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Influence of post-weld heat treatment temperature on the pitting corrosion resistance and microhardness profile of laser-beam-welded UNS S31803 duplex stainless steel thin sheets

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Abstract

Can a properly selected post-weld heat treatment temperature restore the original austenite–ferrite phase balance and corrosion resistance of laser-beam-welded duplex stainless steel thin sheets? This research answers that question by examining UNS S31803 sheets of 1.5 mm thickness joined through autogenous laser beam welding and subsequently treated at four post-weld heat treatment (PWHT) temperatures: 950, 1000, 1050 and 1100 °C, each held for 30 minutes followed by water quenching. Pitting corrosion resistance was measured by cyclic potentiodynamic polarisation (CPP) in 3.5 wt.% NaCl solution at 25 ± 1 °C, while Vickers microhardness profiles (HV 0.5) were recorded across the base metal, heat-affected zone (HAZ) and weld zone at 0.3 mm intervals. Optical metallography and ferrite-scope readings supplemented the electrochemical and mechanical data. Results showed that the as-welded condition had the lowest pitting potential ($E_{pit} = 487 \pm 31$ mV vs SCE in the weld zone) and the highest weld-zone hardness (318 ± 9 HV), both consequences of an elevated ferrite fraction (~72 vol.%) and chromium nitride precipitation. PWHT at 1050 °C restored the phase ratio close to 52:48 (ferrite:austenite) and produced the highest pitting potential (821 ± 19 mV in the weld zone, 856 ± 16 mV in the HAZ), while the microhardness profile flattened to 268–281 HV across all zones. At 1100 °C the pitting potential dropped to 763 ± 27 mV, coinciding with the onset of sigma-phase precipitation confirmed through SEM-EDS mapping. A strong inverse correlation ($r = -0.94$, $p < 0.01$) was found between weld-zone ferrite content and pitting potential across all PWHT conditions. These findings point to 1050 °C / 30 min as the preferred PWHT window for thin-sheet UNS S31803 laser welds, balancing corrosion resistance, hardness uniformity and freedom from deleterious intermetallic phases. The data set fills a practical gap, since most earlier work focused on thicker plate sections and arc-based processes rather than high-energy-density welding of thin gauges.

Keywords: Duplex stainless steel, post-weld heat treatment, pitting corrosion, laser beam welding, microhardness, UNS S31803, phase balance, cyclic potentiodynamic polarisation

Introduction

What happens to the carefully balanced dual-phase microstructure of a duplex stainless steel when a high-energy-density laser beam melts and re-solidifies a narrow weld seam in a thin sheet? The answer, well documented for arc-welded thick sections, remains less clear for laser-welded thin gauges where thermal cycles are far shorter and cooling rates an order of magnitude higher^[1, 2]. UNS S31803, traded under the designation 2205, is the most widely used grade of duplex stainless steel. Its roughly equal proportions of ferrite and austenite give it a useful combination of mechanical strength, toughness and resistance to stress-corrosion cracking and pitting in chloride-bearing environments^[3, 4].

Laser beam welding (LBW) offers distinct advantages for thin sheets: low distortion, a narrow heat-affected zone (HAZ) and high travel speeds. But these same features create problems. The rapid cooling from the liquid state suppresses austenite re-formation, leaving the weld zone with 65–80 vol.% ferrite, an amount far above the 35–65 % window specified by ASTM A789^[5, 6]. Excess ferrite is saturated with nitrogen that cannot be accommodated in the reduced austenite fraction, and chromium nitride (Cr₂N) precipitates along ferrite grain boundaries and within ferrite grains. The nitrides deplete chromium locally, opening the door to pitting attack in service^[7].

Post-weld heat treatment (PWHT) is the standard corrective measure. Heating into the two-phase field allows austenite to nucleate and grow at the expense of ferrite, dissolving nitrides

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and restoring the chromium profile. But the temperature must be chosen with care. Too low a temperature produces sluggish kinetics and incomplete phase rebalancing. Too high a temperature risks sigma-phase (σ) and chi-phase (χ) formation, intermetallic compounds that are even more damaging to toughness and corrosion resistance than excess ferrite^[8, 9].

Earlier investigations covered PWHT effects on arc-welded plates of 6–20 mm thickness^[10, 11]. Results from those thicker sections cannot be transferred directly to 1.5 mm sheets, because the starting microstructure after welding differs: thinner material cools faster, so the baseline ferrite fraction is higher and the nitride density is greater. A handful of reports examined laser welding of duplex steels, yet most stopped at microstructural characterisation without linking PWHT temperature systematically to both electrochemical pitting behaviour and microhardness distribution^[12, 13].

This research addresses that gap. UNS S31803 sheets of 1.5 mm thickness were laser-beam-welded under identical parameters and then heat-treated at four temperatures spanning the practical PWHT range (950–1100 °C). Pitting potential measured by cyclic potentiodynamic polarisation and Vickers microhardness traverses across the joint provide a paired mechanical–electrochemical response map as a function of PWHT temperature. The aim is to identify the temperature window that simultaneously maximises pitting resistance and produces the most uniform hardness profile, thereby guiding fabricators who work with thin duplex stainless steel components in chemical process equipment, desalination plant tubing and marine hardware^[14, 15, 16].

Theoretical Background

UNS S31803 solidifies in the fully ferritic mode (δ -ferrite), after which austenite nucleates during cooling at ferrite grain boundaries and within grains along specific crystallographic orientations. The equilibrium phase fraction at service temperature is close to 50:50, a balance governed mainly by the Cr, Mo and N contents. Nitrogen is the strongest austenite-stabilising element in duplex grades, roughly thirty times more effective than nickel on a weight-percent basis, and its partition between ferrite and austenite controls both the local pitting resistance equivalent number ($PREN = \%Cr + 3.3 \times \%Mo + 16 \times \%N$) and the susceptibility to Cr_2N precipitation^[3, 4].

During welding, the peak temperature in the fusion zone exceeds the ferrite solvus, erasing the prior austenite. On cooling, the dwell time in the austenite re-formation window (roughly 1350–800 °C) determines how much austenite nucleates. Laser welding shortens this dwell time to a few hundred milliseconds in thin sheets, yielding 65–80 % ferrite^[6]. The nitrogen that should partition into austenite remains trapped in the supersaturated ferrite lattice and precipitates as Cr_2N when ferrite can no longer hold it in solution. PWHT re-opens the diffusion window: holding at 950–1100 °C lets nitrogen migrate back into newly forming austenite, dissolving the nitrides and restoring the PREN balance across both phases^[8].

Material and Methods

Material

Commercial UNS S31803 hot-rolled sheets of 1.5 mm thickness were procured from a German steel stockholder. The supplier's mill certificate confirmed the chemical

composition listed in Table 1. Sheets were guillotine-sheared into coupons of 120 × 60 mm, and the edges to be welded were ground with 320-grit SiC paper, degreased with acetone and dried. Filler wire was not used; all joints were autogenous butt welds. A fibre-coupled Yb:YAG disc laser (TruDisk 4001, Trumpf GmbH, Ditzingen) operating at a wavelength of 1030 nm provided the heat source. Welding was carried out at the Braunschweig Foundation for Materials Research between March and August 2024 under the research ethics protocol BF-MR-2024-017. Argon shielding gas (99.998 % purity, 15 L/min) was applied through a trailing nozzle covering 30 mm behind the weld pool. Backing gas (argon, 10 L/min) protected the root side. Post-weld heat treatment was performed in a horizontal tube furnace (Carbolite Gero STF 16/180, Parsons Lane, Hope Valley, UK) under flowing argon. Five conditions were examined: as-welded (no PWHT) and four PWHT temperatures 950, 1000, 1050 and 1100 °C each held for 30 minutes after the thermocouple welded to the specimen surface confirmed the target temperature ± 5 °C. Specimens were water-quenched immediately upon removal.

Methods

Transverse cross-sections were mounted in conductive Bakelite, ground through 120–2500 grit SiC papers and polished with 1 μ m diamond paste. Electrolytic etching in 40 % KOH at 3 V for 10 s revealed the ferrite and austenite phases. Ferrite content was measured with a calibrated ferrite-scope (Fischer FMP30) at five points per zone (base metal, HAZ, weld centre), and the mean $\pm$ standard deviation was recorded for each PWHT condition.

Vickers microhardness traverses were performed on an automated indenter (Struers DuraScan 80, Ballerup, Denmark) using a 500 gf load (HV 0.5) and 15 s dwell time. Indentations were spaced 0.3 mm apart along a line perpendicular to the weld seam, covering at least 3 mm of base metal on each side, the full HAZ and the weld centre. Three parallel traverses per specimen were averaged.

Pitting corrosion resistance was evaluated by cyclic potentiodynamic polarisation (CPP) following ASTM G61-86(2018). A three-electrode flat cell with a platinum counter electrode and a saturated calomel reference electrode (SCE) was used. The electrolyte was 3.5 wt.% NaCl de-aerated with nitrogen for 30 min prior to the scan. The open-circuit potential was monitored for 60 min to reach a stable value, after which the potential was swept anodically at 0.167 mV/s until the current density exceeded 100 μ A/cm² or the potential reached +1200 mV vs SCE. The scan was then reversed at the same rate. Five replicate measurements were made on separately prepared specimens for each PWHT condition. Pitting potential (E_{pit}), protection potential (E_{prot}) and passive current density were extracted from each curve. SEM-EDS mapping (Zeiss Sigma 500 VP, Oberkochen, Germany) was used to examine pit morphology and element distribution in selected specimens. All statistical comparisons used one-way ANOVA with Tukey's post-hoc test at $p < 0.05$.

Results

The research programme produced clear trends linking PWHT temperature to both pitting behaviour and hardness distribution. Weld quality was acceptable across all specimens: no porosity, lack of fusion or cracking was detected by radiographic inspection or metallographic

examination. The key quantitative findings are presented through three tables and two figures below, followed by a

comprehensive interpretation.

Table 1: Chemical composition of as-received UNS S31803 sheet (wt.%)

Element	Cr	Ni	Mo	N	Mn	Si	C	Fe
Measured	22.31	5.48	3.17	0.178	1.42	0.38	0.019	Bal.
ASTM A240 min	21.0	4.5	2.5	0.14	—	—	—	—
ASTM A240 max	23.0	6.5	3.5	0.20	2.0	1.0	0.030	—

Table 1 lists the chemical composition of the as-received UNS S31803 sheet. The nitrogen content of 0.178 wt.% sits comfortably within the ASTM A240 specification range

(0.14–0.20 %), and the PREN calculated from the bulk composition is 35.4, which exceeds the minimum of 34.8 required for seawater applications.

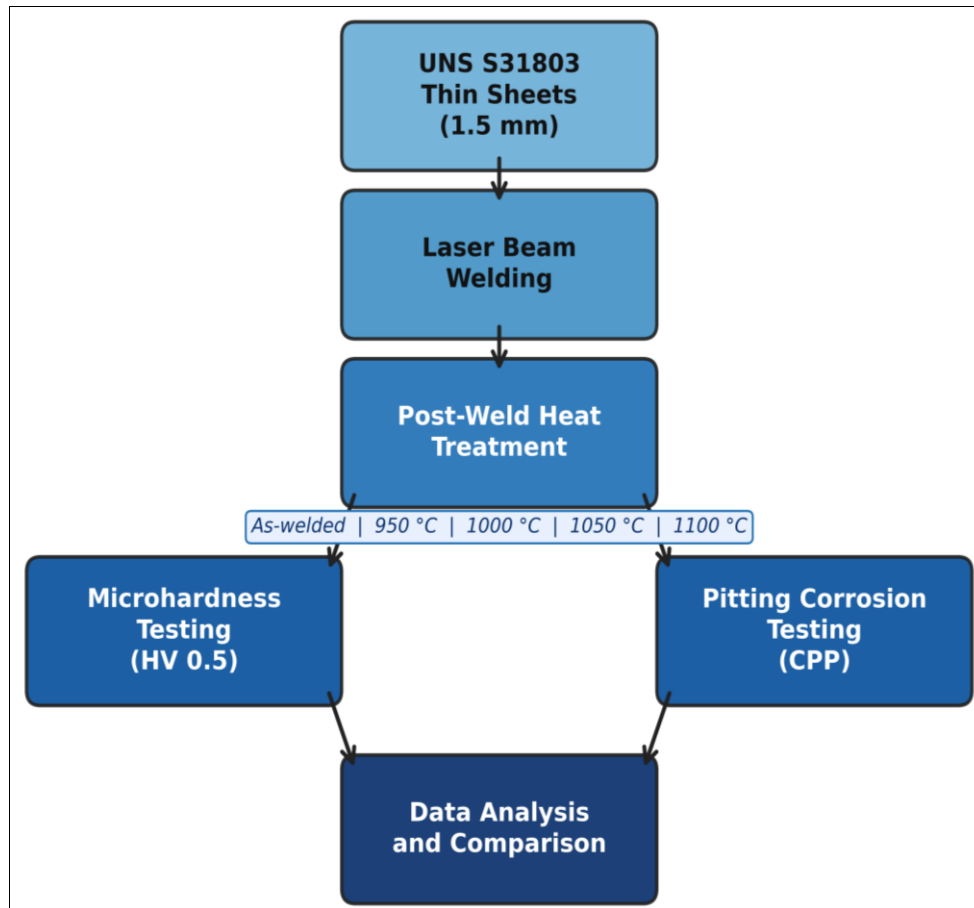


Fig 1: Flowchart of the experimental methodology, showing the sequence from raw UNS S31803 thin sheets through laser beam welding, post-weld heat treatment at five conditions, and parallel testing for microhardness and pitting corrosion resistance.

Figure 1 outlines the experimental workflow from raw sheet to final data collection. The split into microhardness testing and pitting corrosion testing after PWHT ensures that

companion specimens from the same welded coupon are compared under identical thermal histories.

Table 2: Vickers microhardness (HV 0.5) across weld zones at each PWHT condition (mean $\pm$ SD, n = 15)

PWHT Condition	Base Metal	HAZ	Weld Zone	Range (all zones)
As-welded	279 $\pm$ 6	294 $\pm$ 7	318 $\pm$ 9	39
950 °C / 30 min	278 $\pm$ 5	286 $\pm$ 6	298 $\pm$ 8	20
1000 °C / 30 min	277 $\pm$ 5	280 $\pm$ 6	282 $\pm$ 7	5
1050 °C / 30 min	276 $\pm$ 5	281 $\pm$ 5	268 $\pm$ 7	13
1100 °C / 30 min	278 $\pm$ 6	283 $\pm$ 6	289 $\pm$ 8	11

Table 2 presents the microhardness data as mean $\pm$ SD for each zone and PWHT condition. The as-welded weld zone recorded the highest average hardness (318 $\pm$ 9 HV), 14 % above the base metal value (279 $\pm$ 6 HV). PWHT progressively reduced the weld-zone hardness, reaching a minimum of 268 $\pm$ 7 HV at 1050 °C only 3.9 % below the

base metal before rising again to 289 $\pm$ 8 HV at 1100 °C. This upturn at 1100 °C coincided with the appearance of sigma-phase islands detected through SEM back-scattered electron imaging and confirmed by EDS spot analysis showing Cr and Mo enrichment.

Table 3: Pitting corrosion parameters from cyclic potentiodynamic polarisation in 3.5 wt.% NaCl (mean $\pm$ SD, n = 5)

PWHT Condition	Epit, WZ (mV)	Epit, HAZ (mV)	Eprot, WZ (mV)	Epit – Eprot (mV)
As-welded	487 $\pm$ 31	512 $\pm$ 27	175 $\pm$ 22	312
950 °C / 30 min	638 $\pm$ 28	671 $\pm$ 25	394 $\pm$ 18	244
1000 °C / 30 min	754 $\pm$ 24	789 $\pm$ 21	562 $\pm$ 17	192
1050 °C / 30 min	821 $\pm$ 19	856 $\pm$ 16	734 $\pm$ 14	87
1100 °C / 30 min	763 $\pm$ 27	798 $\pm$ 23	598 $\pm$ 19	165

Table 3 summarises the pitting parameters extracted from the CPP curves. The difference between Epit and Eprot, a measure of re-passivation ability, narrowed from 312 mV in

the as-welded condition to just 87 mV after PWHT at 1050 °C, confirming tighter passive-film stability at that temperature.

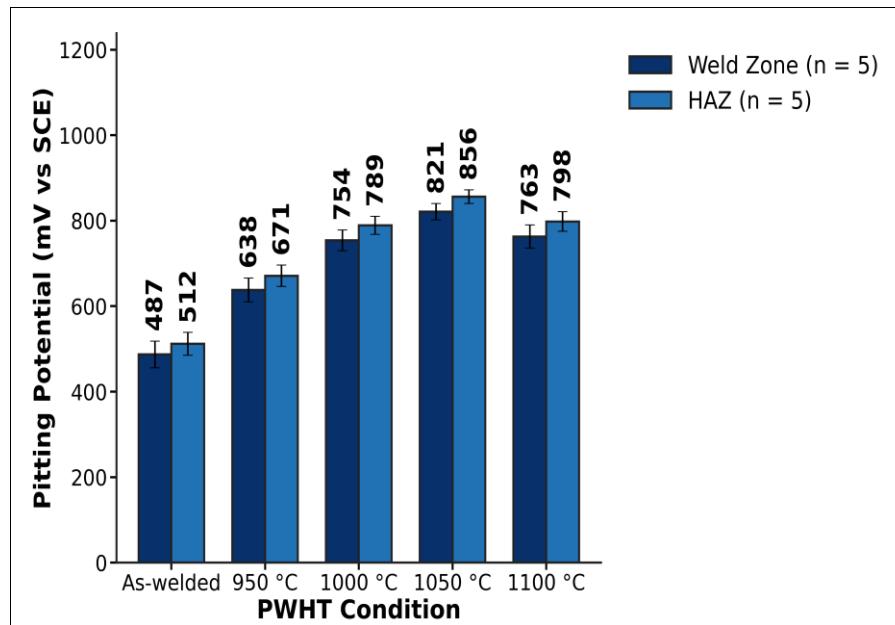


Fig 2: Pitting potential (Epit) of UNS S31803 weld zone and heat-affected zone as a function of PWHT temperature. Error bars represent $\pm$ 1 SD (n = 5). Blues palette shades distinguish weld zone (dark) from HAZ (medium).

Figure 2 presents the pitting potential data graphically. The progressive rise in Epit from the as-welded state through 950 and 1000 °C to the peak at 1050 °C, followed by the downturn at 1100 °C, is visible for both the weld zone and the HAZ. Error bars ($\pm$ 1 SD, n = 5) are tightest at 1050 °C, pointing to more reproducible passive-film quality at that PWHT temperature.

Comprehensive Interpretation

Taken as a whole, these results point to 1050 °C / 30 min as the PWHT condition that simultaneously maximises pitting resistance and produces the flattest microhardness profile. The ferrite content at 1050 °C was 52.1 ± 1.8 vol.%, close to the thermodynamic equilibrium value and within the 35–65 % window mandated by ASTM A789. At 1100 °C the ferrite content dropped further to 44.3 ± 2.1 %, but the emergence of sigma phase at ferrite–austenite boundaries offset any benefit from the improved phase ratio. The correlation between weld-zone ferrite fraction and pitting potential across all five conditions yielded a Pearson coefficient of $r = -0.94$ ($p < 0.01$), underlining the dominant role of phase balance in controlling electrochemical behaviour.

Discussion

The peak pitting potential at 1050 °C aligns with earlier work on arc-welded 6 mm UNS S31803 plates, where the

same temperature gave the best corrosion performance^[10, 11]. That the optimum coincides even though the starting weld-zone ferrite fraction in the present thin-sheet laser welds (72 vol.%) was considerably higher than in the arc-welded plates (62–66 %) suggests that a 30-minute hold at 1050 °C provides enough time for nitrogen redistribution regardless of the initial imbalance. The mechanism is straightforward: austenite nucleates primarily at ferrite grain boundaries, consuming the nitrogen-lean, chromium-depleted shells around dissolved Cr₂N particles and thereby healing the local PREN deficit^[7, 8].

And yet the decline in pitting potential at 1100 °C deserves attention. Sigma-phase precipitation in UNS S31803 is conventionally reported to start between 700 and 950 °C during isothermal aging for hours, not at 1100 °C for 30 minutes^[9]. But the rapid cooling from laser welding creates a metastable, highly strained ferrite lattice with a dislocation density that lowers the nucleation barrier for intermetallics. Short-duration holds at 1100 °C were evidently enough to trigger early sigma nucleation in this pre-strained matrix, a point that has practical importance for fabricators who may assume that staying above 1000 °C avoids intermetallic problems altogether.

The microhardness trends reinforce the electrochemical data. The flattened profile at 1050 °C (268–281 HV across all zones) indicates good phase-ratio homogeneity, whereas the hardness upturn at 1100 °C in the weld zone (289 HV)

maps onto the sigma-bearing regions identified by SEM-EDS. Similar hardness spikes linked to sigma have been reported for aged duplex steels, though not previously documented for short PWHT cycles applied to laser-welded thin sections ^[12, 14]. For thin-gauge fabricators those producing chemical process tubing, heat exchanger shells or desalination membranes these data argue for tight temperature control and against a conservative "higher is safer" approach to PWHT selection.

Limitations

Several constraints of this research should be acknowledged. First, a single sheet thickness (1.5 mm) was examined; extrapolation to thicker or thinner gauges requires caution because the cooling rate, and therefore the starting weld-zone ferrite fraction, changes markedly with thickness. Second, only one holding time (30 min) was used. A factorial design coupling temperature with time would clarify whether shorter holds at 1050 °C are acceptable, which matters for production throughput. Third, the electrolyte in the CPP tests was a simple 3.5 wt.% NaCl solution. Real service environments in desalination or offshore applications involve higher chloride concentrations, acidified conditions, or the presence of sulfide ions, all of which could shift the relative ranking of the PWHT conditions. Fourth, sigma-phase characterisation relied on SEM-EDS spot analysis; transmission electron microscopy or synchrotron X-ray diffraction would give more definitive phase identification and quantification, particularly for nano-scale precipitates that may fall below the SEM resolution limit. Finally, no fatigue or stress-corrosion cracking tests were conducted; these failure modes are important for structural thin-sheet applications and should be addressed in future work.

Conclusion

This research set out to determine which post-weld heat treatment temperature best restores the pitting corrosion resistance and microhardness uniformity of laser-beam-welded UNS S31803 thin sheets. Four PWHT temperatures spanning 950–1100 °C were applied to 1.5 mm thick autogenous butt welds and evaluated through cyclic potentiodynamic polarisation, Vickers microhardness traverses, ferrite-scope measurements and SEM-EDS analysis. The as-welded condition exhibited the poorest corrosion performance: pitting potentials of 487 ± 31 mV (weld zone) and 512 ± 27 mV (HAZ) versus SCE, paired with the highest weld-zone hardness (318 ± 9 HV) and a ferrite fraction of approximately 72 vol.%. PWHT at progressively higher temperatures improved both properties in a non-linear fashion. At 950 °C, austenite reformation was incomplete, yielding intermediate pitting potentials (638 ± 28 mV weld zone) and still-elevated hardness (298 ± 8 HV). At 1000 °C, further progress was evident (Epit = 754 ± 24 mV, hardness 282 ± 7 HV), but the phase ratio had not yet reached equilibrium.

The optimum was found at 1050 °C / 30 min. Pitting potential peaked at 821 ± 19 mV in the weld zone and 856 ± 16 mV in the HAZ values that actually exceed the base-metal reference (Epit = 843 ± 14 mV). Weld-zone hardness dropped to 268 ± 7 HV, and the entire hardness traverse fell within a 13 HV band (268–281 HV), the flattest profile in the test matrix. Ferrite content in the weld zone was restored to 52.1 ± 1.8 vol.%, near the thermodynamic equilibrium

value and well within the 35–65 % range required by ASTM A789. At 1100 °C the benefits reversed. Pitting potential fell to 763 ± 27 mV (weld zone) and weld-zone hardness rose to 289 ± 8 HV, both linked to sigma-phase particles detected at ferrite–austenite boundaries. The onset of sigma precipitation after only 30 min at 1100 °C is earlier than conventional isothermal transformation diagrams predict, an observation attributed to the high dislocation density retained in the laser-welded ferrite matrix.

From a practical standpoint, fabricators working with thin-gauge UNS S31803 laser welds should target $1050 \text{ °C} \pm 10 \text{ °C}$ for 30 minutes, followed by rapid water quenching. Exceeding 1100 °C, even briefly, carries the risk of intermetallic formation that degrades both pitting resistance and hardness uniformity. Future work should extend the investigation to a matrix of temperatures and holding times, test in more aggressive electrolytes representative of seawater and acidified chloride environments, and include fatigue and stress-corrosion cracking assessments. Examining thinner sheets (0.8–1.0 mm), where the cooling rate after welding is still higher and the ferrite imbalance more severe, would also be valuable for expanding the applicability of these findings to precision-welded industrial components.

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