approach to improving mechanical properties is by refining the ferrite morphology. Adding Mg and Ti could refine ferrite in the Ferrite (F) solidification mode, though the effect was less pronounced in the Ferritic-austenitic (FA) solidification mode [18]. The addition of Ti led to the formation of hard and brittle phases, negatively impacting the mechanical properties at room temperature.

As shown in Figure 1, current research on the mechanical properties of deposited metal primarily focuses on its behavior at room temperature. However, studies that explore the enhancement of high-temperature plasticity

at 350°C in 308L stainless steel deposited metal remain limited. By tailoring δ -ferrite morphology from continuous networks to a discontinuous skeletal structure, this study achieves a 32% elongation at 350°C – a 23% improvement over conventional microstructures – addressing a critical gap in the design of nuclear-grade ER308L welds.

2. Materials and methods

In this study, ER308L stainless steel welding wires with a diameter of 1.2 mm were utilized. The base material was commercial 304L stainless steel, measuring 500 mm × 130 mm × 20 mm with a 10° bevel. Three pairs of weld metal test plates were prepared using tungsten inert gas welding with a Miller Dynasty 700 welding machine. The chemical compositions of ER308L wire and base metal are shown in Table 1.

Welding parameters are shown in Table 2. Three pairs of deposited metal weld test plates with different ferrite contents were obtained by adjusting welding heat input and shielding gas composition. Prior to welding, the surface of the base material was polished using a steel wire brush, followed by cleaning with acetone to eliminate any oil residues.

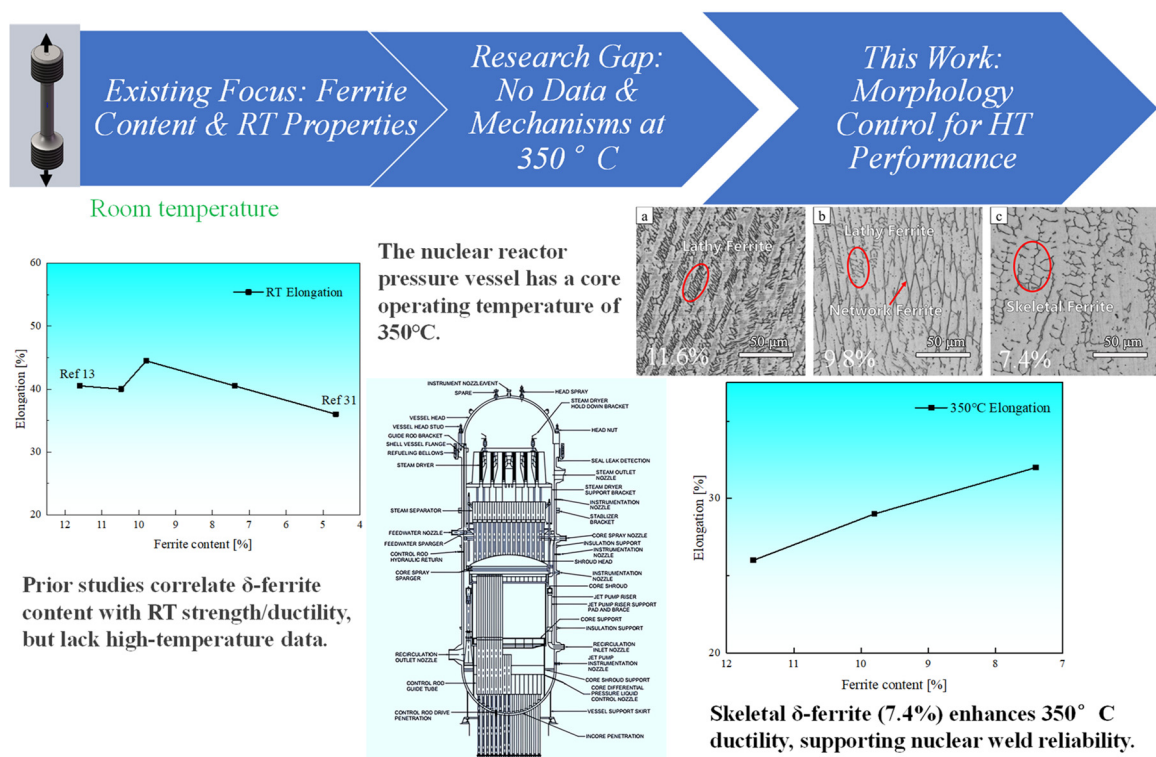


Figure 1. From ferrite content to morphology control: A 350°C-ready pathway for nuclear-grade ER308L weld.

Material	Elemental content (wt%)						
	C	Si	Mn	S	P	Cr	Ni
304L	0.052	0.48	1.08	<0.002	0.0390	18.23	8.39
ER308L-1	0.016	0.36	1.50	<0.002	0.0170	19.57	9.91
ER308L-2	0.016	0.14	1.98	0.0007	0.0140	19.90	9.78
ER308L-3	0.021	0.22	1.69	0.0120	0.0022	19.71	9.90

Table 1. Chemical compositions of ER308L wire and base metal.

Test plate	Current (A)	Voltage (V)	Welding speed (cm/min)	Heat input (kJ/cm)	Shield gas	Gas flow rate (L/min)
No. 1	180–220	10–16	9	15	100% Ar	10
No. 2	180–220	10–16	17	11	100% Ar	10
No. 3	180–220	16–24	24	11	97% Ar + 3% N ₂	10

Table 2. Welding parameters.

Figure 2(a) shows a schematic of the sampling locations. The weld metal was sectioned transversely to prepare metallographic specimens, and tensile specimens were cut along the weld axis for fracture analysis. Figure 2(b) shows a schematic

of the multilayer welding sequence, which can control welding deformation and residual stress. Metallographic specimens were polished with sandpaper and mechanically finished, followed by etching in a CuCl + HCl + H₂O solution (5:15:80) for

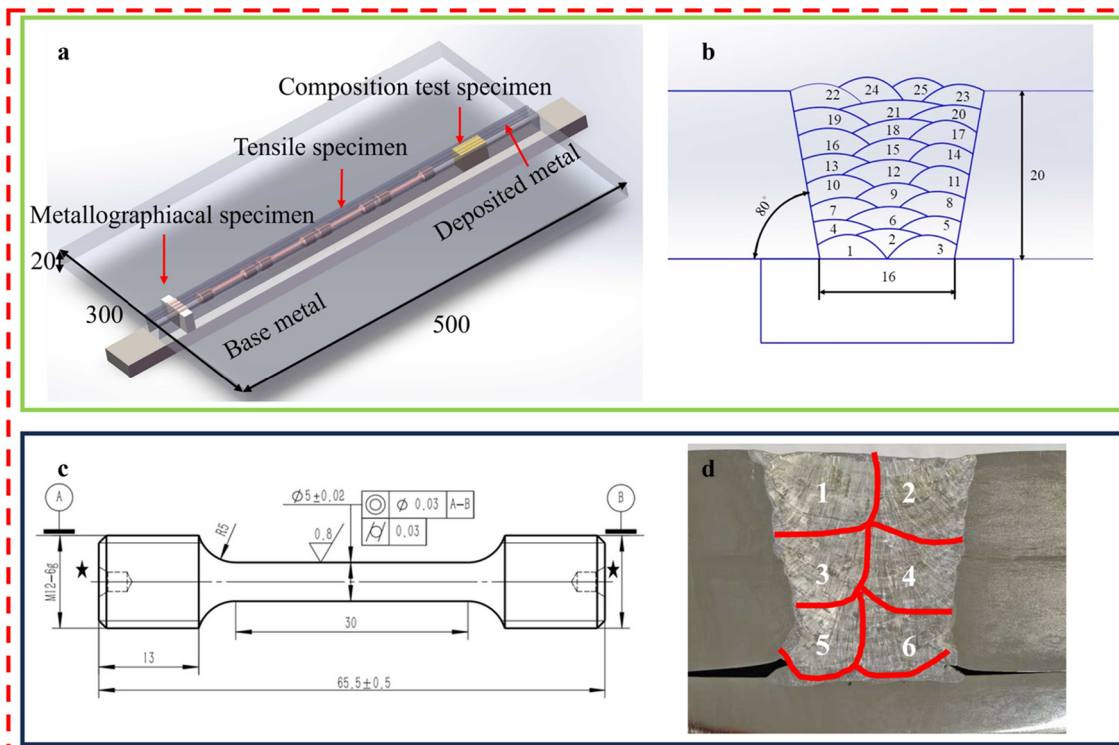


Figure 2. Schematic diagram of experimental material preparation: (a) Dimensions and specimen locations of the weldments. (b) Schematic representation of the deposition process. (c) Tensile specimen machining drawing. (d) Schematic diagram of the testing area of the FERITSCOPE FMP30 ferrite tester.

30 s. Microstructural analysis of the deposited metals was performed using a Leica MEF4 M optical microscope (OM), an FEI Quanta 650 scanning electron microscope (SEM), and an Oxford X-max 50 energy dispersive spectrometer (EDS) for elemental distribution analysis.

As shown in Figure 2(c), standard tensile specimens with a 5 mm diameter were prepared, with a gauge length of 25 mm, and tensile tests were conducted at room temperature and 350°C. Room temperature tests were performed using an ETM205D electronic tensile testing machine. High-temperature tensile tests at 350°C were conducted using an Instron 3369 tensile machine. At 350°C, the tensile strain rate was maintained at 0.0007 s⁻¹ prior to yielding, and the crosshead displacement rate was adjusted to 3 mm/min post-yielding. At room temperature, the tensile strain rate was maintained at 0.00025 s⁻¹ prior to yielding, and the crosshead displacement rate was adjusted to 0.0067 s⁻¹ post-yielding. During tensile tests, force, crosshead displacement, and extensometer strain were recorded in real-time. The force–displacement data were converted to engineering stress–strain curves using the initial specimen dimensions. The yield strength (YS) was determined by the 0.2% offset method, and the ultimate tensile strength (UTS) was calculated as the maximum stress before fracture. After the tensile test, the surface and cross-section of the fracture were investigated by SEM to determine the failure morphology.

The chemical compositions of the deposited metals were analyzed using inductively coupled plasma atomic emission spectroscopy with a measurement accuracy of ±0.01 wt% for major elements (Cr, Ni, Mn, Si) and ±0.001 wt% for C, S, Mo, and N, and ±0.001 wt% for P. The deposited metals' chromium and nickel equivalents (Cr_{eq}, Ni_{eq}) and ferrite contents were predicted based on the Delong diagram. The formulae for calculating Cr_{eq} and Ni_{eq} according to the Delong diagrams are as follows: Cr_{eq} = Cr% + Mo% + (1.5 × Si%) + (0.5 × Nb%) (±0.01 wt%), Ni_{eq} = Ni% + (30 × C%) + (30 × N%) + (0.5 × Mn%) (±0.01 wt%) [19].

The FERITSCOPE FMP30 ferrite tester was calibrated, with a measurement uncertainty of ±1.5% for ferrite content. As shown in Figure 2(d), six random measurements were taken in the direction of the cladding weld channel, and the average of five readings at each point was taken as the measurement result.

3. Results

3.1 Chemical composition and microstructure of deposited metal

The chemical compositions and calculated values of Ni_{eq} and Cr_{eq} for the three deposited metals are summarized in Table 3.

Table 4 shows the results of the ferrite content of the deposited metal Nos 1, 2, and 3. The results of the magnetic method are slightly larger than those of the Delong diagram. Notably, deposited metal No. 3 displays the highest concentrations of Ni and N – crucial elements promoting austenite formation – while having the lowest ferrite content.

Figure 3 shows the microstructures of the three deposited metals. All three deposited metals consist of columnar dendrite austenite and ferrite. Figure 3(a) illustrates deposited metal No. 1, characterized by lathy ferrite morphology. Figure 3(b) depicts deposited metal No. 2, which features lathy and network ferrite morphology. However, the amount of lathy ferrite is significantly reduced compared to that of deposited metal No. 1. Figure 3(c) shows deposited metal No. 3, which exhibits a skeletal ferrite morphology.

Figure 4 presents the SEM and EDS results of δ-ferrite for the three deposited metals. SEM and OM reveal the complete mixed growth of dendritic structures and grain growth with significant orientation. This phenomenon occurs due to the tendency of δ-ferrite to preferentially grow along the direction of heat transfer during dendritic growth, coupled with the aggregation of ferrite-producing elements [11,20]. As shown in Figure 4(h), the δ-ferrite in deposited metal No. 3 exhibits the least continuity and appears in the form of a skeletal. In contrast, as illustrated in Figure 4(b) and (e), the δ-ferrite in both deposited metals No. 1 and No. 2 is characterized by a continuous network and lathy. The EDS mapping of Specimen No. 2 reveals a localized carbon enrichment (0.12 wt%) in interdendritic regions (Figure 4f). This anomaly is attributed to the precipitation of (Fe,Cr,Mn)₃C carbides at δ-ferrite/austenite interfaces during rapid solidification [5]. Based on the EDS results, it is evident that the δ-ferrite in deposited metal No. 3 has the highest chromium content and the lowest nickel content. Conversely, deposited metal No. 1 displays the opposite trend, while deposited metal No. 2 falls in the middle.

Material	C	Si	Mn	S	P	Cr
No. 1	0.011	0.37	1.53	0.021	0.0090	19.54
No. 2	0.012	0.16	1.92	0.014	0.0004	19.90
No. 3	0.011	0.26	1.56	0.008	0.0020	19.57
Material	Ni	Mo	N	Ni _{eq}	Cr _{eq}	Cr _{eq} /Ni _{eq}
No. 1	9.56	0.025	0.035	11.71	20.12	1.72
No. 2	9.60	0.034	0.041	12.15	20.17	1.66
No. 3	9.76	0.017	0.075	13.12	19.98	1.52

Table 3. Chemical composition of the molten metal (wt%), value of Cr_{eq}, Ni_{eq}.

Deposited metal	Ferrite content (%)								
	Based on the magnetic method							Based on the Delong plot	
	Site	1	2	3	4	5	Average	Average	Calculated value
No. 1	1	12.4	12.5	11.0	11.5	9.5	11.4	11.6 ± 1.0	10.2
	2	10.2	11.8	11.5	11.8	10.7	11.2		
	3	11.2	12.1	11.4	11.6	11.0	11.5		
	4	11.9	11.7	11.6	11.5	10.6	11.5		
	5	11.9	11.7	11.5	11.5	11.1	11.5		
	6	11.5	13.5	13.7	13.8	10.5	12.6		
No. 2	1	10.5	10.1	10.4	10.7	11.0	10.5	9.8 ± 0.7	9.4
	2	8.3	8.9	9.6	9.3	9.4	9.1		
	3	7.7	9.0	10.5	10.1	9.4	9.3		
	4	7.9	9.4	10.6	9.7	8.9	9.3		
	5	10.0	9.6	9.9	10.0	9.4	9.8		
	6	10.4	10.5	10.1	11.0	10.4	10.5		
No. 3	1	6.8	7.0	6.5	7.7	7.0	7.0	7.4 ± 0.6	6.2
	2	7.1	6.7	7.0	6.7	7.0	6.9		
	3	6.9	7.3	7.0	7.4	7.5	7.2		
	4	6.7	7.4	7.2	8.2	8.8	7.7		
	5	7.6	8.1	7.9	7.8	8.4	8.0		
	6	7.5	7.3	7.3	7.9	8.3	7.7		

Table 4. Ferrite content of deposited metal.

3.2 Mechanical properties and high-temperature tensile fracture morphology of deposited metal

Table 5 presents the room-temperature and high-temperature tensile test results of the deposited metals. The room-temperature tensile strengths of the three deposited metals all exceed 550 MPa, and the room-temperature

elongations are all greater than 40%. The high-temperature tensile strengths of the three deposited metals at 350°C do not vary significantly. Deposited metal No. 3, with a significant increase in nitrogen content and a more pronounced solid solution strengthening effect [15,21], has a higher high-temperature tensile strength than deposited metal No. 2, which has a higher ferrite content. The strength and plasticity of ASS decrease at

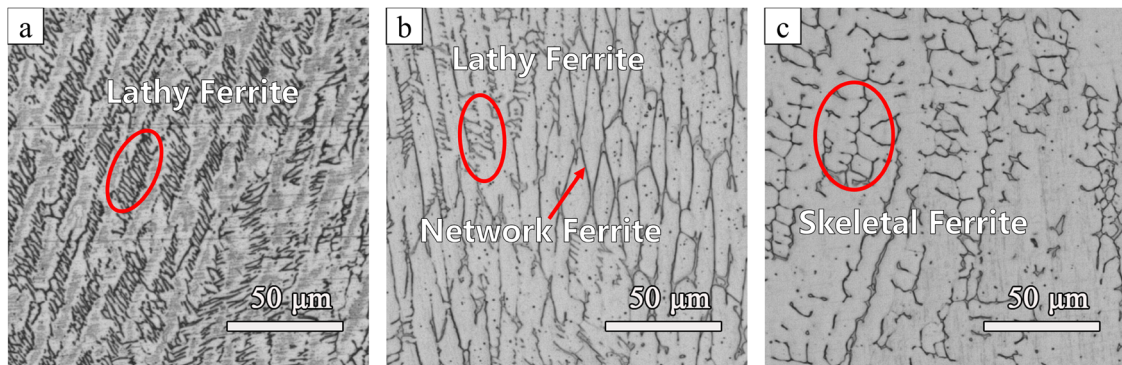


Figure 3. Metallographic microstructures of the deposited metals: (a) No. 1, (b) No. 2, and (c) No. 3.

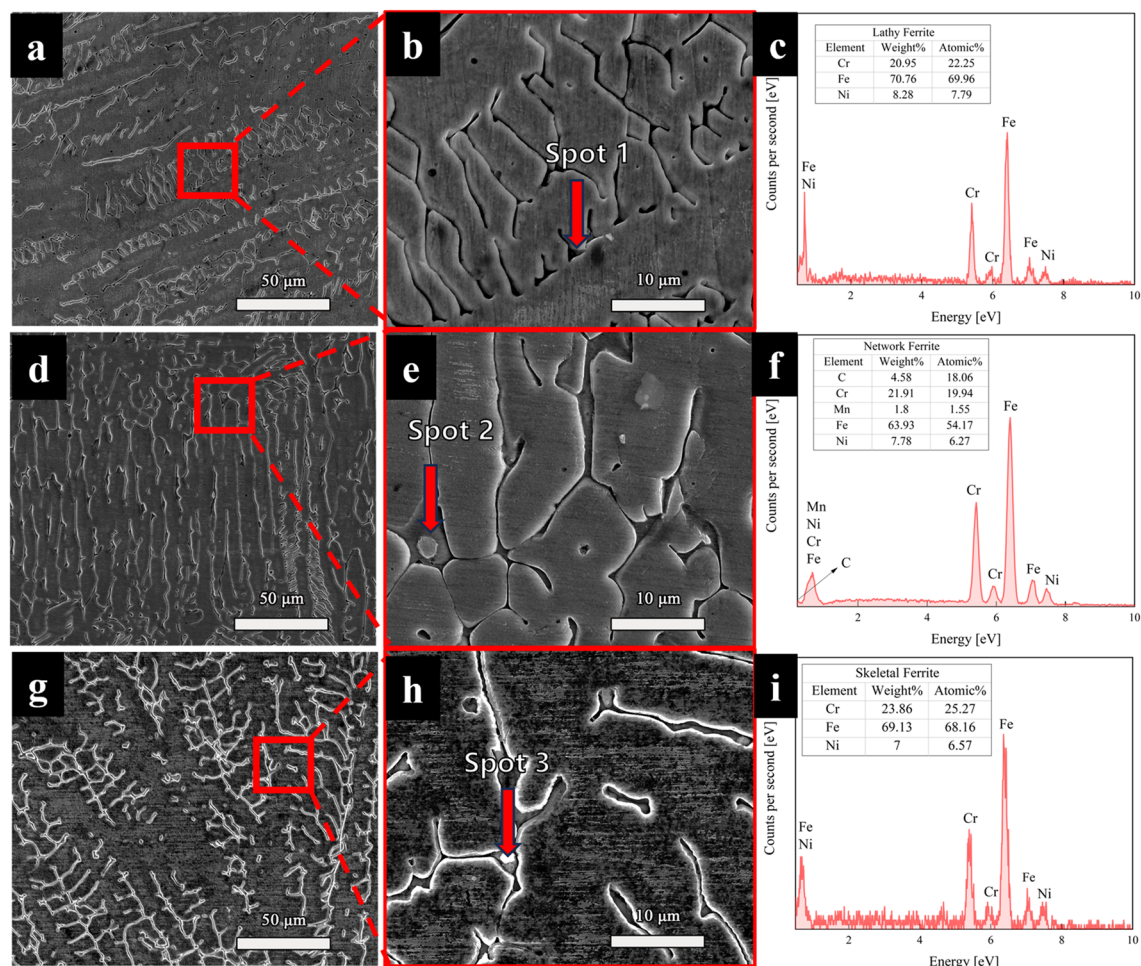


Figure 4. SEM morphology and EDS analysis of δ -ferrite in deposited metals: (a–c) No. 1, (d–f) No. 2, and (g–i) No. 3.

elevated temperatures [22,23]. The high-temperature YSs of deposited metals No. 1 and No. 2 are 339 and 322 MPa, respectively. The YS of deposited metal No. 3 is 284 MPa, a notable decrease compared to the other two. This reduction is attributed to the lack of lathy ferrite in its microstructure and fewer δ/γ phase interfaces, leading to lower resistance to

dislocation slip and yielding at lower stresses. The high-temperature elongations of the three deposited metals increase in order, with the lowest elongation of 26% for No. 1, 29% for No. 2, and the highest elongation of 32% for No. 3.

Figure 5 shows the room-temperature and high-temperature tensile stress–strain curves of the three groups of

Temperature (°C)	Specimen	YS (MPa)	UTS (MPa)	Elongation (%)
23	No. 1	427	570	40.5
	No. 2	421	557	44.5
	No. 3	403	561	40.5
350	No. 1	339	379	26
	No. 2	322	359	29
	No. 3	284	371	32

Table 5. Room temperature and high-temperature tensile properties of the deposited metal.

deposited metals. In the stress–strain curves at both room and high temperatures, all three groups of deposited metals exhibit a typical elastic stage, no obvious yield plateau, and a significant plastic deformation stage. As shown in Figure 4(a), in the room-temperature tensile stress–strain curve, the slope of the elastic stage of No. 2 is greater than that of No. 1 and No. 3, indicating the highest elastic modulus. With the decrease in ferrite content, the strain of the deposited metals first increases and then decreases, with the main differences occurring in the uniform plastic deformation stage. The slope of the uniform plastic deformation stage of No. 2 deposited metal is the largest, showing the best resistance to deformation, due to the strong hindrance of dislocation movement by the network and lathy δ -ferrite. In the high-temperature tensile stress–strain curves, the slope of the elastic stage of No. 3 is greater than that of No. 1 and No. 2, indicating the highest elastic modulus. With the decrease in ferrite content, the strain of the deposited metals gradually increases. No. 3 yields at a lower strength, with the largest slope in the uniform plastic deformation stage, showing the best resistance to deformation. The strength and plasticity of the three deposited metals are satisfactory at room temperature during tensile testing. However, at 350°C, there is a notable decrease in strength and plasticity, while elongation increases significantly with decreasing ferrite content. Therefore, high-temperature tensile fracture morphology was further investigated.

Figure 6 illustrates the SEM results of high-temperature tensile fractures for the three deposited metals. Figure 6(a), (d), and (g) corresponds to the tensile fractures of No. 1, No. 2, and No. 3, respectively, revealing ellipsoidal

macroscopic fracture morphology with distinct necking deformation. Dense, challenging nests can be observed in Figure 6(c), (f), and (i), confirming that all three deposited metals exhibit plastic fracture modes. Significant undulations in the fracture surface are evident in Figure 6(b), (e), and (h), with Figure 6(h) showing a distinct pit, indicating that the plastic deformation of deposited metal No. 3 is greater than that of No. 1 and No. 2.

4. Discussion

4.1 Effect of Cr_{eq}/Ni_{eq} on δ -ferrite

There are four possible modes of solidification and solid-state phase transitions in ASS weld metal: A, AF, FA, and F. Based on the calculated Ni_{eq} and Cr_{eq} values presented in Table 1, it can be concluded that the equilibrium solidification of the weld metal follows the FA mode, which can be divided into four solidification stages [18,24]. In Stage I, the δ -ferrite phase precipitates directly from the liquid phase (L). In Stage II, the quantity of δ -phase gradually increases and grows, while the L diminishes. A peritectic reaction ($L = \delta + \gamma$) occurs at the interface between the L and the δ -phase, leading to the formation of the γ -phase at the δ -phase boundary, accompanied by significant volume contraction around the δ -ferrite. In Stage III, the liquid phase completely disappears, concluding the peritectic reaction, and the δ -phase begins to transform into the γ -phase. Due to the rapid cooling rate and the nonequilibrium solidification during the welding process, the phase

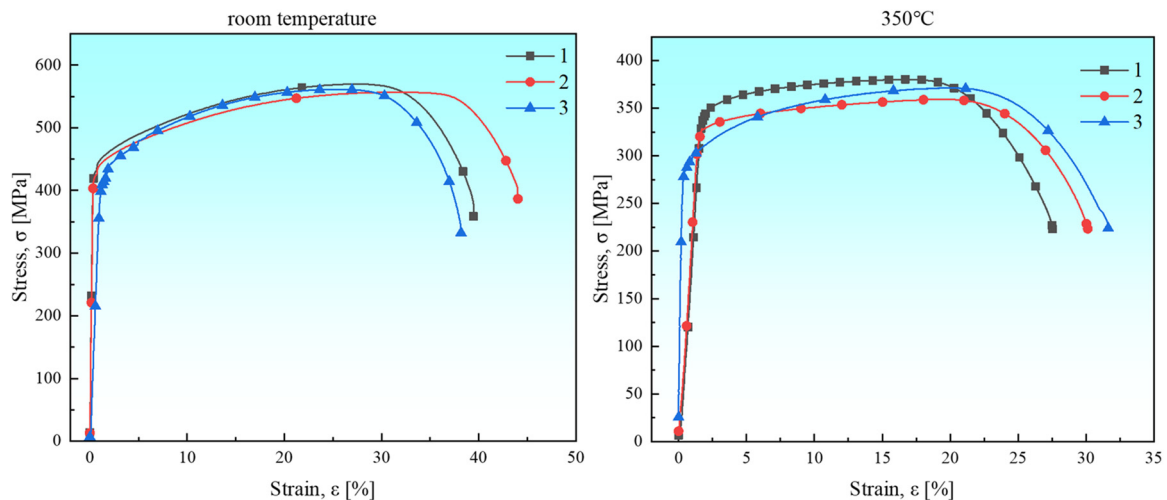


Figure 5. Stress–strain curves of deposited metals: (a) room temperature and (b) 350°C.

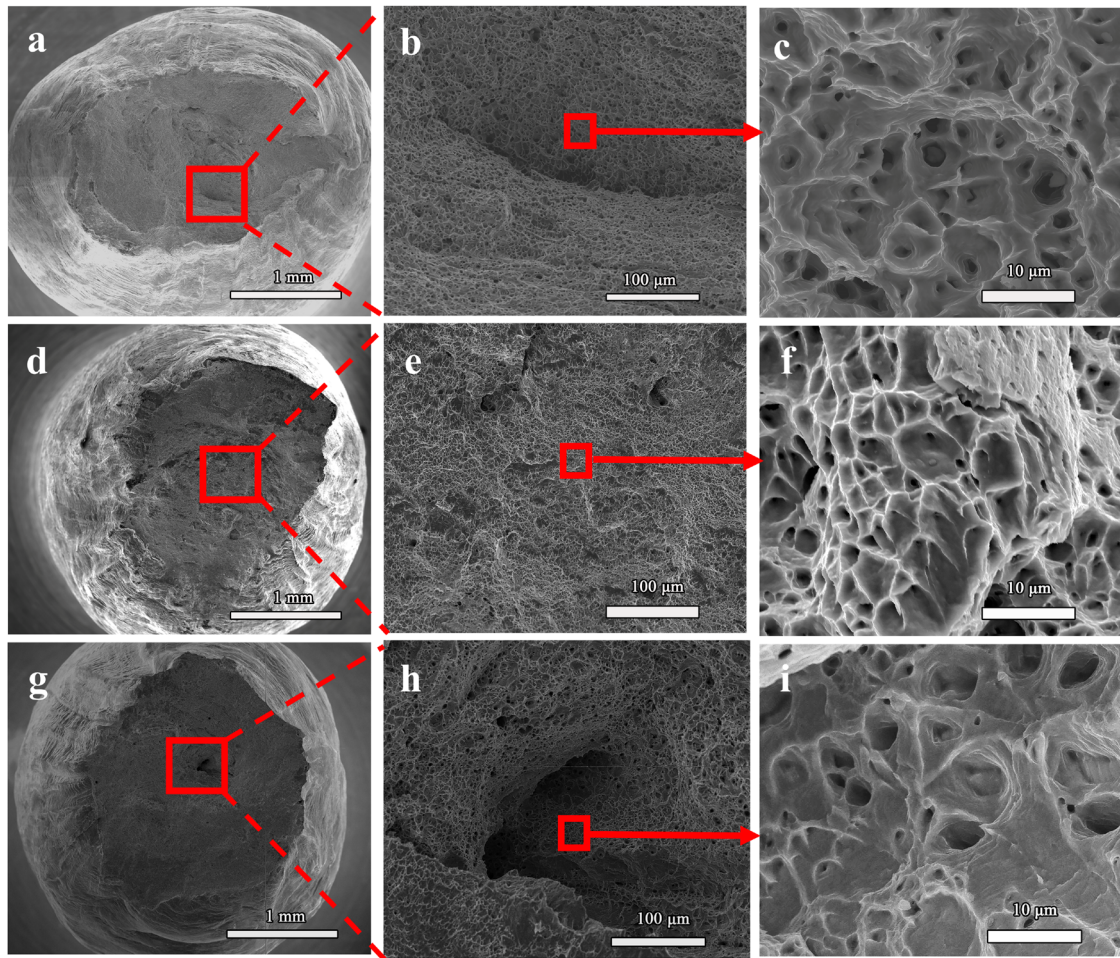


Figure 6. High-temperature tensile fracture morphology of the deposited metal: (a–c) No. 1, (d–f) No. 2, and (g–i) No. 3.

transition in Stage III is not fully completed, resulting in a residual amount of δ -ferrite in the microstructure [25].

The ferrite in deposited metal No. 3 exhibits a skeletal morphology due to its relatively low Cr_{eq}/Ni_{eq} ratio [26]. Near the Cr_{eq}/Ni_{eq} ratio interval of AF, austenite forms by consuming ferrite. As the phase transformation progresses, ferrite-producing elements such as chromium become increasingly concentrated in the residual ferrite, while austenite-producing elements, including nickel, carbon, and nitrogen, are depleted. Consequently, deposited metals No. 1 and No. 2 possess relatively high Cr_{eq}/Ni_{eq} ratios, leading to the formation of lathy ferrite in their room temperature microstructures. This phenomenon is attributed to the limited diffusion during the ferrite–austenite phase transformation; as the diffusion distance decreases, the transformation occurs more efficiently in closely spaced laminae. That enables the residual ferrite to adopt a lathy morphology that intersects the growth

direction of the original dendritic or cellular crystals [17], thereby replacing the skeletal ferrite.

4.2 High-temperature plasticity enhancement mechanism

Figure 7 presents the SEM results of the high-temperature tensile fracture profiles. As illustrated in Figure 7(a), (d), and (g), the fracture widths of deposited metals No. 1 and No. 2 are comparable. In contrast, the diameter of the fracture for No. 3 is significantly smaller, indicating better plasticity of No. 3. Figure 7(b), (e), and (h) depicts the fracture paths of the tensile fractures for the three deposited metals, revealing that the fracture paths of metals No. 2 and No. 3 are more tortuous compared to that of No. 1. As shown in Figure 7(c) and (f), elongated ferrite strips are present near the tensile fractures of No. 1 and No. 2, with fractured ferrite strip observed in the

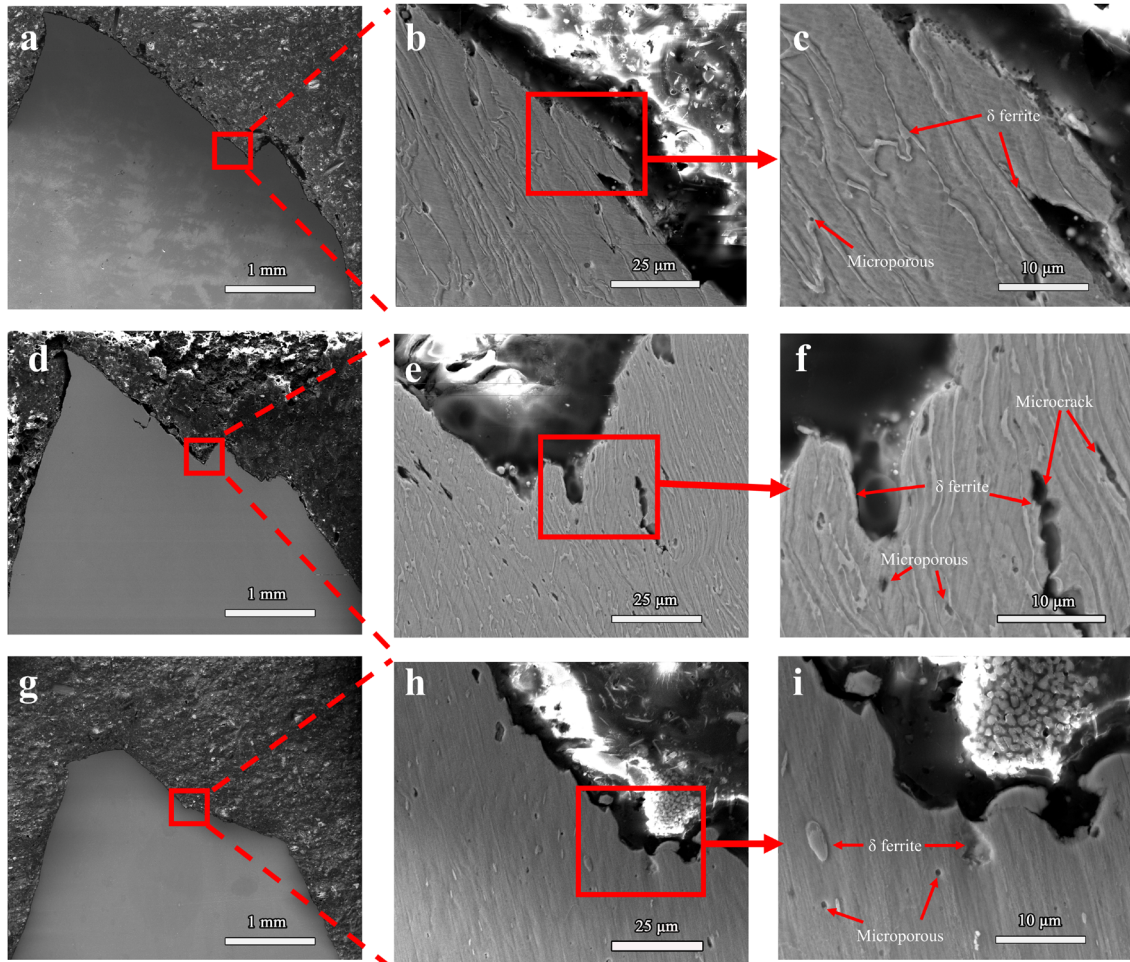


Figure 7. SEM results of high-temperature stretching fracture profile: (a–c) No. 1, (d–f) No. 2, and (g–i) No. 3.

fracture path of No. 1. It can be demonstrated that the crack extends along the austenite– δ ferrite interface [27]. The ferrite in the vicinity of the tensile fracture for deposited metal No. 3 is ellipsoidal, as shown in Figure 7(i). The fracture paths in Figure 7(c) and (f) are characterized by intergranular fracture, while the fracture path in Figure 7(i) exhibits trans-granular fracture. Micropores and microcracks are evident at the internal δ and γ phase interfaces near the fracture edges in all three deposited metals. The austenite–ferrite interface is the primary site for microcrack initiation at 350°C [28].

Moreover, it has been shown that in the fatigue fracture of ASS, cracks tend to expand rapidly along reticulated or flocculated ferrite [29,30]. At the same time, discontinuous rod ferrite can mitigate the rapid crossover of cracks, enhancing overall performance [13,31]. Consequently, in deposited metals with 9–12% ferrite content, the δ -ferrite exhibits a continuous structure, including lathy and network ferrite,

providing a rapid pathway for expanding micropores, resulting in limited plastic deformation during the overall tensile process. In contrast, in metals with 6–7% ferrite content, the δ -ferrite exhibits a discontinuous skeletal structure, necessitating microcrack penetration through the austenite grains for further extension. During this plastic deformation, the austenite grains absorb some of the crack expansion energy. In summary, there is a negative correlation between ferrite content and high-temperature plasticity in 308L deposited metal; as ferrite content decreases, the high-temperature plasticity of ER308L deposited metal increases.

5. Conclusions

This study investigates the optimization of δ -ferrite morphology to enhance the high-temperature elongation of

ER308L deposited metal for nuclear applications. Microstructural characterization reveals that reducing δ -ferrite content from 11.6 to 7.4% transforms the morphology from continuous lathy and network structures to a discontinuous skeletal form. This transition correlates with an increase in high-temperature elongation from 26 to 32%, while tensile strength remains stable. High-temperature tensile tests identify the austenite-ferrite interface as the primary site for crack initiation. Reducing ferrite content to 6–7% disrupts continuous δ -ferrite structures, thereby enhancing resistance to crack extension and improving high-temperature plasticity.

Future efforts should focus on evaluating the long-term thermal stability of skeletal δ -ferrite under reactor-relevant conditions and integrating multi-scale models to predict its evolution during multi-pass welding. In engineering practice, this optimized microstructure can mitigate crack risks in reactor internals and primary coolant piping under thermal transients, aligning with Generation III + nuclear system demands.

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Author contributions

Cong jiang: methodology, investigation, writing-original draft. Yanchang Qi: methodology, investigation, writing-review and editing. ZiXin Xu: investigation. Guangchang Yang: methodology. Chengyong Ma: Review and editing.

Conflict of interest statement

Authors state no conflict of interest.

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