



SILESIAAN UNIVERSITY OF TECHNOLOGY

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**Material characterization of SLM-printed
duplex stainless steel**

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Declaration of Originality

I, the undersigned, declare that this doctoral dissertation, entitled “*Material characterization of SLM-printed duplex stainless steel*”, is my original work prepared under the supervision of Zbigniew Brytan, PhD, DSc Eng. at the Department of Engineering Materials and Biomaterials, Faculty of Mechanical Engineering.

Except where explicitly acknowledged, cited, or otherwise indicated, the materials, data, methods, figures, interpretations, and conclusions presented in this thesis are my own. To the best of my knowledge, this thesis has not been submitted, either in whole or in part, for the award of any academic degree, diploma, or equivalent qualification at this or any other institution.

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Abstract

Laser powder bed fusion (LPBF) enables fabrication of 2507 super duplex stainless steel (SDSS), but rapid solidification and cyclic reheating generate a non-equilibrium metallurgical state unlike wrought material. This thesis compares standard, +3 wt.% Ni, and +6 wt.% Ni SDSS 2507 variants within an integrated processing–microstructure–mechanical–electrochemical framework. It demonstrates that LPBF parameter selection, stress relief, solution annealing, and Ni modification jointly control austenite stabilisation, ferrite–austenite balance, and response. Alloys were fabricated at $P = 180 \text{ W}$, $v = 300 \text{ mm s}^{-1}$, $h = 120 \text{ }\mu\text{m}$, $t = 30 \text{ }\mu\text{m}$, and volumetric energy density $V_{\text{ED}} = 166.7 \text{ J mm}^{-3}$. Characterisation used scanning electron microscopy (SEM), X-ray diffraction (XRD), electron backscatter diffraction (EBSD), transmission/scanning transmission electron microscopy (TEM/STEM), hardness, tensile, Charpy impact testing, cyclic potentiodynamic polarisation (CPP), electrochemical impedance spectroscopy (EIS), machine-learning-assisted screening based on the apparent CPP-derived screening descriptor $R_{\text{p,app}}$. Response was governed by alloy chemistry, LPBF history, phase constitution, defects, orientation heterogeneity, and local chemistry, not by porosity, Ni content, or heat treatment temperature alone. Standard and +3 wt.% Ni alloys remained ferrite-rich after LPBF and stress relief. After solution annealing at $1100 \text{ }^{\circ}\text{C}$, the standard alloy reached 50.9% α -Fe and 49.1% γ -Fe, while the +3 wt.% Ni alloy became austenite-rich with 69.4% γ -Fe. The +6 wt.% Ni alloy showed higher EBSD-mapped austenite: 52.1% γ -Fe in the as-built (AS), 54.2% after stress relief at 350 and $425 \text{ }^{\circ}\text{C}$ (SR350 and SR425), 60.8% after stress relief at $525 \text{ }^{\circ}\text{C}$ (SR525), and 89.1% after solution annealing at $1150 \text{ }^{\circ}\text{C}$ (SA1150), supporting criterion-specific selection. XRD comparison showed no universal recovery trend: standard stress relief reduced ferrite reflection broadening descriptors, whereas Ni-modified alloys responded non-monotonically. In +6 wt.% Ni, SR425 gave the strongest broadening increase, while SR525 showed the highest kernel average misorientation (KAM)-based orientation heterogeneity. TEM/STEM analysis of the standard alloy stress-relieved at $550 \text{ }^{\circ}\text{C}$ (SR550) showed ferritic nanoscale/substructural contrast, without confirmed sigma, chi, or chromium nitride precipitates. As-built/stress-relieved states retained higher hardness and strength, whereas solution annealing improved elongation and impact response with strength reduction. The +3 wt.% Ni alloy stress-relieved at $400\text{--}450 \text{ }^{\circ}\text{C}$ gave the highest tensile-strength route, reaching $\sim 1414 \text{ MPa}$. Electrochemical ranking changed with corrosion current density (J_{corr}), polarization resistance (R_{p}), charge-transfer resistance (R_{ct}), breakdown potential (E_{br}), and repassivation potential (E_{rp}): standard stress relief at $400 \text{ }^{\circ}\text{C}$ (SR400), +3 wt.% Ni stress relief at $500/550 \text{ }^{\circ}\text{C}$ for CPP descriptors and $450/550 \text{ }^{\circ}\text{C}$ for EIS descriptors, +6 wt.% Ni SR350 for breakdown/repassivation descriptors, and +6 wt.% Ni SR525 for resistance descriptors. The contribution is a criterion-specific pathway map linking LPBF processing, Ni modification, thermal exposure, phase constitution, mechanical response, and laboratory electrochemical descriptors within the domain.

Streszczenie

Selektywne stapianie laserowe proszku w złożu proszkowym, określane jako laser powder bed fusion (LPBF), umożliwia wytwarzanie elementów ze stali nierdzewnej typu super duplex 2507 (SDSS). Jednak bardzo duże szybkości krzepnięcia oraz cykliczne oddziaływanie cieplne kolejnych przejść wiązki laserowej prowadzą do ukształtowania nierównowagowego stanu metalurgicznego, istotnie odmiennego od stanu typowego dla materiałów przerabianych plastycznie. W pracy porównano standardową stal SDSS 2507 oraz jej warianty modyfikowane dodatkiem 3% mas. Ni i 6% mas. Ni. Analizę wykonano w ujęciu zintegrowanym, obejmującym warunki procesu LPBF, mikrostrukturę, właściwości mechaniczne oraz odpowiedź elektrochemiczną. Wykazano, że dobór parametrów LPBF umożliwiających uzyskanie próbek o małej porowatości, zastosowana obróbka cieplna w postaci wyżarzania odpężającego i przesycania, a także modyfikacja składu chemicznego dodatkiem niklu, łącznie wpływają na stabilizację austenitu, równowagę fazową ferryt–austenit oraz właściwości użytkowe badanego materiału. Materiały wytworzono przy następujących parametrach procesu: moc lasera $P = 180 \text{ W}$, prędkość skanowania $v = 300 \text{ mm s}^{-1}$, odległość między ścieżkami $h = 120 \text{ }\mu\text{m}$, grubość warstwy $t = 30 \text{ }\mu\text{m}$ oraz objętościowa gęstość energii $V_{ED} = 166,7 \text{ J mm}^{-3}$. Charakterystykę materiałów wykonano z zastosowaniem skaningowej mikroskopii elektronowej (SEM), dyfrakcji rentgenowskiej (XRD), dyfrakcji elektronów wstecznie rozproszonych (EBSD), transmisyjnej oraz skaningowo-transmisyjnej mikroskopii elektronowej (TEM/STEM), pomiarów twardości, statycznej próby rozciągania, próby udarności metodą Charpy'ego, cyklicznej polaryzacji potencjodynamicznej (CPP), elektrochemicznej spektroskopii impedancyjnej (EIS) oraz analizy przesiewowej z wykorzystaniem uczenia maszynowego, opartej na pozornej rezystancji polaryzacyjnej $R_{p,app}$. Uzyskane wyniki wykazały, że odpowiedź materiału była determinowana przez sprzężone oddziaływanie składu chemicznego stopu, historii cieplnej procesu LPBF, składu fazowego, stanu defektów, heterogeniczności orientacji krystalograficznej oraz lokalnych różnic składu chemicznego. Nie była ona natomiast kontrolowana wyłącznie przez porowatość, zawartość niklu lub temperaturę obróbki cieplnej. Stal o standardowym składzie chemicznym gatunku 2507 oraz stop z dodatkiem 3% mas. Ni po procesie LPBF i po wyżarzaniu odpężającym zachowały strukturę silnie wzbogającą w ferryt. Po przesycaniu w temperaturze $1100 \text{ }^{\circ}\text{C}$ stal standardowa osiągnęła udział $50,9\%$ $\alpha\text{-Fe}$ i $49,1\%$ $\gamma\text{-Fe}$, natomiast stop z dodatkiem 3% mas. Ni uzyskał strukturę wzbogającą w austenit, z udziałem $69,4\%$ $\gamma\text{-Fe}$. Stop z dodatkiem 6% mas. Ni charakteryzował się wyższym udziałem austenitu wyznaczonym metodą EBSD: $52,1\%$ $\gamma\text{-Fe}$ w stanie bezpośrednio po wytworzeniu (AS), $54,2\%$ po wyżarzaniu odpężającym w temperaturze 350 i $425 \text{ }^{\circ}\text{C}$ (SR350 i SR425), $60,8\%$ po wyżarzaniu odpężającym w temperaturze $525 \text{ }^{\circ}\text{C}$ (SR525) oraz $89,1\%$ po przesycaniu w temperaturze $1150 \text{ }^{\circ}\text{C}$ (SA1150). Wyniki te wskazują, że dobór ścieżki technologicznej powinien być uzależniony od przyjętego kryterium oceny właściwości materiału. Analiza efektów wyżarzania odpężającego na podstawie wyników XRD nie

wykazała uniwersalnego trendu zdrowienia mikrostruktury. W przypadku stali o standardowym składzie chemicznym gatunku 2507 wyżarzanie odpężające prowadziło do zmniejszenia wartości parametrów opisujących poszerzenie refleksów dyfrakcyjnych pochodzących od ferrytu, natomiast stopy modyfikowane niklem wykazywały odpowiedź niemonotoniczną. W stopie z dodatkiem 6% mas. Ni stan SR425 charakteryzował się największym wzrostem poszerzenia refleksów dyfrakcyjnych, natomiast stan SR525 wykazywał największą lokalną heterogeniczność orientacji krystalograficznej, określoną na podstawie parametru KAM. Analiza TEM/STEM stali typu 2507 po wyżarzaniu odpężającym w temperaturze 550 °C (SR550) wykazała lokalne zróżnicowanie kontrastu w osnowie ferrytycznej w skali nanometrycznej, jednak nie potwierdziła obecności wydzielen fazy sigma, chi ani azotków chromu w analizowanym obszarze. Materiały w stanie bezpośrednio po druku oraz po wyżarzaniu odpężającym charakteryzowały się większą twardością i wytrzymałością mechaniczną, natomiast przesycanie prowadziło do zwiększenia wydłużenia oraz poprawy odporności na obciążenia uderowe, przy jednoczesnym obniżeniu wytrzymałości. Stop z dodatkiem 3% mas. Ni po wyżarzaniu odpężającym w zakresie 400–450 °C wykazał najwyższą wytrzymałość na rozciąganie, osiągając wartość około 1414 MPa. Ocena odpowiedzi elektrochemicznej była zależna od przyjętego parametru oceny, tj. gęstości prądu korozyjnego J_{corr} , rezystancji polaryzacyjnej R_p , rezystancji przeniesienia ładunku R_{ct} , potencjału przebiccia E_{br} oraz potencjału repasacji E_{rp} . Najkorzystniejsze odpowiedzi uzyskano odpowiednio dla: stali standardowej po wyżarzaniu odpężającym w temperaturze 400 °C (SR400), stopu z dodatkiem 3% mas. Ni po wyżarzaniu odpężającym w temperaturze 500 lub 550 °C w odniesieniu do parametrów CPP oraz 450 lub 550 °C w odniesieniu do parametrów EIS, stopu z dodatkiem 6% mas. Ni w stanie SR350 w odniesieniu do potencjału przebiccia i repasacji oraz stopu z dodatkiem 6% mas. Ni w stanie SR525 w odniesieniu do parametrów rezystancyjnych. Głównym wkładem pracy jest opracowanie kryterialnej mapy ścieżek technologiczno-materiałowych, która łączy warunki procesu LPBF, modyfikację składu chemicznego dodatkiem niklu, oddziaływanie obróbki cieplnej, skład fazowy, właściwości mechaniczne oraz laboratoryjne parametry odpowiedzi elektrochemicznej. Zaproponowane podejście umożliwia dobór wariantu materiałowo-technologicznego w zależności od wymaganego kryterium właściwości, przy uwzględnieniu ograniczeń wynikających z badanego układu materiałowego oraz zastosowanej metodyki eksperymentalnej.

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List of acronyms

AS	As-built condition	ASTM	American Society for Testing and Materials
BCC	Body-centred cubic	CI	Confidence index
CPP	Cyclic potentiodynamic polarization	CPE	Constant phase element
CPT	Critical pitting temperature	CSL	Coincident site lattice
DSS	Duplex stainless steel	EBSD	Electron backscatter diffraction
ECM	Equivalent circuit model	EDS	Energy-dispersive X-ray spectroscopy
EIS	Electrochemical impedance spectroscopy	EN	European standard/number designation
FCC	Face-centred cubic	FC	Furnace cooling
FE-SEM	Field-emission scanning electron microscopy	FIB	Focused ion beam
FWHM	Full width at half maximum	GBA	Grain-boundary austenite
GOS	Grain orientation spread	GROD	Grain reference orientation deviation
HAADF-STEM	High-angle annular dark-field scanning transmission electron microscopy	HAGB	High-angle grain boundary
HDSS	Hyper duplex stainless steel	HRA	Rockwell A hardness
HRTEM	High-resolution transmission electron microscopy	HV	Vickers hardness
ICSD	Inorganic Crystal Structure Database	IPF	Inverse pole figure
ISO	International Organization for Standardization	JCPDS	Joint Committee on Powder Diffraction Standards
KAM	Kernel average misorientation	KS	Kurdjumov–Sachs
LAGB	Low-angle grain boundary	LPBF	Laser powder bed fusion
LSTM	Long short-term memory	MAE	Mean absolute error
ML	Machine learning	MRD	Multiples of random distribution
NW	Nishiyama–Wassermann	OCP	Open-circuit potential
ODF	Orientation distribution function	OR	Orientation relationship
PBF-LB	Powder bed fusion–laser beam; literature nomenclature	PBF-LB/M	Powder bed fusion–laser beam/metal; literature nomenclature
PN-EN	Polish Standard–European Norm	PREN	Pitting resistance equivalent number

RMSE	Root mean square error	SA	Solution-annealed condition
SAED	Selected-area electron diffraction	SAF	Sandvik Austenitic Ferritic
SDSS	Super duplex stainless steel	SEM	Scanning electron microscopy
S/TEM	Scanning/transmission electron microscopy	SR	Stress-relieved condition
STEM	Scanning transmission electron microscopy	STEM-EDS	Scanning transmission electron microscopy–energy-dispersive X-ray spectroscopy
TEM	Transmission electron microscopy	TTT	Time–temperature–transformation
UNS	Unified Numbering System	UTS	Ultimate tensile strength
VED	Volumetric energy density	WQ	Water quenching
XRD	X-ray diffraction	YS	Yield strength

Chapter 1

Introduction

1.1 Background and LPBF metallurgical challenge

Super duplex stainless steels (SDSS), particularly UNS S32750/Grade 2507, are used in chloride containing oil and gas, marine, offshore, and chemical processing environments because they combine high strength with resistance to localized corrosion. This performance depends on a controlled ferrite austenite microstructure. Ferrite contributes strength and resistance to chloride stress-corrosion cracking, whereas austenite improves ductility, toughness, and strain accommodation [1, 2, 3]. The engineering response of SDSS 2507 is therefore governed not by composition alone, but by phase balance, phase morphology, and local chemistry.

Laser powder bed fusion (LPBF) enables geometrically complex SDSS components, but localized melting, steep thermal gradients, rapid solidification, and repeated layer wise reheating impose a non-equilibrium thermal cycle that differs fundamentally from conventional thermomechanical processing [4, 5, 6]. In SDSS 2507, this can restrict diffusion controlled austenite formation and leave the as-built material ferrite rich or predominantly ferritic. Although recent LPBF/PBF-LB studies show that dense DSS and SDSS 2507 parts can be produced, low porosity alone does not establish metallurgical suitability [7, 8, 9]. A dense build may still contain an unfavorable ferrite/austenite balance, residual strain, crystallographic texture, chemical heterogeneity, and precipitation-sensitive ferrite; the central challenge is, therefore, to control the LPBF inherited phase-defect state that governs the subsequent mechanical and electrochemical response.

In this thesis, LPBF-fabricated SDSS 2507 is treated as a non-equilibrium starting state, not as a direct equivalent of wrought or conventionally solution-annealed SDSS. The as-built material may contain a ferrite rich cellular matrix, limited intercellular or boundary associated austenite, process defects, segregation, stored strain, and texture. These features define the starting condition for post build treatment. Stress relief is examined as a route for reducing residual stress related instability, but it may also promote ferritic decomposition, solute clustering, nitride formation, or early stage precipitation depending on temperature, holding time, and alloy chemistry. Solution annealing is examined as a high temperature route for austenite recovery, chemical redistribution, secondary phase dissolution, and duplex rebalancing, with possible strength loss from reduced LPBF derived refinement and stored strain [10, 11].

Nickel modification is used as an alloy design variable to test whether austenite stabilization can assist duplex recovery after LPBF. Additional Ni may promote austenite formation, but it can also change phase connectivity, ferrite stability, Cr/Ni equivalent balance, chemical partitioning, precipitation response, mechanical strength, and passive-film behavior. The Ni-modified alloys are therefore treated as experimental SDSS 2507 variants, not as commercial replacements for UNS S32750.

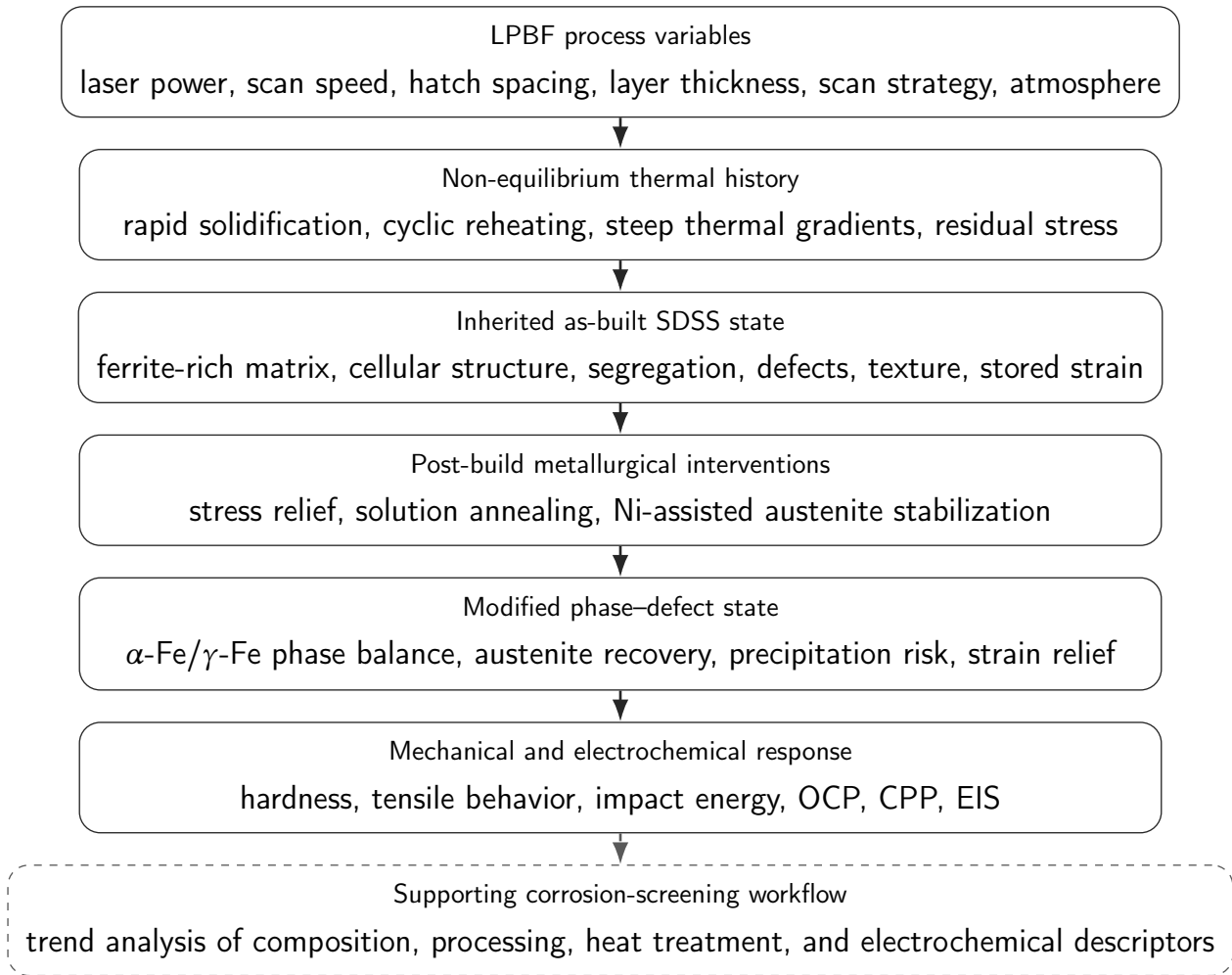


Figure 1.1: Framework linking LPBF processing, thermal history, inherited phase-defect state, post-build treatment, Ni-assisted austenite stabilization, and mechanical/electrochemical response. The dashed block shows the supporting corrosion screening workflow used only for trend comparison with measured OCP, CPP, and EIS results.

Figure 1.1 summarizes the thesis framework. LPBF processing defines the non-equilibrium thermal history and inherited phase-defect state. Heat treatment and Ni modification then change phase balance, residual-strain indicators, precipitation susceptibility, and corrosion-relevant microstructure, which control the mechanical and electrochemical responses. The corrosion-screening workflow is used only as a supporting trend-analysis layer; OCP, CPP, and EIS measurements, interpreted with microstructural evidence, remain the primary corrosion basis.

Table 1.1 positions the thesis against representative LPBF/PBF-LB duplex and super duplex stainless steel studies. The literature shows that dense DSS and SDSS parts can be fabricated, but density alone does not define metallurgical suitability. Phase balance, residual strain, mechanical response, and corrosion behavior remain governed by thermal history and post-build treatment. This motivates the present assessment of Ni-assisted austenite stabilization and heat treatment in LPBF-fabricated SDSS 2507. Corrosion-informatics studies are included only as supplementary descriptor-level context for CPP and EIS interpretation.

Table 1.1: Representative studies positioning the thesis gap in LPBF-fabricated duplex and super duplex stainless steels, separating metallurgical evidence from supplementary corrosion-informatics work.

Study	Material	Focus	Thesis positioning
<i>Metallurgical and phase-corrosion evidence</i>			
Zhu et al. [12]	2507 DSS, LPBF	Process optimisation; properties	Demonstrated dense LPBF 2507 with high hardness and strength, and corrosion behaviour reported as comparable to rolled 2507. This establishes LPBF feasibility for 2507, but does not address Ni-modified alloy design or systematic post-build phase rebalancing.
Gandhi et al. [11]	2507 SDSS, PBF-LB	Solution annealing	Showed that short annealing can promote ferrite/austenite rebalancing, nitride dissolution, and austenite growth. This supports treating solution annealing as phase recovery and chemical redistribution rather than simple property enhancement.
Zhou et al. [13]	Ni-overalloyed 22Cr DSS, LPBF	Ni addition; annealing; corrosion	Reported that Ni increased the as-built austenite fraction, while annealing modified phase balance and pitting behaviour. Strength improved, but pitting resistance did not necessarily improve. This justifies evaluating Ni addition through coupled phase, mechanical, and corrosion responses.
Zhou et al. [14]	Ni-overalloyed 22Cr DSS, LPBF	Laser re-melting	Re-melting improved austenite distribution and elemental homogeneity; repeated re-melting increased pitting potential and critical pitting temperature. This separates process-path effects from nominal composition effects.
Shen et al. [15]	2205 DSS, LPBF	Phase transformation; corrosion	Found that LPBF 2205 showed high strength but reduced elongation and corrosion resistance relative to cast material. This provides a lower-alloy DSS benchmark showing that strength improvement alone does not guarantee acceptable corrosion performance.
Zhao et al. [16]	2205 DSS, LPBF	In-situ EBSD tensile testing	Demonstrated slip, boundary evolution, and phase-dependent strain accommodation during deformation. This supports EBSD-based interpretation of deformation and strain partitioning in LPBF DSS.
Maier et al. [17]	DSS 1.4462, PBF-LB/M	Process-property comparison	Showed that process route and heat treatment strongly affect microstructure and functional properties. This reinforces the need to interpret LPBF duplex stainless steel as a process-specific material state.
Kangazian et al. [18]	SAF 3207 HDSS, LPBF	High-alloy duplex processing	Demonstrated microstructural control in LPBF SAF 3207 hyper duplex stainless steel. This extends the phase-balance problem to more highly alloyed duplex systems.
<i>Supplementary corrosion-informatics evidence</i>			
Coelho et al.; Qiao et al.; Hou et al. [19, 20, 21]	Stainless steels	Data-driven corrosion screening	Showed that machine-learning models can relate chemistry, environment, processing descriptors, and electrochemical metrics when the target variable and applicability domain are clearly defined. This supports corrosion informatics as a bounded descriptor-level screening layer rather than the primary metallurgical gap.

1.2 Research gaps, questions, and hypotheses

LPBF-fabricated DSS/SDSS commonly develops ferrite-rich, non-equilibrium microstructures, and post-build heat treatment can partly recover duplex balance depending on alloy chemistry, thermal exposure, and the inherited as-built condition [10, 7, 11]. However, predictive control of LPBF-fabricated SDSS 2507 remains limited because processing, heat treatment, phase balance, defects, mechanical response, and corrosion behavior are often examined separately rather than as one coupled metallurgical problem [9, 8, 22]. This thesis addresses the following gaps.

- **Uncritical transfer of wrought or welded SDSS heat-treatment logic.** Stress relief and solution annealing are often interpreted using conventional SDSS metallurgy, although LPBF-fabricated SDSS starts from a ferrite-rich, chemically heterogeneous, defect-containing, and residually strained condition [1, 3, 2]. In this thesis, the term phase–defect state denotes the combined phase constitution, defect population, residual-strain indicators, and LPBF-inherited microstructural heterogeneity. The heat-treatment response must therefore be evaluated from the inherited as-built phase–defect state.
- **Insufficiently resolved role of deliberate Ni modification.** Ni is an austenite-stabilizing element, but its role in LPBF-fabricated SDSS 2507 remains insufficiently resolved with respect to austenite stabilization, phase morphology, precipitation sensitivity, mechanical response, and corrosion behavior. In this thesis, Ni modification is assessed as an alloy-design variable under the investigated thermal pathways, not as an automatically beneficial or strictly monotonic composition series.
- **Separated interpretation of mechanical and corrosion behavior.** Mechanical and corrosion responses are often interpreted independently, although both depend on phase balance, local chemistry, defect morphology, residual strain, and passive-film stability [23, 24, 25]. An interpretation requires these responses to be related to the same microstructural basis.
- **Single-technique methodological limitations.** X-ray diffraction (XRD), electron backscatter diffraction (EBSD), scanning electron microscopy/energy-dispersive X-ray spectroscopy (SEM/EDS), transmission electron microscopy (TEM), porosity analysis, mechanical testing, and electrochemical testing each have method-specific limitations. Conclusions must therefore rely on convergent evidence rather than on a single characterization or testing method.
- **Bounded use of data-driven corrosion screening.** Machine-learning models can screen composition, processing, descriptor, and electrochemical spaces, but their interpretation depends on training-data coverage, protocol consistency, feature design, and applicability domain. In this thesis, corrosion informatics is used as a descriptor-level screening and transfer-assessment tool. The +3 wt.% Ni dataset provides the closest experimental reference

condition, while the standard SDSS 2507 and +6 wt.% Ni datasets assess qualitative cross-alloy consistency. The model is not used as an independent validation of localized-corrosion resistance. These gaps lead to the following research questions:

1. How does deliberate Ni addition affect the as-built LPBF phase-defect state of SDSS 2507, and to what extent does it promote ferrite–austenite phase balance relative to the standard alloy?
2. How do the selected stress-relief and solution-annealing treatments modify phase balance, microstructure, residual-strain indicators, and precipitation-sensitive features?
3. How do the Ni-modified alloy conditions respond under the investigated post-build thermal pathways with respect to austenite stabilization, phase morphology, and thermal response?
4. How do the as-built and heat-treated microstructural states affect hardness, tensile response, and impact behavior?
5. How do alloy chemistry and post-build thermal condition influence open-circuit potential (OCP), cyclic potentiodynamic polarization (CPP), electrochemical impedance spectroscopy (EIS) response, and screening-level localized-corrosion susceptibility?
6. Which investigated alloy–heat-treatment pathway provides the most balance among phase constitution, mechanical performance, and electrochemical screening response?

The data-driven component adds a machine-learning-assisted electrochemical screening framework to the experimental corrosion analysis. Using the +3 wt.% Ni dataset for calibration, the workflow evaluates whether corrosion-response trends can be compared qualitatively across the standard SDSS 2507, +3 wt.% Ni, and +6 wt.% Ni alloy systems. This calibration does not constitute external validation and is used only to assess internal descriptor-level consistency. This provides a supplementary basis for organising cross-alloy descriptor trends and comparing them with the experimental cyclic potentiodynamic polarization (CPP) and electrochemical impedance spectroscopy (EIS) response. These working hypotheses are evaluated only within the investigated LPBF route, alloy compositions, alloy-specific heat-treatment schedules, specimen orientations, surface preparation, mapped regions, and electrochemical protocol. The thesis tests three working hypotheses:

1. **Effect of Ni addition on the as-built LPBF phase state.** This hypothesis tests whether deliberate Ni addition shifts the as-built LPBF phase constitution toward increased austenite formation relative to the standard alloy, while recognizing that excessive Ni may promote austenite-rich rather than balanced duplex states. Because the applied LPBF parameter set was optimized primarily for low porosity on the selected machine, complete wrought-like ferrite–austenite phase balance is not assumed in the as-built condition.

2. **Effect of post-build heat treatment on phase balance and microstructure.** This hypothesis tests whether stress relief and solution annealing modify the LPBF-inherited microstructure through different thermal pathways. Stress relief is expected to mainly affect the retained as-built state, whereas solution annealing is expected to promote stronger ferrite–austenite rebalancing. The extent of these changes is expected to depend on Ni content and heat-treatment condition.
3. **Effect of phase–defect–strain state on material response.** This hypothesis tests whether the mechanical and electrochemical responses depend on the combined phase constitution, defect state, residual-strain indicators, and heat-treatment condition rather than on Ni content or porosity alone. Ni addition and heat treatment are therefore expected to produce criterion-dependent changes in strength, ductility, impact response, and electrochemical behaviour.

These hypotheses are tested using SEM, XRD, EBSD, TEM/STEM, where available, mechanical testing, CPP, EIS, and the bounded corrosion-screening workflow. The screening model is used only to compare descriptor-level trends with the measured electrochemical response.

1.3 Aim and objectives

This thesis determines how post-build heat treatment and deliberate Ni modification affect the LPBF-inherited microstructure, mechanical response, and electrochemical behavior of SDSS 2507. Standard UNS S32750, SDSS 2507 +3 wt.% Ni, and SDSS 2507 +6 wt.% Ni are compared using a common LPBF processing basis and alloy-specific heat-treatment pathways; therefore, the comparison is cross-alloy rather than a simple Ni-content series. The machine-learning-assisted analysis is used only as a supplementary descriptor-level comparison with the measured electrochemical trends. This aim is pursued through six specific objectives:

- (1) Establish the feedstock characteristics and LPBF process-window basis used to fabricate standard SDSS 2507 and the Ni-modified variants.
- (2) Determine how the selected alloy-specific post-build stress-relief and solution-annealing treatments modify the LPBF-inherited phase balance, microstructure, residual-strain indicators, and precipitation-sensitive features.
- (3) Assess how deliberate Ni modification affects phase constitution, austenite stabilization, and thermal response under the investigated alloy-specific heat-treatment pathways, while accounting for differences in heat-treatment temperature, holding time, cooling route, and available characterization evidence.
- (4) Characterize the as-built and heat-treated microstructures using complementary scanning electron microscopy (SEM), X-ray diffraction (XRD), electron backscatter diffraction (EBSD), and transmission electron microscopy (TEM) evidence, with method-specific interpretation

according to data availability. SEM and XRD are used for cross-alloy microstructural and phase-identification comparison, EBSD is used for crystallographic and orientation-spread analysis where available, and TEM is used for selected nanoscale assessment.

- (5) Evaluate hardness, tensile behavior, and impact response as functions of alloy chemistry and post-build thermal condition, with direct cross-alloy comparisons restricted to metallurgically comparable treatment states.
- (6) Assess electrochemical behavior in a chloride-containing environment using open-circuit potential (OCP), cyclic potentiodynamic polarization (CPP), and electrochemical impedance spectroscopy (EIS), and compare these results with the bounded descriptor-level corrosion-screening workflow. The experimental electrochemical data remain the primary corrosion evidence, while the machine-learning-assisted framework is used only to assess qualitative consistency of selected descriptor trends.

1.4 Scope and thesis organization

This thesis focuses on LPBF-fabricated standard SDSS 2507 and two experimentally Ni-modified variants: SDSS 2507 +3 wt.% Ni and SDSS 2507 +6 wt.% Ni. The study covers the as-built condition and selected post-build stress-relief and solution-annealing treatments. The interpretation is limited to the investigated LPBF route, alloy compositions, alloy-specific heat-treatment schedules, specimen orientations, surface preparations, exposed sections, and electrochemical testing environment.

Because the heat-treatment matrix was alloy-specific rather than fully factorial, cross-alloy comparisons in this thesis are interpreted as pathway-based comparisons. The results are therefore not used to isolate direct temperature-by-temperature, time-by-time, or cooling-history effects across all alloy systems.

The central contribution is an integrated assessment of how LPBF processing, Ni modification, and post-build heat treatment affect phase constitution, microstructure, mechanical response, and electrochemical behavior. The thesis does not seek a single condition that maximizes one property. Instead, it evaluates which alloy–heat-treatment condition gives the most balance among phase constitution, strength, ductility, impact response, laboratory electrochemical descriptors related to passivity, repassivation tendency, and screening-level localized-corrosion susceptibility.

The corrosion analysis is based on open-circuit potential (OCP), cyclic potentiodynamic polarization (CPP), and electrochemical impedance spectroscopy (EIS) measurements in 3.5 wt.% NaCl. These methods are used to compare electrochemical response, repassivation tendency, polarization resistance, and charge-transfer resistance across the investigated alloy–heat-treatment states. Because quantitative post-corrosion morphology, pit-depth statistics, pit-density analysis, and critical pitting temperature measurements are outside the present dataset, the corrosion results are interpreted as screening-level indicators rather than as proof of service localized-corrosion resistance.

The data-driven component is included as a supplementary corrosion-screening layer. It compares descriptor-level trends in composition, processing, heat treatment, and apparent polarization resistance with the measured electrochemical response. The +3 wt.% Ni alloy provides the closest experimental reference condition for model assessment, while the standard SDSS 2507 and +6 wt.% Ni variants are used to examine cross-alloy transfer consistency. The model output is therefore not treated as independent validation of localized-corrosion resistance, but as a bounded support tool for comparing electrochemical trends across the investigated alloy states.

Fatigue, stress-corrosion cracking, crevice corrosion, critical pitting temperature, quantitative pit morphology, long-term immersion, long-term thermal aging, rough as-built surface exposure, and service-scale component validation are outside the present scope and are treated as future work. Within these constraints, the thesis establishes a comparative pathway map for LPBF SDSS 2507 alloy design and post-build treatment, linking alloy chemistry, thermal exposure, phase constitution, mechanical performance, and electrochemical screening response.

The thesis is organized as follows. Chapter 2 reviews SDSS 2507 metallurgy, LPBF-induced non-equilibrium microstructures, post-build heat-treatment response, Ni-assisted austenite stabilization, mechanical behavior, corrosion response, methodological constraints, and data-driven corrosion screening. Chapter 3 describes the alloy preparation, LPBF processing, alloy-specific heat-treatment matrices, sample preparation, characterization methods, mechanical testing, electrochemical procedures, and machine-learning-assisted screening methodology. Chapter 4 presents the experimental results and descriptor-based electrochemical assessment. Chapter 5 discusses the relationships among LPBF route, alloy design, post-build heat treatment, microstructure, mechanical response, and electrochemical behavior. Chapter 6 summarizes the conclusions, original contributions, limitations, and recommendations for future work.

Chapter 2

Literature Review

This chapter critically synthesizes the literature needed to interpret LPBF-fabricated SDSS 2507 as a non-equilibrium alloy system rather than a direct extension of wrought duplex stainless steel metallurgy. It is structured around three questions: what is established, what remains unresolved, and how the present thesis contributes original insight. The central gap is that LPBF-fabricated SDSS is often examined through isolated variables, whereas its performance is governed by a coupled metastable state involving ferrite retention, chemical heterogeneity, residual strain, defects, texture, and precipitation risk. This gap motivates the comparative standard LPBF SDSS 2507 of +3 wt.% Ni and +6 wt.% Ni alloy design used in this dissertation.

2.1 SDSS 2507 metallurgy

Duplex stainless steels are Fe–Cr–Ni–Mo–N alloys designed to develop a mixed ferrite–austenite microstructure after appropriate thermomechanical processing and heat treatment. Their engineering value arises from phase complementarity rather than from either phase alone: ferrite contributes high strength and resistance to chloride stress-corrosion cracking, whereas austenite improves ductility, toughness, and strain accommodation [1, 26, 27, 3]. In super duplex stainless steels, this balance is achieved at higher Cr, Mo, and N contents, producing higher localized-corrosion resistance but also a narrower thermal-processing window [2, 28, 27].

UNS S32750, commonly known as SDSS 2507, belongs to the super duplex class and is typically alloyed with approximately 25 wt.% Cr, 7 wt.% Ni, 4 wt.% Mo, and 0.3 wt.% N [29, 30, 27]. The high Cr and Mo contents support passive-film stability and chloride-pitting resistance, while nitrogen increases austenite stability, solid-solution strength, and pitting resistance [1, 2, 28]. However, the same high alloy content increases susceptibility to deleterious phase formation during unsuitable thermal exposure, especially Cr- and Mo-rich intermetallics and chromium nitrides [2, 31, 32, 27]. Therefore, phase balance in SDSS 2507 is not merely a descriptive microstructural parameter; it is a governing variable linking processing history to strength, toughness, passivity, and localized-corrosion resistance.

In wrought SDSS 2507, an approximately balanced ferrite–austenite microstructure is normally targeted after solution annealing and rapid cooling. Departures from this balance can degrade performance in different ways. Excessive ferrite may increase strength but reduce toughness and limit austenite-mediated strain accommodation, whereas excessive austenite may reduce strength and alter phase connectivity. More critically, phase imbalance can change element partitioning between ferrite and austenite, thereby affecting phase-specific pitting resistance and the local thermodynamic driving force for secondary-phase precipitation [1, 2, 31, 3].

The pitting resistance equivalent number (PREN) is commonly used to compare the nominal resistance of duplex stainless steels to chloride-induced localized corrosion. For duplex grades, the

most widely used expression is

$$\text{PREN}_{16} = \text{Cr} + 3.3 \text{Mo} + 16 \text{N}, \quad (2.1)$$

where Cr, Mo, and N are expressed in wt.%. Alternative formulations use a higher nitrogen coefficient to emphasize the strong beneficial effect of nitrogen in duplex stainless steels:

$$\text{PREN}_{30} = \text{Cr} + 3.3 \text{Mo} + 30 \text{N}. \quad (2.2)$$

For W-containing duplex grades, the tungsten contribution may be included by modifying the molybdenum term:

$$\text{PREN}_{\text{W},16} = \text{Cr} + 3.3(\text{Mo} + 0.5 \text{W}) + 16 \text{N}. \quad (2.3)$$

An analogous W-containing expression may also be written using the higher nitrogen coefficient. Therefore, the selected PREN formula should always be stated explicitly, because values calculated using different nitrogen coefficients are not directly comparable [27].

PREN is useful as a first-order compositional ranking parameter, but it is not a microstructural or electrochemical proof of corrosion resistance. In duplex stainless steels, effective localized-corrosion resistance also depends on phase-specific partitioning of Cr, Mo, and N; the phase with the lower local PREN may become the limiting phase for pitting resistance [1, 29]. This limitation is particularly important for additively manufactured duplex and super duplex stainless steels, because rapid solidification, cyclic reheating, solute redistribution, and cellular segregation can produce local chemistry and phase distributions that differ from the nominal bulk composition [33, 34, 35].

To position UNS S32750 within the broader duplex stainless steel family, Table 2.1 summarizes the nominal composition ranges and representative compositions of selected lean duplex, standard duplex, super duplex, and hyper duplex grades. This comparison highlights why SDSS 2507 is classified as a super duplex stainless steel: it combines high Cr, Mo, and N contents with a PREN typically above 40, but this high-alloy chemistry also increases sensitivity to phase imbalance and secondary-phase precipitation.

The comparison in Table 2.1 clarifies the compositional distinction between standard duplex and super duplex grades. Compared with standard duplex grades such as S31803/S32205, SDSS 2507 contains higher Cr, Mo, and N contents, which improves nominal pitting resistance but also increases the need for careful thermal control during LPBF and post-build heat treatment.

The implication for LPBF is direct. The phase constitution expected in wrought SDSS 2507 cannot be assumed after rapid solidification. Recent LPBF/PBF-LB studies on SDSS 2507 show that dense parts can be produced, but the as-built microstructure is commonly ferrite-rich because rapid solidification suppresses diffusion-controlled austenite formation [7, 36]. Post-build solution annealing can promote austenite recovery and improve duplex phase balance, but it can also change strength, ductility, residual strain, precipitation state, and corrosion response [36, 11].

Table 2.1: Nominal chemical composition ranges and representative compositions (wt.%) of selected duplex and super duplex stainless steel grades according to UNS designations. Adapted from [1, 27, 29, 30].

Type	UNS grade	Cr	Ni	Mo	N	PREN
Lean duplex	Range	20–24	1–5	0.1–0.3	0.10–0.22	22–27
	S32003	21.0	1.5	0.0	0.18	24.5
	S32101	22.0	1.5	0.3	0.18	26.0
	S32202	22.0	1.5	0.5	0.20	27.0
Standard duplex	Range	21–23	4.5–6.5	2.5–3.5	0.10–0.22	28–38
	S31803	22.0	5.0	3.0	0.16	33–35
	S32205	22.0	5.5	3.1	0.18	≥34
Super duplex	Range	24–29	4.5–8.2	3–5	0.10–0.35	38–45
	S32750	24–26	6–8	4–5	0.20–0.30	42–43
	S32760	25–27	6–8	4.5–6	0.20–0.30	43–44
Hyper duplex	Range	27.0	6.5	4.8	0.40	>45
	S32707	27.0	6.5	4.8	0.40	≥48

Therefore, the desirable duplex balance must either be promoted during fabrication or recovered through alloy modification and post-build thermal treatment. This is the metallurgical basis for comparing the standard SDSS 2507 alloy with +3 wt.% Ni and +6 wt.% Ni variants in this dissertation.

2.2 Alloying and phase stability

Phase stability in duplex and super duplex stainless steels is governed by the balance between ferrite-stabilizing and austenite-stabilizing elements. Chromium and molybdenum stabilize ferrite and improve resistance to localized corrosion, whereas nickel and nitrogen stabilize austenite and promote ferrite-to-austenite transformation during cooling or annealing [1, 3, 27]. This alloying balance controls phase fraction, phase morphology, elemental partitioning, precipitation susceptibility, and electrochemical response.

Chromium is indispensable for passivation, but its benefit depends on remaining in solid solution and being sufficiently available within both ferrite and austenite. Chromium depletion adjacent to sigma phase, chromium nitrides, or other Cr-rich precipitates can reduce localized-corrosion resistance even when the nominal bulk Cr content remains high [2, 32, 28]. Molybdenum improves resistance to pit initiation and pit propagation in chloride-containing environments, but it also stabilizes ferrite and promotes Mo-rich intermetallics such as chi and sigma during inappropriate thermal exposure [2, 31, 32]. Nitrogen contributes to austenite stability, solid-solution strengthening, and pitting resistance through its strong contribution to PREN; however, in ferrite-rich LPBF microstructures, nitrogen supersaturation may promote chromium nitride precipitation during thermal exposure [3, 11]. Carbon is intentionally kept low to suppress carbide sensitization.

In this thesis, nitrogen is not treated as an independent alloy-design variable. It is considered part of the Cr–Ni–Mo–N background chemistry of SDSS 2507 and as a factor that may influence austenite stabilization, phase-specific PREN, and nitride-related precipitation risk. The deliberate compositional variable examined here is nickel addition. Therefore, nitrogen is discussed only where it affects the interpretation of Ni-assisted austenite stabilization, ferrite retention, and post-build heat-treatment response in LPBF-fabricated SDSS 2507.

2.2.1 Nickel-assisted austenite stabilization

Nickel is central to this dissertation because it is used as a deliberate compositional variable to test whether austenite stabilization can counteract the ferrite-rich non-equilibrium state produced by LPBF. In conventional duplex metallurgy, Ni increases the austenite-stabilizing tendency and shifts the ferrite–austenite balance toward a higher austenite fraction [1, 3, 28]. This approach is also well established in duplex stainless-steel welding metallurgy. During welding, rapid thermal cycling and high cooling rates can suppress austenite reformation and leave excessive ferrite in the weld metal. Duplex stainless-steel filler materials are therefore commonly designed with higher Ni content than the base metal to promote austenite reformation during cooling and restore a more favourable ferrite–austenite phase balance in the weld zone [1, 3]. By analogy, Ni addition in LPBF-processed duplex stainless steel may act as an austenite-stabilizing strategy to compensate partly for the strongly non-equilibrium solidification and rapid cooling conditions characteristic of additive manufacturing.

Under LPBF conditions, this effect is particularly relevant because high cooling rates restrict diffusion-controlled austenite formation during solidification and early cooling. Recent [7, 11, 37] LPBF studies on 2507 and Ni-overalloyed duplex stainless steels similarly show that rapid solidification promotes ferrite-rich as-built states, while alloy modification or post-build annealing can increase the austenite fraction. The expected benefit of Ni enrichment is therefore not limited to phase fraction. By promoting austenite formation, Ni may also alter phase connectivity, ferrite/austenite interface density, local strain partitioning, and the extent of ferrite supersaturation. These changes can influence hardness, tensile response, impact behavior, and corrosion performance. However, Ni enrichment should not be treated as intrinsically beneficial. Recent [13] work on Ni-overalloyed LPBF duplex stainless steel shows that increased Ni can improve austenite formation and tensile response without necessarily improving pitting resistance. Therefore, the +3 wt.% Ni and +6 wt.% Ni alloys in this thesis are treated as experimental alloy modifications whose value must be judged by the coupled phase-balance–property–corrosion response rather than by Ni content alone.

2.3 Phase transformations and secondary phases

Figure 2.1 illustrates the crystallographic distinction between ferrite and austenite and the compositional progression from ferritic to increasingly austenitic microstructures with progressive austenite stabilization [38]. The defining feature of SDSS is the coexistence of ferrite and austenite. In conventionally processed SDSS, the final phase balance is approached through solidification and

subsequent solid-state transformation under cooling conditions that permit substantial austenite formation [1]. In LPBF, the thermal path is radically different: rapid solidification favors ferrite retention, while cyclic reheating may only partially promote austenite nucleation at cell boundaries, melt-pool boundaries, grain boundaries, or reheated regions [39, 40].

A terminology distinction is necessary. In duplex stainless-steel literature, ferrite is sometimes denoted as δ when referring to the high-temperature solidification product and as α when referring to the ferritic BCC phase retained at room temperature. In this thesis, δ is used where the discussion concerns solidification and ferrite retention from the LPBF thermal path, whereas discussions on ferrite/austenite phase-balance refer to the retained BCC ferritic phase and FCC austenite.

Duplex and super duplex stainless steels generally solidify through a ferrite-containing mode, followed by partial solid-state transformation of ferrite to austenite during cooling. The extent of this transformation is governed by composition, cooling rate, ferrite grain size, segregation, and the availability of nucleation sites [1, 3, 2]. Therefore, the duplex structure is not produced by solidification alone; it requires sufficient thermal exposure for diffusion-controlled austenite nucleation and growth from the ferritic matrix.

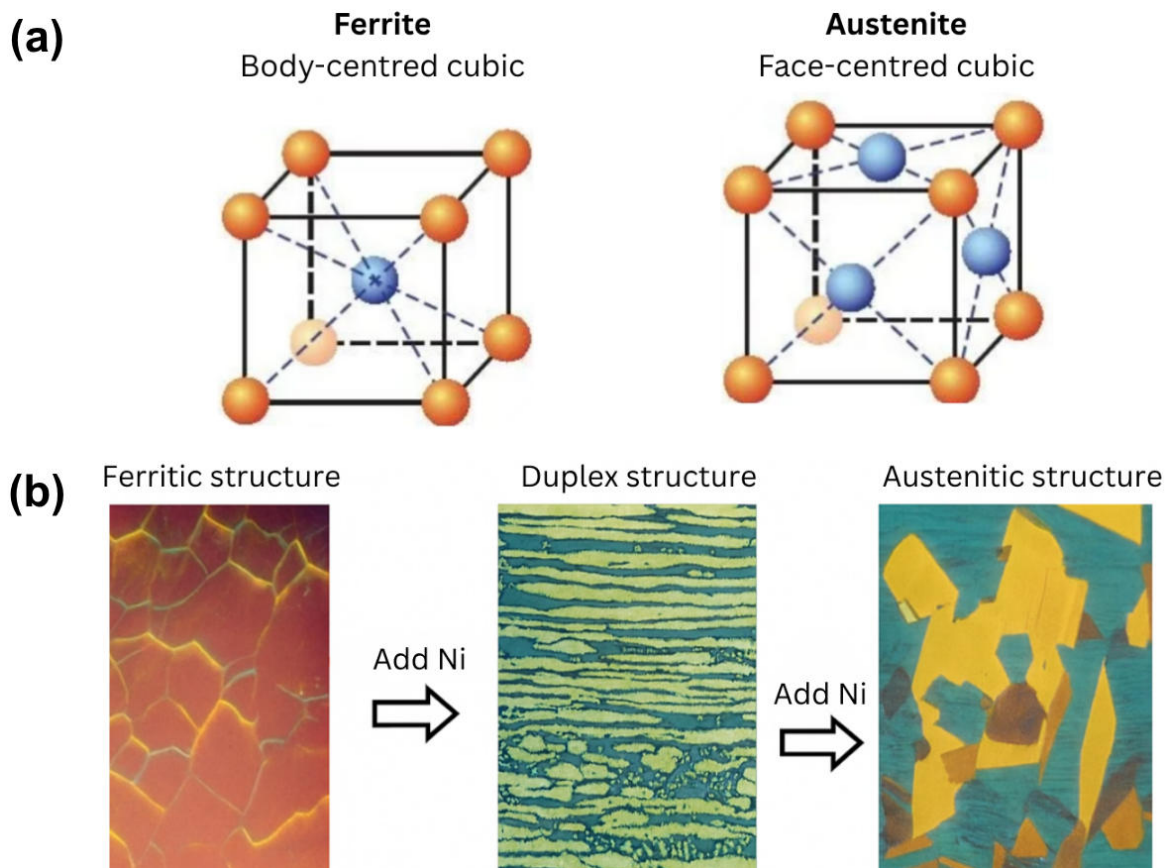


Figure 2.1: Crystal-structure and composition–microstructure schematic for duplex stainless steels. Ferrite is BCC, austenite is FCC, and progressive austenite stabilization shifts the microstructure from ferritic to duplex and then toward austenite-rich states. Adapted from [38].

The most important transformation in SDSS 2507 is the ferrite-to-austenite transformation during cooling or subsequent thermal exposure. Austenite may form as grain-boundary allotriomorphs,

Widmanstätten side plates, intragranular austenite, or secondary austenite, depending on cooling rate, local chemistry, nucleation-site availability, and reheating history. In LPBF, this transformation is kinetically restricted because the solidification and cooling cycles are too rapid for extensive diffusion-controlled austenite growth. Therefore, the LPBF as-built condition is not simply a refined wrought microstructure; it is a metastable ferritic matrix from which duplex balance must later be recovered.

Austenite morphology is important because it controls phase connectivity, boundary density, and chemical partitioning. Grain-boundary austenite can provide continuous transformation paths along ferrite boundaries, whereas Widmanstätten and intragranular austenite reflect different combinations of undercooling, supersaturation, and nucleation-site density [41, 42, 43]. Secondary austenite is particularly important in heat-treated or reheated duplex steels because it may form with chemistry different from primary austenite, especially when associated with nitride dissolution, local Cr/Mo depletion, or partitioning during intermediate-temperature exposure [42, 3].

Beyond the primary ferrite–austenite balance, SDSS 2507 is sensitive to deleterious secondary phases. Sigma, chi, chromium nitrides, secondary austenite, and ferritic decomposition can redistribute Cr, Mo, and N in ways that simultaneously reduce toughness and destabilize passivity [44]. For a thesis focused on coupled mechanical and corrosion performance, these phases must be treated as performance-controlling features rather than as isolated metallographic observations. The precipitation sensitivity of SDSS 2507 is a direct consequence of its high Cr, Mo, and N contents. These elements are beneficial for pitting resistance and strength when retained in solid solution, but they also increase the driving force for intermetallic or nitride formation during unsuitable thermal exposure [2, 31, 32]. This creates the central heat-treatment challenge for SDSS: thermal exposure must be sufficient to recover or rebalance austenite, but not so prolonged in precipitation-sensitive ranges that it promotes local Cr/Mo/N depletion.

Figure 2.2 is important because SDSS 2507 is expected to solidify through a ferrite-containing mode rather than as fully austenitic material. The Cr-equivalent/Ni-equivalent balance controls whether solidification follows A, AF, FA, or F modes. Here, A denotes fully austenitic solidification, AF denotes primary austenitic solidification followed by ferrite formation, FA denotes primary ferritic solidification followed by austenite formation, and F denotes fully ferritic solidification. For high-Cr, Mo-containing duplex alloys, primary ferrite formation is favored; therefore, FA or F-type solidification is expected rather than fully austenitic solidification. Under conventional cooling, part of this primary ferrite can transform to austenite. Under LPBF cooling, however, the ferrite-to-austenite transformation is kinetically restricted, so the solidification path can produce a ferrite-rich inherited state. This provides the metallurgical basis for using Ni modification and post-build heat treatment as interventions in the present thesis.

The solidification-mode framework is useful, but it should not be treated as a complete predictor for LPBF SDSS. Equivalent-ratio diagrams were largely developed for welding and conventional solidification contexts, whereas LPBF involves repeated remelting, rapid solidification, steep thermal gradients, scan-vector-dependent heat accumulation, and melt-pool-scale segregation.

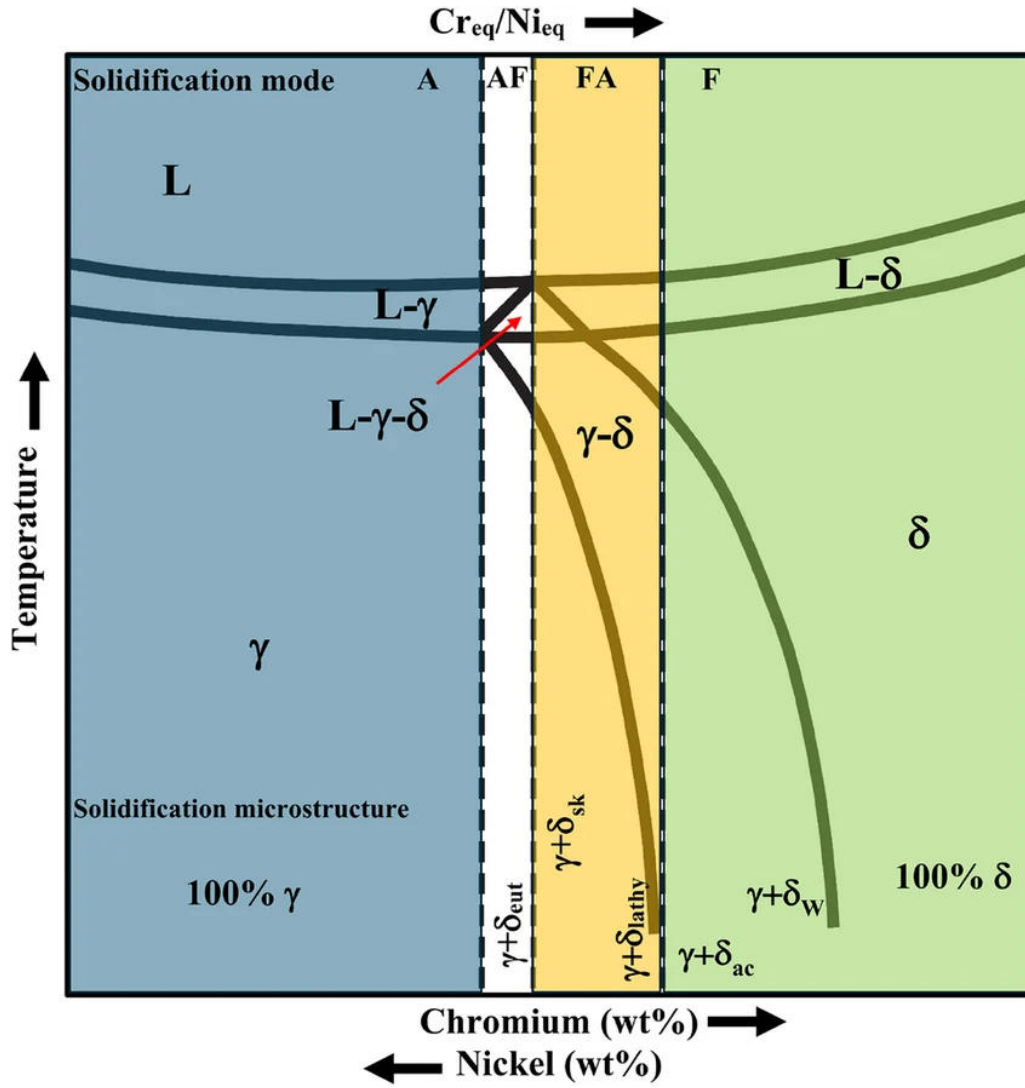


Figure 2.2: Schematic pseudobinary Fe–Cr–Ni phase diagram showing the effect of the Cr-equivalent/Ni-equivalent ratio on A, AF, FA, and F solidification modes. Adapted from [45].

Consequently, nominal Cr-equivalent/Ni-equivalent balance provides only a first-order expectation; the observed phase state must still be established experimentally using XRD, EBSD, SEM, and TEM evidence.

Figure 2.3 illustrates why the heat-treatment window for duplex stainless steels is narrow. Sigma-phase precipitation is typically favoured at intermediate-to-high temperatures, especially near 700–950 °C, depending on alloy composition and exposure time. By contrast, ferritic decomposition associated with 475 °C embrittlement occurs at lower intermediate temperatures. Therefore, sigma formation should not be overemphasized at approximately 550 °C unless it is supported by long exposure times or direct microstructural evidence. At this temperature, ferritic decomposition, clustering, nitride formation, or early-stage precipitation may provide more interpretations.

Low-temperature ferritic decomposition deserves explicit attention because LPBF SDSS can retain a high ferrite fraction in the as-built state. In the 300–550 °C range, ferrite may undergo Cr-rich/Fe-rich separation associated with α/α' decomposition and age-hardening behaviour [46,

47, 48]. Such decomposition may increase hardness and reduce toughness without producing obvious SEM-scale precipitates. Therefore, changes after low-temperature stress relief should not be attributed automatically to sigma phase unless direct crystallographic or nanoscale evidence is available.

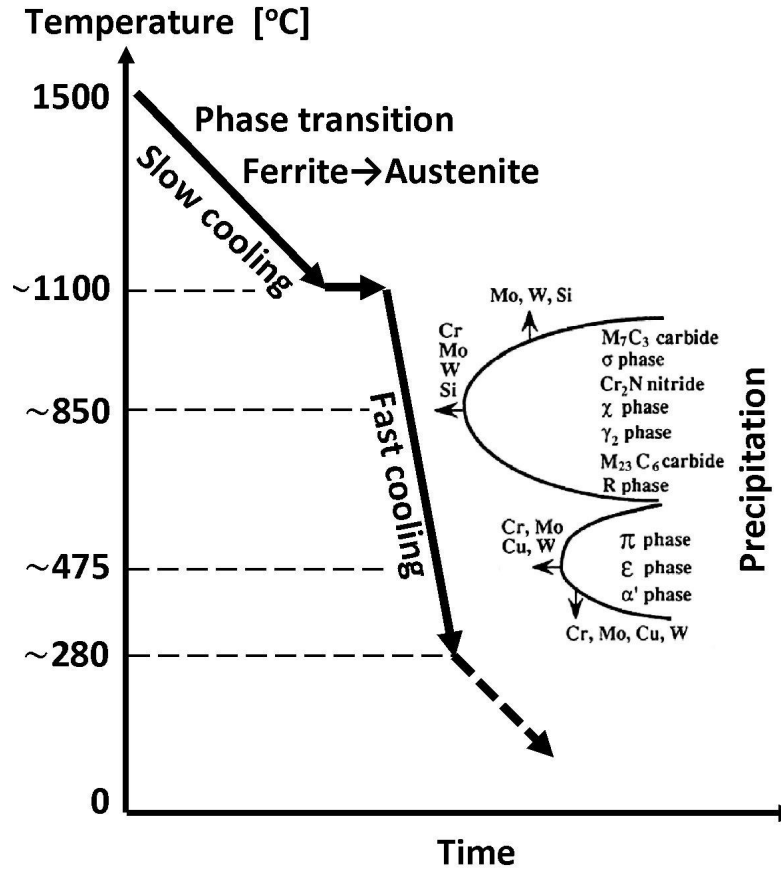


Figure 2.3: Schematic TTT diagram for duplex stainless steels showing thermal ranges associated with precipitation and ferritic decomposition. Adapted from [49].

Transformation behavior in SDSS 2507 is controlled not only by temperature, but also by exposure time, alloy chemistry, ferrite fraction, segregation, dislocation density, and nucleation-site density. These factors are especially important in LPBF material because cellular segregation, retained ferrite, residual strain, and repeated reheating can localize or accelerate transformations relative to wrought SDSS. Table 2.2 summarizes the main transformation windows relevant to LPBF-fabricated SDSS 2507, and temperature ranges are approximate. Transformation onset and severity depend on composition, prior phase balance, cooling rate, exposure time, and the LPBF-inherited microstructural state.

Table 2.3 summarizes the main secondary phases and transformation risks in heat-treated LPBF-fabricated SDSS 2507. Chromium nitrides are treated separately because ferrite-rich LPBF microstructures can retain nitrogen supersaturation when austenite formation is kinetically restricted. Phase detection must be separated from phase consequence. Sigma and chi may be detectable by scanning electron microscopy (SEM), electron backscatter diffraction (EBSD), or X-ray diffraction

Table 2.2: Transformation windows relevant to LPBF-fabricated SDSS 2507. Adapted from [50, 51].

Temperature	Dominant transformation risk	Relevance to LPBF SDSS 2507
300–400 °C	Residual-stress relief; early clustering or ferritic decomposition may occur during sufficient exposure.	May reduce residual-stress effects, but is unlikely to restore duplex balance. Interpretation should focus on stress relaxation, local strain, and early ferritic instability rather than austenite recovery.
425–550 °C	475 °C embrittlement regime; α/α' separation, clustering, and possible nitride-related phenomena.	Particularly important for ferrite-rich LPBF states. Hardness increase or toughness loss in this range should not be assigned to sigma phase without direct evidence.
600–1000 °C	Sigma, chi, chromium nitrides, and secondary austenite may form depending on time and chemistry.	High-risk precipitation range. Local Cr/Mo/N depletion can degrade toughness and pitting response even when the precipitate fraction is small.
1050–1150 °C	Dissolution of secondary phases, austenite recovery, homogenization, and duplex rebalancing.	Useful for restoring duplex morphology, but may reduce strength by removing LPBF-inherited dislocation density, residual strain, and cellular refinement.

(XRD) once sufficiently large or abundant. In contrast, α' , G phase, solute clustering, and early nitride precipitation may require transmission electron microscopy (TEM), atom-probe tomography, or indirect evidence from hardness, magnetic response, or diffraction broadening. Therefore, the absence of SEM-visible precipitates or clear XRD peaks does not prove that nanoscale decomposition or precipitation is absent.

Nitride precipitation should not be grouped with intermetallic precipitation. Sigma and chi are mainly Cr–Mo-rich intermetallics, whereas chromium nitrides are controlled by nitrogen supersaturation, ferrite fraction, cooling rate, and austenite reformation. This distinction matters because retained ferrite in LPBF SDSS may accommodate nitrogen less effectively than a balanced ferrite–austenite microstructure.

Secondary phases in LPBF-fabricated SDSS 2507 should therefore be evaluated primarily by their effect on local chemistry rather than by their visibility in SEM or XRD alone. Even a small fraction of sigma, chi, or chromium nitride can be damaging if it produces Cr-, Mo-, or N-depleted regions adjacent to precipitates, whereas the absence of obvious precipitates in SEM or XRD does not exclude nanoscale decomposition, clustering, or nitride formation. These inherited non-equilibrium features can modify transformation kinetics and precipitate distribution relative to wrought material; therefore, post-build thermal treatment should be interpreted as a controlled intervention in a metastable LPBF microstructure rather than as a direct transfer of wrought SDSS heat-treatment practice.

Table 2.3: Secondary phases and transformation risks relevant to LPBF-fabricated SDSS 2507. Adapted from [52, 53, 50, 51].

Feature	Temperature	Preferred location	Main consequence
Sigma phase (σ)	600–1000 °C	Ferrite–austenite interfaces; ferrite	Cr–Mo-rich intermetallic; reduces toughness and causes local Cr/Mo depletion.
Chi phase (χ)	700–900 °C	Ferrite; interfaces	Mo-rich intermetallic; may precede or accompany sigma and degrade corrosion response.
Chromium nitrides (Cr ₂ N, CrN)	Rapid cooling or 700–1000 °C exposure in N-supersaturated ferrite	Ferrite interiors; ferrite–ferrite and ferrite–austenite boundaries	Consume Cr and N locally; promote Cr-depleted regions and nitride-assisted pitting risk.
Secondary austenite (γ_2)	700–1000 °C	Ferrite regions; interfaces	May increase austenite fraction but can be Cr-, Mo-, or N-depleted.
Ferritic decomposition / α'	300–525 °C	Ferrite	Cr-rich/Fe-rich separation; increases hardness and embrittlement risk.
G phase / solute clustering	300–500 °C	Ferrite; α/α' regions	Long-term embrittlement risk; requires high-resolution confirmation.
Carbides	600–1050 °C	Grain and phase boundaries	Less dominant in low-carbon SDSS, but may contribute to local Cr depletion.

2.4 LPBF microstructural evolution

2.4.1 Process variables and solidification

LPBF fabricates metallic components by selectively melting successive powder layers using a focused laser beam. The resulting thermal history is governed by laser power, scan speed, hatch spacing, layer thickness, beam size, scan strategy, atmosphere, and thermal boundary conditions. In SDSS 2507, these variables simultaneously influence densification, ferrite/austenite balance, segregation, residual stress, defect formation, crystallographic texture, and the response to subsequent heat treatment [54, 12].

The LPBF thermal cycle differs fundamentally from conventional casting, welding, and wrought processing because melting, solidification, and reheating occur locally and repeatedly within overlapping melt pools. As a result, the final microstructure is controlled not only by the nominal parameter set but also by melt-pool geometry, local heat accumulation, powder-bed absorptivity, fluid flow, remelting depth, and the thermal history imposed by neighbouring scan tracks and subsequent layers [55, 56, 4]. This is particularly important for duplex and super duplex stainless steels because ferrite/austenite balance is sensitive to both cooling rate and solid-state diffusion time.

A recurring weakness in LPBF literature is the tendency to treat optimization as a density problem. For SDSS, this is insufficient. A parameter set that minimizes porosity may still produce an

unacceptable ferrite-rich microstructure, strong texture, residual strain, nitrogen redistribution, or precipitation susceptibility. The LPBF process window for SDSS is therefore constrained by metallurgy, not only by densification. Figure 2.4 summarizes the principal LPBF system components and process variables that define the thermal history.

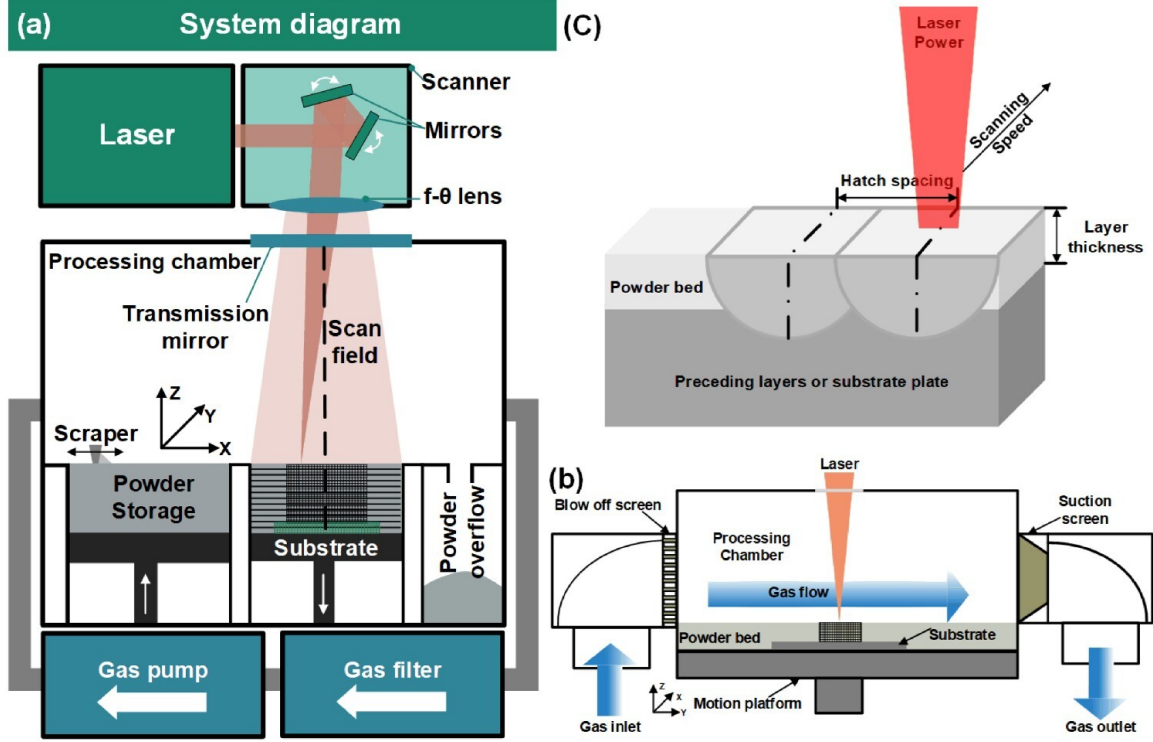


Figure 2.4: Schematic LPBF system showing the laser, optics, powder spreading mechanism, inert-gas flow, and principal process parameters governing melt-pool behavior. Adapted from [57].

Volumetric energy density is often used as a compact process descriptor,

$$E_v = \frac{P}{vht}, \quad (2.4)$$

where E_v is the volumetric energy density [J mm^{-3}], P is the laser power [$\text{W} = \text{J s}^{-1}$], v is the scan speed [mm s^{-1}], h is the hatch spacing [mm], and t is the layer thickness [mm]. Although E_v is useful for first-order process comparison, it is not a mechanistic predictor of LPBF microstructure. Different combinations of P , v , h , and t can produce the same nominal E_v but different melt-pool geometries, cooling rates, keyhole tendency, lack-of-fusion risk, thermal overlap, and phase-transformation conditions [58, 59, 60, 61]. For LPBF-fabricated SDSS, energy density must therefore be interpreted together with direct porosity, melt-pool, phase-fraction, and microstructural evidence. Figure 2.5 identifies the main process parameters commonly used to describe LPBF processing.

Classical solidification concepts remain useful for interpreting LPBF microstructure. The temperature gradient, G , and solidification growth rate, R , influence the transition among planar, cellular, columnar dendritic, and equiaxed dendritic morphologies. In solidification theory, G/R

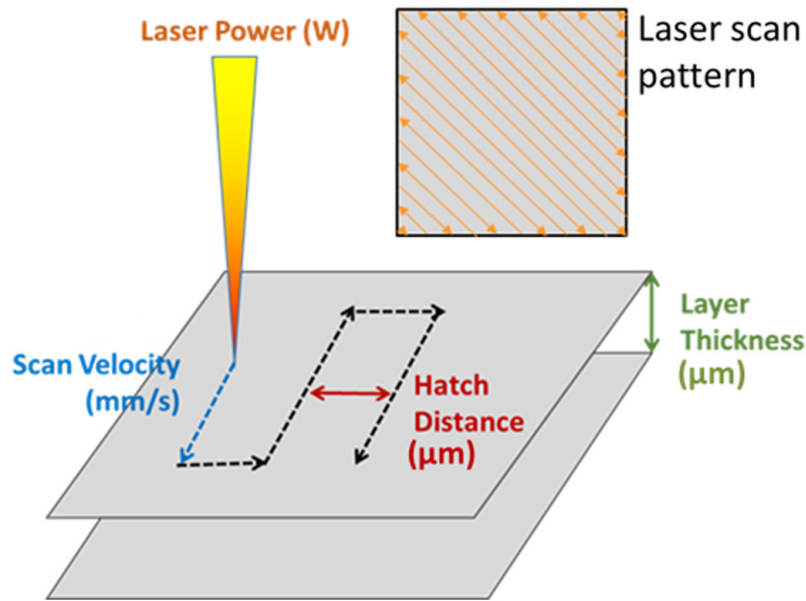


Figure 2.5: Main LPBF parameters: laser power, scan speed, hatch spacing, layer thickness, and scan strategy. Adapted from [62].

controls the stability of the solidification front, whereas GR is associated with cooling rate and microstructural refinement [63, 64]. LPBF imposes high G , high R , and rapid spatial changes in both quantities, producing heterogeneous solidification conditions within and between melt pools. This promotes fine cellular or columnar structures, chemical gradients at cellular boundaries, and epitaxial grain growth across layer boundaries when the crystallographic orientation of the underlying material is retained [4, 6]. Figure 2.6 schematically illustrates how G and R control solidification morphology and scale. In the schematic, S/L denotes the solid–liquid interface.

For SDSS 2507, the key consequence of rapid solidification is ferrite retention. Austenite formation is diffusion-controlled and therefore kinetically restricted during the short thermal cycles of LPBF. Austenite may subsequently form during reheating or post-build annealing, but its morphology, chemistry, and distribution depend on the inherited ferritic cellular structure and local segregation state. Consequently, LPBF-fabricated SDSS should be interpreted as a non-equilibrium ferrite-rich starting condition rather than as a directly miniaturized version of wrought duplex stainless steel.

2.4.2 As-built LPBF microstructures

A recurring observation across LPBF-processed duplex and super duplex stainless steels is the formation of a ferrite-rich as-built microstructure, often with fine cellular or subcellular morphology and limited austenite located at intercellular regions, melt-pool boundaries, or reheated zones [54, 12, 40]. This fundamentally differs from the balanced duplex microstructure expected in wrought SDSS. The ferrite-rich as-built state arises because the LPBF cooling path suppresses extensive ferrite-to-austenite transformation. Any austenite that forms during fabrication is expected to nucleate preferentially where thermodynamic driving force and diffusion paths are favourable, including ferrite grain boundaries, cellular boundaries, locally Ni- or N-enriched regions, and thermally reheated zones [3, 41, 42]. Therefore, the spatial distribution of austenite is as important

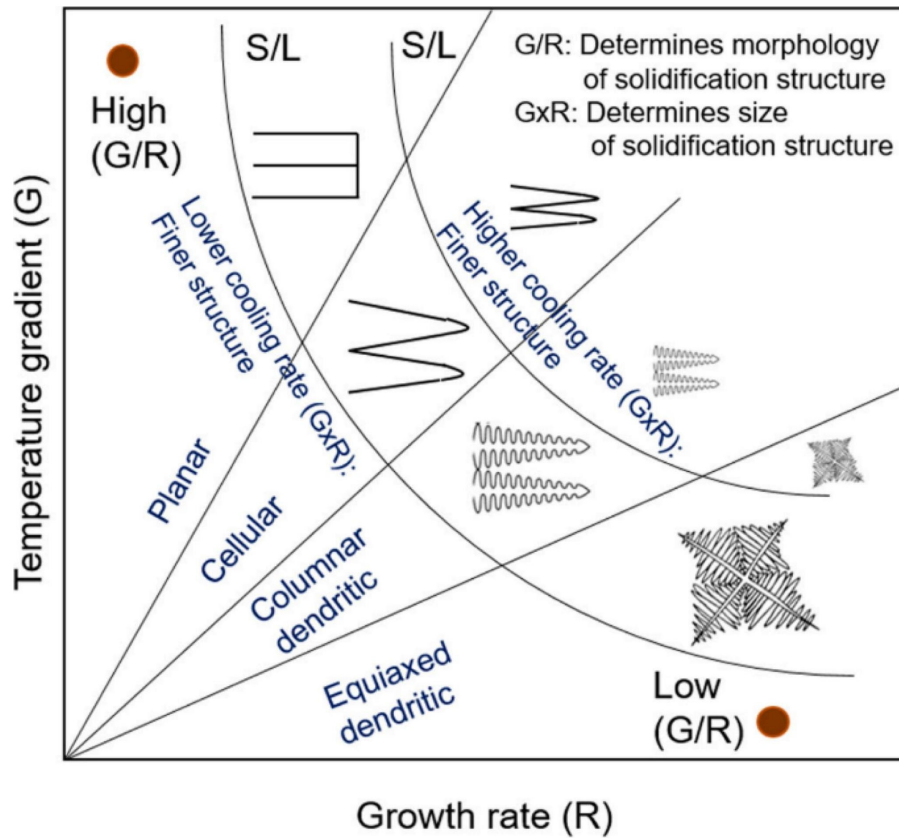


Figure 2.6: Schematic effect of temperature gradient G and solidification growth rate R on solidification morphology and microstructural scale. S/L denotes the solid–liquid interface. Adapted from [65].

as the phase fraction itself.

The as-built state may also contain elemental segregation, high dislocation density, residual stress, crystallographic texture, and nanoscale precipitates or precursors [40, 44]. These features should not be interpreted independently. Cellular ferrite affects segregation; segregation affects austenite nucleation and passive-film chemistry; residual strain affects both mechanical response and corrosion sensitivity; and defects influence both fracture and pit initiation. Thus, the as-built LPBF state is a coupled metallurgical condition rather than merely a phase-fraction problem.

Cellular and subcellular structures in LPBF steels often reflect solute redistribution during rapid solidification and subsequent dislocation arrangement during thermal contraction. In stainless steels, such cell boundaries can be enriched in substitutional elements and may act as preferred sites for phase transformation, precipitation, or localized corrosion depending on chemistry and heat treatment [6, 4]. For SDSS 2507, this means that cellular morphology is not merely a refinement feature; it is a potential pathway for austenite nucleation, chemical heterogeneity, and electrochemical localization.

Table 2.4 summarizes representative phase-fraction trends reported for LPBF-fabricated SDSS and shows why the as-built phase balance cannot be inferred from processing density alone.

Table 2.4: Representative phase-fraction trends in LPBF-fabricated SDSS. Adapted from [66].

Condition	Parameters	Ferrite (vol.%)	Austenite (vol.%)	Metallurgical implication
As-built	200 W / 400 mm/s	95	5	Highly ferritic; rapid cooling dominates.
As-built	350 W / 800 mm/s	50	45	More balanced but still sensitive to minor secondary phases.
Solution annealed	350 W / 800 mm/s	48	50	Duplex balance largely recovered.
As-built	500 W / 1200 mm/s	60	30	Coarser structure and increased secondary-phase risk.

2.4.3 Texture and anisotropy

LPBF/SLM is intrinsically directional. Build orientation and scan strategy influence thermal gradients, epitaxial grain growth, crystallographic texture, melt-pool overlap, phase distribution, and residual stress. In duplex stainless steels, these features can affect tensile behaviour, impact response, crack propagation, and corrosion response because they modify phase connectivity, boundary character, and local strain accommodation. Figure 2.7 illustrates orientation-dependent microstructures reported for SLM-processed DSS SAF 2205 in as-built and heat-treated conditions.

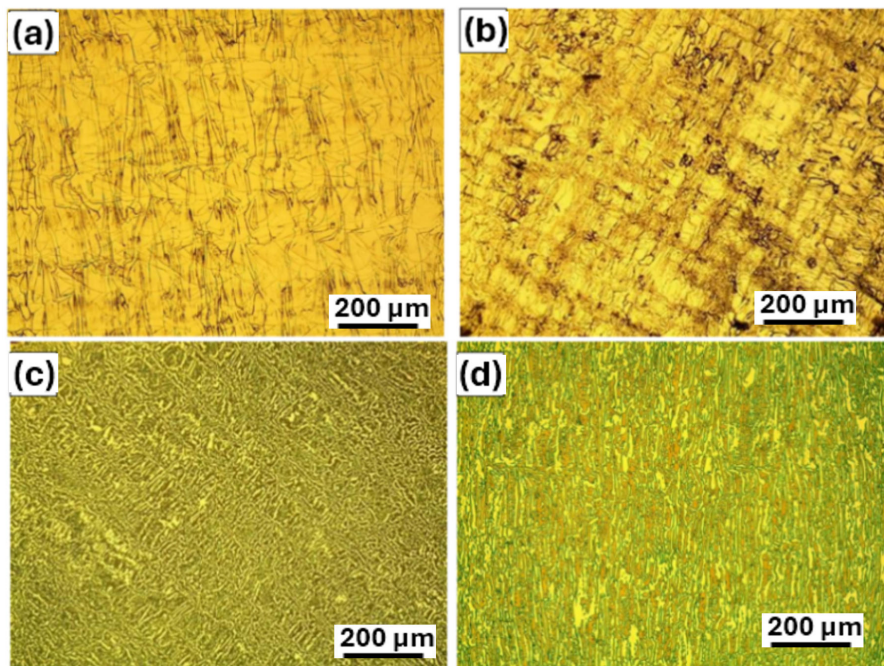


Figure 2.7: Representative optical micrographs of DSS 2205 processed by LPBF: (a) as-built, parallel to the build direction; (b) as-built, perpendicular to the build direction; (c) heat-treated, parallel to the build direction; and (d) heat-treated, perpendicular to the build direction. Adapted from [67].

Texture and anisotropy should not be treated as secondary mechanical artefacts. In LPBF/SLM duplex stainless steels, they are introduced during solidification and modified by cyclic reheating and post-build heat treatment. Therefore, build direction and scan strategy must be considered

when interpreting both mechanical and electrochemical datasets.

Epitaxial growth across remelted layer boundaries can promote columnar grain structures and preferred crystallographic orientations, especially where the local heat-flow direction remains persistent across layers [68, 56, 4]. Scan-vector rotation can reduce, but not necessarily eliminate, directional texture because local thermal gradients and melt-pool overlap remain anisotropic. In duplex stainless steels, such texture may also influence ferrite–austenite morphology and connectivity, thereby affecting anisotropy in tensile deformation, impact fracture, and corrosion propagation paths.

2.4.4 Defect inheritance

LPBF defects include lack-of-fusion pores, keyhole pores, gas porosity, balling-related discontinuities, cracks, delamination, distortion, and residual stress. Their significance in SDSS is broader than structural discontinuity: surface-connected or near-surface defects can act as stress concentrators and as preferential sites for passive-film breakdown. Residual stress and local strain heterogeneity may further increase scatter in both mechanical and electrochemical measurements. Figure 2.8 shows the typical processing regimes associated with dense material, lack of fusion, keyholing, and balling. Defect morphology is strongly linked to the local melt-pool regime. Lack-of-fusion defects generally arise when insufficient energy input or inadequate hatch overlap prevents complete melting and bonding between neighboring tracks or layers. Keyhole pores are associated with excessive energy density, vapor depression instability, and collapse of the keyhole cavity. Gas pores may be inherited from powder feedstock or trapped during solidification, while balling is associated with melt-track instability and poor wetting [58, 59, 60, 61]. These defect classes differ in morphology, location, and severity; therefore, total porosity alone is an incomplete descriptor of LPBF build quality.

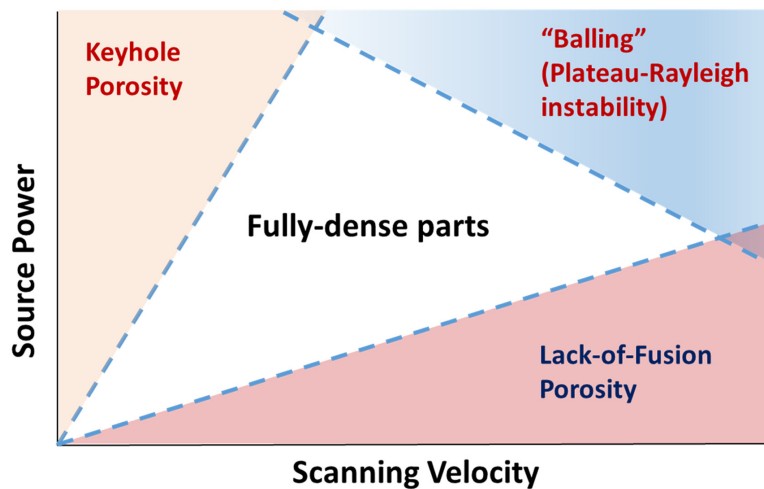


Figure 2.8: LPBF power–velocity process map showing dense, lack-of-fusion, keyhole, and balling regimes. Adapted from [62].

Figure 2.9 summarizes the origin and hierarchy of residual stresses in LPBF, which are relevant to distortion, cracking, phase stability, and corrosion sensitivity. Residual stress in LPBF originates

from steep thermal gradients, constrained thermal contraction, repeated reheating, and layer-wise accumulation of plastic strain [69, 70, 4]. In SDSS, this is not merely a dimensional-stability issue. Residual stress may modify cracking susceptibility, local misorientation, phase-boundary strain accommodation, and the electrochemical response of surface-connected defects. Stress-relief heat treatments must therefore be assessed not only by stress reduction but also by possible ferritic decomposition, nitride formation, or other thermally activated changes in the ferrite-rich LPBF state.

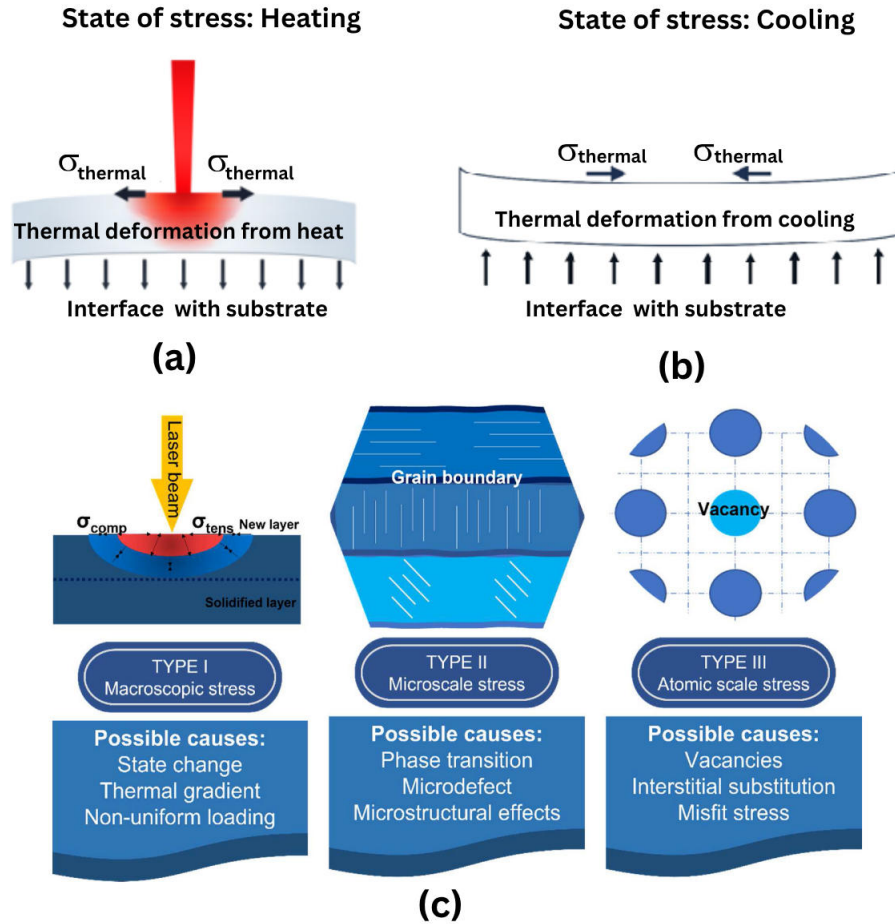


Figure 2.9: Residual-stress development in LPBF, including heating-induced compressive stress, cooling-induced tensile residual stress, and stress classification by length scale. Adapted from [71, 72].

2.5 Heat-treatment response

Post-build heat treatment is used to relieve residual stress, promote austenite formation, homogenize segregation, dissolve detrimental phases, and recover a more favorable duplex balance [12, 44]. However, LPBF-fabricated SDSS begins from a ferrite rich, defect-containing, chemically heterogeneous, and residually strained state; therefore, heat-treatment concepts derived from wrought or welded SDSS cannot be transferred uncritically.

The relevant transformation windows have already been summarized in Table 2.2; therefore, they are not repeated here. In LPBF-fabricated SDSS 2507, these windows must be interpreted with respect to the inherited as built condition. Low temperature stress relief may reduce residual

stress related instability but is unlikely to restore duplex phase balance. Intermediate temperature exposure may promote ferritic decomposition, solute clustering, chromium nitride precipitation, or early stage precipitation. High temperature solution annealing can promote austenite recovery, secondary phase dissolution, and chemical homogenization, but it may also reduce strength by removing LPBF inherited cellular refinement, residual strain, and high dislocation density.

This distinction is critical because LPBF changes both the starting phase constitution and the stored energy state relative to conventionally processed SDSS. Rapid solidification and repeated reheating can retain a high ferrite fraction, introduce cellular segregation, increase dislocation density, and generate residual stress. These inherited features can accelerate or localize transformations during post-build heating [4, 6, 2]. Consequently, the same nominal heat-treatment temperature may produce different responses in wrought, welded, and LPBF-fabricated SDSS.

The 300–550 °C range requires particular caution because ferrite rich duplex stainless steels are susceptible to low temperature ferritic instability. Exposure near the 475 °C embrittlement regime can promote α/α' decomposition, Cr-rich/Fe-rich separation, and nanoscale clustering without producing SEM-visible precipitates [46, 48, 47]. Therefore, an increase in hardness, a loss of toughness, or electrochemical degradation after stress relief should not be automatically assigned to the sigma phase unless sigma or chi is directly confirmed. Chromium nitride formation is also possible where nitrogen remains supersaturated in ferrite or where austenite formation is insufficient to accommodate nitrogen in solid solution [42, 3].

The 600–1000 °C range remains the classical high risk precipitation window for duplex and super duplex stainless steels. Sigma and chi phases are especially important because they are enriched in Cr and Mo and can generate adjacent depleted regions, reducing toughness and pitting resistance even when the precipitate fraction is small [2, 31, 32]. Secondary austenite may also form in this range, but its effect is not necessarily beneficial because its local Cr, Mo, and N content may differ from primary austenite. Solution annealing should therefore be understood as a phase rebalancing and homogenization treatment rather than a universal improvement. It can restore austenite and dissolve harmful phases when the temperature and holding time are appropriate, but it may also coarsen the microstructure and reduce strength. The heat treatment response of LPBF fabricated SDSS 2507 should consequently be interpreted as a balance among phase restoration, precipitation risk, defect relaxation, chemical homogenization, and strength retention.

2.6 Mechanical performance

Mechanical behavior in LPBF-fabricated DSS and SDSS is governed by phase fraction, phase morphology, grain size, texture, residual stress, dislocation density, porosity, and secondary phases. As-built LPBF states may exhibit high strength because of ferrite retention, cellular refinement, residual strain, and high dislocation density, but these same features may reduce ductility or impact toughness. Heat treatment commonly reduces strength while improving ductility and toughness, reflecting a trade-off between LPBF-inherited strengthening and duplex restoration [12, 66].

The mechanical response should therefore be interpreted as a coupled outcome of strengthening

and damage-tolerance mechanisms. Ferrite-rich LPBF microstructures can increase yield strength through phase constitution, fine cellular substructure, residual strain, and high dislocation density. However, ductility and impact toughness depend strongly on austenite fraction, austenite morphology, phase connectivity, defect population, and the presence or absence of brittle secondary phases [1, 26, 6]. Thus, high strength in the as-built state does not necessarily imply superior overall mechanical performance.

This distinction is important because LPBF mechanical properties are often controlled by the competition between microstructural strengthening and defect-limited failure. Fine cellular structures, high dislocation density, and residual strain can increase hardness and yield strength, whereas lack-of-fusion defects, keyhole pores, surface roughness, residual tensile stress, and brittle precipitates can reduce elongation, impact energy, and fracture resistance [23, 73, 74, 5]. Consequently, the same LPBF condition may appear favorable in hardness or yield strength but unfavorable in tensile ductility or impact toughness.

Figure 2.10 places conventional duplex stainless steel families in a yield-strength and PREN framework. This comparison is useful as a wrought-material benchmark because it shows why SDSS grades are attractive for chloride-containing environments: they combine high nominal pitting resistance with high strength. However, the figure should not be interpreted as an LPBF-specific property map, because LPBF introduces non-equilibrium solidification, ferrite retention, residual strain, texture, porosity, and possible chemical segregation that can shift both mechanical and corrosion response away from conventional property envelopes. To complement the conventional PREN–yield-strength benchmark in Figure 2.10, Table 2.5 summarizes representative literature data for LPBF/PBF-LB-manufactured DSS and SDSS. The table highlights that LPBF materials can exhibit high strength, but their performance depends strongly on phase balance, heat treatment, residual strain, defect state, and corrosion-relevant local chemistry rather than on nominal PREN alone.

The comparison in Table 2.5 shows that LPBF-manufactured DSS/SDSS does not follow the conventional PREN–yield-strength landscape in a simple way. High strength is often obtained in the as-built state because of ferrite retention, cellular refinement, residual strain, and high dislocation density, whereas heat treatment can improve phase balance and ductility at the expense of strength. Therefore, LPBF DSS/SDSS must be interpreted using a processing–microstructure–property framework rather than by transferring conventional wrought-material PREN–strength relationships directly. A critical interpretation of mechanical performance must also avoid reducing strength to the ferrite fraction alone. Porosity, lack-of-fusion defects, crystallographic texture, dislocation density, residual stress, and phase connectivity can each dominate the measured response depending on specimen orientation and test mode. Impact toughness is particularly sensitive to brittle phases, residual stress, and defects, so it cannot be inferred from tensile strength or hardness alone.

Build orientation and scan strategy are relevant because LPBF produces direction-dependent melt-pool overlap, columnar grain growth, texture, residual stress, and defect alignment. These features

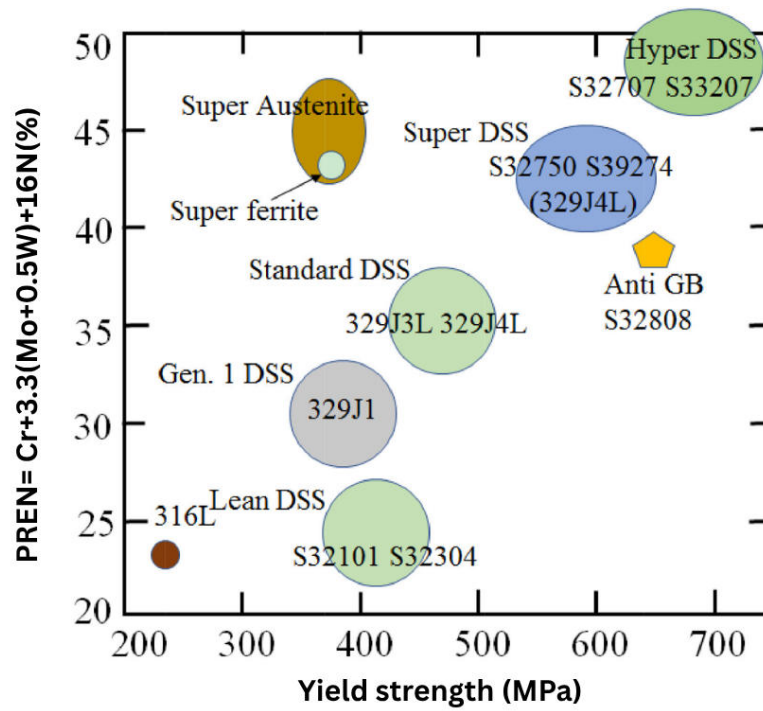


Figure 2.10: PREN versus yield-strength landscape for conventional duplex stainless steel families. The figure is used as a wrought-material benchmark and should not be interpreted as an LPBF-specific property map. Adapted from [75].

can generate anisotropy in tensile strength, elongation, and fracture path even when the nominal composition and process parameters are unchanged [68, 56, 4]. In duplex and super duplex stainless steels, anisotropy may be amplified further by phase connectivity because ferrite and austenite differ in strength, strain accommodation, and crack-arrest capacity. Defects are especially important in LPBF mechanical testing because different defect types have different mechanical consequences. Lack-of-fusion defects are generally more damaging than small spherical gas pores because they are irregular, sharp, and often aligned with melt-pool or layer boundaries. They can reduce ductility, promote premature fracture, and increase scatter in tensile and impact data [58, 59, 60, 61]. Thus, density or porosity fraction alone is insufficient; defect morphology, size, orientation, and location must also be considered.

Heat treatment changes the balance between strength and damage tolerance. Solution annealing can reduce strength by lowering dislocation density, relieving residual strain, coarsening the microstructure, and weakening LPBF-inherited cellular strengthening. At the same time, it can improve ductility and impact response by increasing austenite fraction, reducing residual stress, dissolving brittle precipitates where present, and improving phase balance [2, 3, 4]. Stress-relief treatments may produce smaller changes in tensile strength, but their effect on toughness can be substantial if ferritic decomposition, nitride precipitation, or local embrittlement occurs.

For SDSS 2507, the mechanical penalty from secondary phases is especially severe because sigma, chi, nitrides, and ferritic decomposition products can embrittle the ferritic matrix or phase boundaries while also altering local chemistry. This means that hardness increases after intermediate-

Table 2.5: Representative literature data for LPBF/PBF-LB-manufactured DSS and SDSS.

Study	Alloy	Condition	Structure / density	Mechanical response	Corrosion response
Zhu et al. [12]	2507 DSS, LPBF	As-built	Ferritic LPBF microstructure; relative density > 98%; hardness > 400 HV	YS = 1198 MPa; UTS = 1269 MPa	$i_{\text{corr}} \approx 10^{-6} \text{ A cm}^{-2}$; passive window $\sim 1.2 \text{ V}$; response comparable to rolled 2507.
Xie et al. [76]	SAF 2507 DSS, LPBF	As-built	Density = 98.71%; island-like structures reported	Strength = 1173 MPa; elongation = 18%	Electrochemical behaviour better than cast material.
Xie et al. [10]	SAF 2507 DSS, LPBF	Solution-treated at 1100 °C	Austenite recovery after solution treatment	Strength = 860 MPa; elongation = 45%	Similar electrochemical behaviour to LPBF as-built condition and better than cast material.
Davidson et al. [66]	SDSS, PBF-LB	As-built; different build orientations	Predominantly ferritic microstructure; build orientation affects Widmanstätten γ -austenite formation, texture, and local chemistry	Higher hardness and strength than solution-annealed states	Corrosion response affected by build orientation, phase morphology, texture, residual stress, and chemical homogeneity.
Davidson et al. [66]	SDSS, PBF-LB	Solution-annealed	Increased γ -austenite content and weakened retained texture	Lower tensile strength but significantly increased elongation	Improved pitting response attributed to stress removal and more equilibrated microstructure.
Haghdadi et al. / Laleh et al. [77, 78]	2205 DSS, LPBF	As-built and heat-treated	As-built state commonly ferrite-rich; heat treatment promotes austenite recovery	High as-built strength but strength–ductility trade-off after heat treatment	Corrosion cannot be inferred from nominal PREN alone because phase imbalance, nitrogen redistribution, nitrides, and local chemistry affect pitting/repassivation.

temperature exposure should not be interpreted automatically as beneficial strengthening. They may indicate ferritic decomposition or precipitation-induced embrittlement, particularly if accompanied by reduced elongation, lower impact energy, or more brittle fracture morphology [31, 32, 46, 47].

2.7 Corrosion and electrochemical response

Corrosion performance in SDSS 2507 depends on passive-film stability, phase balance, phase-specific chemistry, secondary phases, residual stress, and defects. Nominally high PREN does not guarantee high pitting resistance if Cr or Mo is locally depleted near precipitates or if surface-

connected defects provide preferential pit-initiation sites [79, 44]. This limitation is particularly important for LPBF-fabricated SDSS because rapid solidification, ferrite retention, cellular segregation, residual strain, and defect inheritance can produce local corrosion responses that differ from those expected from nominal composition alone. Electrochemical results were interpreted by separating the information provided by OCP, CPP, Tafel analysis, and EIS. Open-circuit potential reflects the time-dependent mixed potential and was therefore used to track potential stabilization rather than as a direct corrosion-resistance metric. Cyclic potentiodynamic polarization provided information on breakdown and repassivation behaviour, with interpretation linked to hysteresis, scan rate, surface preparation, and metastable-pitting response. Tafel-derived parameters were used comparatively because activation-controlled regions can be limited in passive alloys. EIS equivalent-circuit fitting was used to describe the dominant impedance response, with fitted parameters assessed together with OCP and CPP evidence.

Table 2.6: Representative electrochemical response of LPBF/PBF-LB-manufactured DSS and SDSS.

Study	Alloy / condition	Evidence	Finding
Zhu et al. [12]	SDSS 2507, LPBF; as-built	$i_{\text{corr}} \sim 10^{-6} \text{ A cm}^{-2}$; passive window $\sim 1.2 \text{ V}$	Dense LPBF 2507 showed corrosion behavior comparable to rolled 2507 despite a ferritic LPBF microstructure.
Xie et al. [10]	SAF 2507, LPBF; as-built and 1100 °C solution-treated	Potentiodynamic comparison with cast 2507	LPBF and solution-treated specimens performed better than cast material; phase fraction alone did not control corrosion resistance.
Davidson et al. [66]	SDSS, PBF-LB; as-built, different build orientations	Pitting response compared across build orientations	Corrosion response depended on build orientation, phase morphology, texture, residual stress, and chemical homogeneity.
Davidson et al. [66]	SDSS, PBF-LB; solution-annealed	Pitting response compared with as-built material	Solution annealing improved pitting response through stress relief and a more equilibrated duplex microstructure.
Laleh et al. [80]	2205 DSS, LPBF; as-built and post-processed	Repassivation and localized-corrosion assessment	Poor as-built repassivation was linked to non-equilibrium phase composition and elemental distribution; post-processing improved resistance.
Gandhi et al. [11]	SDSS 2507, PBF-LB; solution-annealed	Thermodynamic and microstructural assessment	Solution annealing supported ferrite/austenite rebalancing, nitride dissolution, and austenite recovery.

The studies summarized in Table 2.6 show that corrosion resistance in LPBF/PBF-LB-manufactured DSS and SDSS cannot be inferred from nominal PREN, bulk ferrite/austenite fraction, or electrochemical parameters alone. Processing condition, build orientation, residual strain, local chemical partitioning, nitride or intermetallic precipitation, and surface-connected defects can influence breakdown, repassivation, and apparent corrosion-current response. Electrochemical results should therefore be interpreted together with phase analysis.

Pitting corrosion involves local passive-film breakdown, pit stabilization, hydrolysis-driven acidification, chloride enrichment, and autocatalytic dissolution. In duplex stainless steels, pit initiation may occur preferentially at inclusions, ferrite/austenite interfaces, chromium-depleted regions, nitrides,

intermetallics, or surface-connected LPBF defects. Figure 2.11 illustrates this chloride-assisted breakdown sequence. Cl^- ions first adsorb and accumulate at weak regions of the passive film, including defects, inclusions, phase boundaries, chemically depleted zones, or surface-connected pores. This adsorption can destabilize the Cr-rich oxide/hydroxide film by increasing local defect concentration, promoting cation-vacancy generation, and accelerating local metal-ion dissolution [24, 81].

As vacancies accumulate and condense at the metal/film interface, local film detachment or thinning may occur. Once the passive film ruptures, the exposed metal becomes anodically active and a metastable pit may form. After film rupture, dissolved metal cations hydrolyze inside the confined pit environment, producing local acidification. Chloride ions migrate into the pit to maintain electroneutrality, which further increases the local chloride concentration and sustains anodic dissolution. This creates an autocatalytic pit chemistry characterized by low pH, high Cl^- concentration, and rapid metal dissolution [24]. If the passive film reforms before this local chemistry stabilizes, the pit repassivates; otherwise, the metastable pit develops into a stable growing pit. For LPBF-fabricated SDSS, this mechanism is particularly relevant because ferrite/austenite interfaces, Cr- or Mo-depleted regions, nitrides, intermetallics, lack-of-fusion defects, pores, and melt-pool-related surface features can provide preferential sites for chloride-assisted passive-film breakdown.

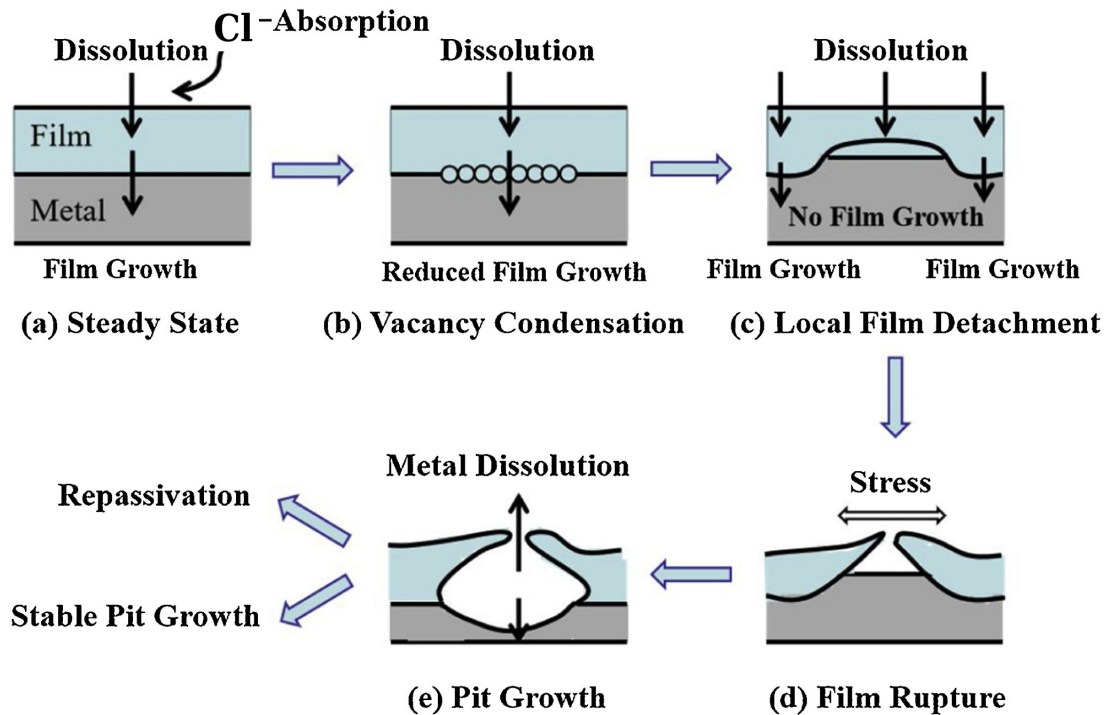


Figure 2.11: Schematic representation of chloride-assisted passive-film instability and pit-nucleation stages, including Cl^- adsorption, vacancy condensation, local film detachment, rupture, metastable pit formation, and pit growth. Adapted from [81].

Corrosion and mechanical performance are coupled because both are governed by the inherited LPBF phase–defect state. Ferrite/austenite balance, phase morphology, segregation, residual

strain, porosity, texture, and precipitation influence strength, ductility, impact toughness, passive-film stability, and pit-initiation susceptibility. A condition that increases strength through ferrite retention, cellular refinement, or high dislocation density may reduce toughness or corrosion resistance if it also retains sharp defects, residual tensile stress, chemical heterogeneity, or unstable ferrite [4, 6, 23]. Conversely, solution annealing may improve ductility, phase balance, and passive-film stability while reducing strength by removing LPBF-inherited strengthening features. This coupling is especially important for SDSS 2507 because mechanical degradation and localized corrosion can originate from the same microstructural features. Lack-of-fusion defects can act as fracture initiators and pit-initiation sites; Cr–Mo-rich intermetallics can reduce toughness while producing adjacent Cr/Mo-depleted regions; and ferritic decomposition can increase hardness while reducing damage tolerance [58, 59, 31, 32, 46]. Therefore, strength, ductility, impact toughness, passivity, and pitting resistance should be interpreted as coupled outcomes rather than independent performance metrics. The relevant question is not whether LPBF-fabricated SDSS is mechanically strong or corrosion resistant in isolation, but which microstructural state provides the best compromise among strength, ductility, impact toughness, passivity, and localized corrosion resistance.

2.8 Data-driven corrosion screening

Data-driven corrosion screening can serve as a supplementary tool for relating alloy composition, processing history, environment, and electrochemical response. Machine learning (ML) has been applied to corrosion-rate prediction, pitting descriptors, corrosion potential, pitting potential, critical pitting temperature (CPT), and impedance-based passivity classification [19, 82, 83, 84, 20, 21, 85]. These applications show that ML can support corrosion screening, trend organization, and consistency assessment, provided that the predicted target, input descriptors, and applicability domain are clearly defined.

Corrosion-informatics models require a clearly defined electrochemical target because corrosion current density, polarization resistance, pitting potential, repassivation potential, and charge transfer resistance describe different aspects of the corrosion response. A high polarization resistance, for example, may indicate low near-open-circuit dissolution, but it does not by itself demonstrate high pitting resistance or stable repassivation. Model outputs must therefore be interpreted only with respect to the predicted quantity. This target-specific approach is supported by recent studies: Pei et al. predicted atmospheric corrosion from environmental descriptors [82]; Jiang et al. modeled time-dependent corrosion-potential evolution using long short-term memory (LSTM) networks [83]; Coelho et al. estimated pitting descriptors from localized polarization data [84]; Qiao et al. predicted pitting potential in chloride-containing environments [20]; and Hou et al. developed interpretable descriptors for critical pitting temperature (CPT) [21]. Together, these studies support corrosion screening when the target variable, input descriptors, and applicability domain are explicitly controlled.

This requirement is critical for LPBF-fabricated SDSS 2507. Nominal composition alone does not

define corrosion behavior because LPBF introduces non-equilibrium phase balance, segregation, residual strain, texture, defects, and surface-condition sensitivity. Post-build heat treatment further modifies ferrite/austenite balance, Cr–Mo–N partitioning, and precipitation risk. Descriptors such as pitting resistance equivalent number (PREN), chromium equivalent, nickel equivalent, heat-treatment temperature, holding time, and composition processing interaction terms must therefore be interpreted together with measured phase constitution and electrochemical data.

The main limitation is the applicability domain. Literature-derived datasets often combine different surface preparations, scan rates, electrolytes, reference electrodes, exposure histories, and reporting conventions. In passive alloys, this is especially important because cyclic potentiodynamic polarization (CPP) reflects breakdown, hysteresis, and repassivation tendencies, whereas electrochemical impedance spectroscopy (EIS) reflects impedance near the selected potential and depends on the equivalent-circuit model.

Therefore, for LPBF-fabricated SDSS 2507, ML assisted corrosion screening should be treated as a bounded support layer. It can compare alloy processing heat treatment trends with measured OCP, CPP, EIS, and microstructural evidence, and it can test whether predicted trends are consistent with the experimental response. Mechanistic interpretation and corrosion-resistance claims, however, must remain anchored in direct evidence. Confirmation of service-relevant resistance would require post-test morphology, pit-depth statistics, pit-density analysis, or critical pitting temperature measurements.

2.9 Methodological constraints

The interpretation of structure, properties, and corrosion response in this thesis was strengthened by explicitly considering the operating limits of each characterization and testing method. X-ray diffraction (XRD) provided bulk phase-identification and comparative peak-broadening information; however, phase quantification and line-broadening analysis can be influenced by texture, overlapping ferrite/austenite reflections, residual strain, and instrumental broadening. Williamson–Hall analysis was therefore used as a comparative descriptor of peak broadening rather than as an absolute dislocation-density measurement, unless supported by rigorous instrumental correction, peak selection, peak-shape control, and strain-anisotropy treatment [86, 87].

Electron backscatter diffraction (EBSD) provided phase maps, grain morphology, texture, boundary statistics, and local orientation-spread descriptors. Kernel average misorientation (KAM), grain orientation spread (GOS), and related metrics were interpreted as comparative indicators of local orientation heterogeneity, not as direct residual-stress or stored-energy measurements without independent calibration [88, 89]. Scanning electron microscopy/energy-dispersive X-ray spectroscopy (SEM/EDS) supported morphology and micron-scale chemical assessment, while transmission electron microscopy (TEM) enabled selected nanoscale examination. Because EDS has limited sensitivity for nitrogen and TEM is spatially local, these methods were interpreted as complementary evidence streams rather than isolated proof of bulk-scale behaviour [90, 91].

The electrochemical assessment was also interpreted using a convergent-evidence approach. Open-

circuit potential (OCP) was treated as a time-dependent mixed potential; cyclic potentiodynamic polarization (CPP) was used to compare breakdown and repassivation behaviour; and electrochemical impedance spectroscopy (EIS) provided fitted impedance descriptors near the tested condition. Tafel-derived quantities and equivalent-circuit parameters were therefore used comparatively, recognizing their sensitivity to surface condition, scan rate, electrolyte history, pit stabilization, and model non-uniqueness [24, 25, 92, 93].

Consequently, this thesis does not rely on single-technique inference. Instead, phase constitution, defect state, local orientation heterogeneity, mechanical response, CPP behaviour, EIS response, and descriptor-level corrosion screening were interpreted together. The data-driven corrosion-screening workflow was used as a supplementary tool to organize composition–processing–electrochemical trends, while the primary corrosion interpretation remained anchored in the measured electrochemical response and supporting microstructural evidence.

2.10 Literature evidence matrix

Table 2.7 summarizes the main evidence patterns in LPBF DSS/SDSS research and identifies the unresolved issues that motivate this dissertation.

The evidence matrix leads to three main conclusions. First, LPBF-fabricated DSS/SDSS cannot be interpreted solely by density or nominal composition; the relevant material state is defined by the coupled evolution of phase balance, defect population, residual strain, texture, segregation, and precipitation sensitivity. Second, mechanical and corrosion performance must be treated as linked outcomes of this inherited phase–defect state rather than as independent property categories. A condition that improves strength through ferrite retention, cellular refinement, or residual strain may not necessarily improve ductility, impact toughness, repassivation ability, or pitting resistance. Third, the combined role of Ni-assisted austenite stabilization and nitrogen-related phase chemistry remains insufficiently resolved under LPBF thermal histories. These conclusions justify the comparative framework of the present thesis, in which standard SDSS 2507 and Ni-modified variants are assessed through integrated microstructural, mechanical, and electrochemical characterization.

2.11 Critical synthesis

The literature shows that laser powder bed fusion (LPBF)-fabricated duplex stainless steels (DSS) and super duplex stainless steels (SDSS) commonly develop ferrite-rich, non-equilibrium microstructures because rapid solidification and short thermal cycles restrict diffusion-controlled austenite formation [54, 12, 4, 6]. Post-build heat treatment can promote austenite recovery, residual-strain reduction, solute redistribution, and secondary-phase dissolution where thermodynamics and kinetics permit [44, 2, 3]. However, these benefits are coupled to competing risks. Solution annealing may reduce LPBF-inherited strengthening from cellular refinement, residual strain, and high dislocation density, whereas stress relief may enter temperature ranges associated with ferritic decomposition, solute clustering, chromium nitride precipitation, or early-stage intermetallic formation. The heat-treatment response of LPBF-fabricated SDSS must therefore

Table 2.7: Critical evidence matrix for LPBF DSS/SDSS metallurgy, corrosion assessment, and data-driven corrosion screening.

Study focus	What is established	Main limitation	Relevance to this thesis
Process parameters	Power, scan speed, hatch spacing, layer thickness, and scan strategy control density, melt-pool stability, texture, and phase constitution [4, 6].	Density is often overemphasized relative to duplex balance, residual strain, segregation, and precipitation sensitivity.	Supports treating LPBF as a metallurgically constrained process, not only a densification route.
Heat treatment	Annealing can restore austenite, reduce residual strain, and dissolve secondary phases where thermodynamically and kinetically possible; stress relief may improve dimensional and residual-stress stability [2, 31, 32].	Wrought heat-treatment logic is often transferred to LPBF without accounting for the ferrite-rich, defect-rich, chemically heterogeneous starting state.	Justifies comparing stress relief and solution annealing across standard and Ni-modified alloys.
Mechanical response	Strength, ductility, hardness, and impact toughness depend on phase balance, defects, texture, residual stress, and heat treatment [23, 73].	Mechanical properties are often interpreted without sufficient linkage to corrosion-relevant microstructure and defect morphology.	Supports combined interpretation of strength, ductility, impact energy, fracture morphology, and microstructural state.
Corrosion response	CPE, OCP, and EIS reveal strong sensitivity to phase balance, precipitates, surface condition, passive-film stability, and defects [24, 25, 93].	Electrochemical metrics are often reported without pit morphology, phase-specific attack, or defect correlation.	Supports linking electrochemical behavior to microstructure.
Characterization limits	XRD, EBSD, SEM/EDS, TEM, and EIS provide complementary but method-limited descriptions of phase constitution, defect state, residual strain, local chemistry, and electrochemical response [86, 88, 93].	Single-technique interpretation can overstate certainty, especially for nanoscale precipitation, local chemistry, residual stress, and equivalent-circuit fitting.	Supports convergent interpretation using multiple microstructural, mechanical, and electrochemical datasets.
Ni/N modification	Ni and N can promote austenite stabilization and influence corrosion resistance through phase balance and phase-specific chemistry [1, 3, 2].	Their combined effect under LPBF rapid solidification and post-build heat treatment remains insufficiently resolved.	Defines the central alloy-design question for the +3 wt.% Ni and +6 wt.% Ni variants.
Data-driven corrosion screening	Machine-learning models can relate alloy chemistry, processing route, heat treatment, environment, and electrochemical metrics in large feature spaces [19, 82, 20, 21].	Model interpretation depends on target definition, training-data coverage, protocol consistency, and applicability domain [19, 84, 85].	Supports using corrosion informatics as a supplementary screening layer rather than as independent proof of corrosion mechanism [19, 83, 21].

be interpreted as a balance among phase restoration, precipitation risk, defect relaxation, and strength retention.

Predictive control of LPBF-fabricated SDSS 2507 remains limited by five unresolved issues. First, ferrite/austenite balance, porosity, texture, residual strain, segregation, and precipitation are not consistently treated as coupled consequences of the LPBF thermal history. Second, mechanical and corrosion responses are often interpreted separately, although both are governed by the inherited phase–defect state, including phase connectivity, local chemistry, residual stress, and defect morphology [23, 24, 93]. Third, heat-treatment practices are often transferred from wrought or welded SDSS without sufficient allowance for the ferrite-rich, chemically heterogeneous, defect-containing, and residually strained LPBF starting condition [1, 3, 31]. Fourth, the role of deliberate nickel (Ni) enrichment remains insufficiently resolved: Ni can promote austenite stabilization, but it may also alter strength, phase connectivity, chromium–molybdenum–nitrogen partitioning, precipitation response, and passive-film behavior. Fifth, data-driven corrosion screening remains insufficiently integrated with LPBF SDSS metallurgy. Machine-learning models can compare composition, processing, and electrochemical trends, but their outputs require assessment against phase balance, defect state, passive-film stability, and experimental corrosion measurements.

These unresolved issues define the contribution of this dissertation. The comparative alloy design examines standard UNS S32750/SDSS 2507 together with +3 wt.% Ni and +6 wt.% Ni variants under a common LPBF processing basis and selected post-build heat treatments. This design tests whether Ni-assisted austenite stabilization can improve the balance among phase constitution, mechanical performance, and corrosion response without unacceptable trade-offs in defect sensitivity, precipitation behavior, strength–ductility balance, toughness, or localized-corrosion resistance. Table 2.8 summarizes how the main unresolved literature gaps are translated into the methodological response used in this thesis.

Table 2.8: Literature-gap-to-methodology mapping for the present thesis.

Unresolved literature issue	Methodological response in this thesis
Ferrite-rich LPBF starting state	SEM, XRD, and EBSD were used to evaluate phase constitution, phase distribution, and LPBF-inherited microstructural features.
Heat-treatment response of LPBF SDSS	Stress-relief and solution-annealing treatments were compared to assess phase recovery, residual-strain indicators, and precipitation-sensitive features.
Effect of deliberate Ni enrichment	Standard SDSS 2507, +3 wt.% Ni, and +6 wt.% Ni variants were processed under a common LPBF basis to evaluate Ni-assisted austenite stabilisation.
Nanoscale precipitation uncertainty	Targeted TEM, STEM, SAED, and STEM-EDS were used where bulk and mesoscopic methods could not resolve nanoscale features.
Mechanical–corrosion coupling	Hardness, tensile, impact, fracture-surface SEM, OCP, CP, EIS, and corrosion observations were linked to the same phase–defect state.
Descriptor-level corrosion screening	A bounded data-driven workflow was used to compare composition–processing–electrochemical trends with measured corrosion behaviour.

This synthesis provides the basis for the research questions, hypotheses, and objectives in Chapter 1 by showing that LPBF processing, Ni modification, post-build heat treatment, characterization, mechanical testing, electrochemical evaluation, and data-driven corrosion screening are connected parts of one metallurgical framework. Accordingly, the methodology in Chapter 3 targets the main unresolved issues identified in the literature. Within this framework, SEM, XRD, EBSD, TEM, mechanical testing, CPP, and EIS provide complementary evidence for comparing phase constitution, microstructural morphology, crystallographic state, selected nanoscale features, mechanical response, fracture behavior, passive-film response, repassivation tendency, and charge-transfer resistance across the investigated alloy–heat-treatment conditions. Further work using atom-probe tomography, quantitative pit-depth and pit-density statistics, critical pitting temperature testing, phase-specific electrochemistry, stress-corrosion cracking tests, fatigue tests, rough-surface exposure, and long-term immersion or aging would extend the present analysis toward nanoscale solute partitioning, phase-specific dissolution, service localized-corrosion resistance, and long-term degradation behavior.

Chapter 3

Research Methodology

3.1 Feedstock and alloy design

Three powder feedstocks were investigated to assess composition-dependent responses in laser powder bed fusion (LPBF)-fabricated super duplex stainless steel (SDSS) 2507. The standard feedstock was a gas-atomized 2507-type SDSS powder derived from UNS S32750/EN 1.4410 chemistry and supplied by Sandvik Osprey Ltd. The modified feedstocks were produced by adding nominal +3 wt.% nickel (Ni) and +6 wt.% Ni to the standard alloy system. The Ni-modified variants were treated as experimental alloy modifications rather than as commercial UNS S32750 equivalents because the Ni addition alters the chromium/nickel (Cr/Ni) equivalent balance and may affect phase stability, chemical partitioning, and corrosion-relevant local chemistry.

The Ni-modified powders were prepared by blending the standard SDSS 2507 powder with pure Ni powder using a Turbula mixer for 20 min. This duration was selected to promote Ni distribution while limiting handling-induced powder degradation and fine-particle agglomeration. Because blended powder systems may retain local compositional heterogeneity, chemical and microstructural characterization were required to assess homogenization after LPBF. The blended powders were therefore treated as compositionally modified feedstocks, with final homogenization achieved through melting, mixing, and solidification during LPBF.

Table 3.1: Nominal alloy chemistry of standard and Ni-modified SDSS 2507 powders.

Alloy	Fe	Cr	Ni	Mo	Mn	Si	N	Cu	C	P	S	O
Standard	Bal.	25.0	6.3	3.8	0.4	0.35	0.30	0.15	0.016	0.024	0.001	0.017
+3 wt.% Ni	Bal.	24.4	9.1	3.7	0.4	0.29	0.30	0.14	0.016	0.023	0.001	0.016
+6 wt.% Ni	Bal.	23.7	11.9	3.6	0.4	0.28	0.30	0.14	0.015	0.023	0.001	0.016

Table 3.1 summarizes the nominal chemical compositions of the investigated powders. The progressive increase in Ni provides the intended austenite-stabilizing variable, while the concurrent changes in Cr, Mo, and N must be considered when interpreting phase stability and corrosion response.

3.2 LPBF processing

Specimens were fabricated by laser powder bed fusion (LPBF) using an AM125 system (Renishaw, UK) equipped with a Yb-fiber laser operating at a wavelength of $\lambda = 1.074 \mu\text{m}$ and a maximum laser power of $P_{\text{max}} = 200 \text{ W}$. Processing was conducted under an argon atmosphere to limit oxidation; the build chamber was purged and maintained at oxygen levels below 50 ppm during fabrication. Square coupons with plan dimensions of $1.0 \times 1.0 \text{ cm}$ were produced for metallography, X-ray diffraction (XRD), hardness, porosity, and electrochemical testing. Separate horizontally oriented

blanks were fabricated for tensile and Charpy impact testing because the square coupons were not geometrically compatible with these specimen geometries. A meander scanning strategy with 67° interlayer rotation was used to reduce directional heat accumulation and texture reinforcement. The selected processing parameters were laser power $P = 180$ W, scan speed $v = 300$ mm/s, hatch spacing $h = 120$ μ m, and layer thickness $t = 30$ μ m, corresponding to a volumetric energy density of $V_{ED} = 166.7$ J/mm³.

Before applying the alloy-specific heat-treatment matrix, LPBF parameters were screened using volumetric energy density as a first-order descriptor of energy input. The process matrix in Table 3.2 spans a broad energy-density range and includes both preheated and non-preheated builds. Because multiple variables differ between these conditions, the matrix is interpreted as process-window screening rather than as a controlled assessment of base-plate preheating.

Table 3.2: LPBF process-parameter matrix used to screen laser power and scan speed through calculated volumetric energy density.

$v = 300$ mm/s, $t = 30$ μ m, $h = 100$ μ m, preheating to 150 °C						
Sample ID	P ₁	P ₂	P ₃	P ₄	P ₅	P ₆
P (W)	200	190	180	170	160	150
V_{ED} (J/mm ³)	222.22	211.11	200.00	188.89	177.78	166.67
$P = 180$ W, $t = 30$ μ m, $h = 120$ μ m, no preheating						
Sample ID	A ₁	A ₂	A ₃	A ₄	–	–
v (mm/s)	250	300	375	500	–	–
V_{ED} (J/mm ³)	200.00	166.67	133.33	100.00	–	–
Sample ID	B ₁	B ₂	B ₃	B ₄	–	–
v (mm/s)	391	470	587	783	–	–
V_{ED} (J/mm ³)	127.66	106.38	85.11	63.83	–	–
Sample ID	C ₁	C ₂	C ₃	C ₄	–	–
v (mm/s)	525	630	787	1050	–	–
V_{ED} (J/mm ³)	95.24	79.37	63.49	47.62	–	–

The volumetric energy density was used as a first-order descriptor of the applied LPBF energy input and was calculated as [94]

$$V_{ED} = \frac{P}{vht}, \quad (3.1)$$

where V_{ED} is the volumetric energy density [J mm^{−3}], P is the laser power [W], v is the scan speed [mm s^{−1}], h is the hatch spacing [mm], and t is the layer thickness [mm]. The calculated E_v values for the investigated parameter sets are listed in Table 3.2. Because V_{ED} combines several independent process variables into a single scalar value, it was used only for process-window comparison and not as a complete predictor of melt pool behavior, phase constitution, residual stress, or corrosion performance. The process-window screening was performed to identify a common low-porosity LPBF condition for the subsequent comparative alloy study. The screening

results, including optical defect maps and the V_{ED} –porosity relationship, are presented in Section 4.2. The final common processing condition was selected only after evaluating the measured porosity response.

3.3 Heat-treatment conditions

Post-build heat treatments were applied to the baseline UNS S32750 super duplex stainless steel (SDSS) 2507 alloy, SDSS 2507 +3 wt.% Ni, and SDSS 2507 +6 wt.% Ni. The as-built (AS) condition was retained as the reference state for each alloy system. The experimental design was alloy-specific rather than fully factorial: the baseline alloy was examined after exposure at 300, 400, and 550 °C, the +3 wt.% Ni alloy over the 400–550 °C range, and the +6 wt.% Ni alloy over the 350–525 °C range. These schedules were selected to sample low-, intermediate-, and higher-temperature post-build thermal-response regimes for each composition. Direct comparison is therefore strongest within each alloy system, whereas cross-alloy comparisons are interpreted as pathway-dependent trends rather than direct composition–temperature–time effects at matched thermal schedules.

The low- to intermediate-temperature treatments are referred to as nominal stress-relief treatments, denoted SR. They were used to examine the response of the LPBF-inherited microstructure to controlled thermal exposure, including possible recovery-related changes and modification of the phase constitution. Depending on alloy composition, temperature, and holding time, thermal exposure within this range may also involve ferrite ageing and associated phase-decomposition or precipitation phenomena [3, 46, 47, 95]. Residual-stress relaxation, phase decomposition, and secondary-phase precipitation were therefore not assumed *a priori*; these responses were interpreted only where supported by the corresponding characterization results. Furnace cooling (FC) was applied after the nominal stress-relief treatments. The reported holding time therefore represents the nominal isothermal dwell, while the complete thermal history also includes cooling within the furnace. Consequently, the SR conditions should be interpreted as defined thermal-processing routes rather than as purely isothermal ageing experiments.

The high-temperature solution-annealing treatments, denoted SA, were selected to promote ferrite–austenite phase rebalancing, chemical homogenization, and dissolution of secondary phases where thermodynamically and kinetically possible [1, 2, 31]. The baseline and +3 wt.% Ni alloys were solution annealed at 1100 °C for 15 min, whereas the +6 wt.% Ni alloy was solution annealed at 1150 °C for 10 min. Water quenching (WQ) followed solution annealing to limit precipitation during cooling through intermediate-temperature transformation ranges.

The heat-treatment designations, temperatures, holding times, cooling routes, and treatment classes are summarized in Table 3.3. The condition labels denote applied thermal-processing routes rather than predefined microstructural states. Subsequent interpretations of phase rebalancing, ferrite ageing, secondary-phase formation or dissolution, and local orientation heterogeneity are therefore based on the corresponding microstructural and crystallographic characterization results.

Table 3.3: Alloy-specific post-build heat-treatment matrix for the investigated LPBF SDSS 2507 systems.

Condition	Temperature	Time	Cooling	Treatment class
Baseline SDSS 2507				
AS	–	–	–	As-built reference
SR300	300 °C	5 h	FC	Nominal stress relief
SR400	400 °C	1 h	FC	Nominal stress relief
SR550	550 °C	5 h	FC	Nominal stress relief
SA1100	1100 °C	15 min	WQ	Solution annealing
SDSS 2507 +3 wt.% Ni				
AS	–	–	–	As-built reference
SR400	400 °C	1 h	FC	Nominal stress relief
SR450	450 °C	1 h	FC	Nominal stress relief
SR500	500 °C	1 h	FC	Nominal stress relief
SR550	550 °C	1 h	FC	Nominal stress relief
SA1100	1100 °C	15 min	WQ	Solution annealing
SDSS 2507 +6 wt.% Ni				
AS	–	–	–	As-built reference
SR350	350 °C	2 h	FC	Nominal stress relief
SR425	425 °C	1 h	FC	Nominal stress relief
SR525	525 °C	1 h	FC	Nominal stress relief
SA1150	1150 °C	10 min	WQ	Solution annealing

3.4 Sample preparation

Metallographic specimens were sectioned from the printed coupons along selected build-related planes. Where orientation effects were evaluated, the section plane is reported explicitly as XY or XZ to avoid ambiguity between build-plane and build-direction microstructures. Samples were mounted in epoxy resin and ground using SiC papers from 500 to 2000 grit, followed by diamond polishing from 9 μm to 1 μm and final polishing with 0.04 μm colloidal silica. Electrolytic etching was performed using 10% oxalic acid solution. The solution was prepared by dissolving 1.26 g of oxalic acid dihydrate in distilled water and diluting it to a final volume of 100 mL. Etching conditions were optimized within 3–6 V and 5–60 s, depending on the alloy condition; a representative condition was 6 V for 20 s. The etching response was treated qualitatively because phase contrast can depend on local composition, residual strain, and surface preparation.

3.5 Microstructural and XRD characterization

Microstructural characterization was first performed using light optical microscopy, scanning electron microscopy (SEM), energy-dispersive X-ray spectroscopy (EDS), and electron backscatter diffraction (EBSD). Optical microscopy was conducted using a ZEISS Axio Observer microscope with

differential interference contrast capability. SEM examination was performed using a Zeiss Supra 35 microscope operated at 20 kV. EBSD was performed using a Zeiss SUPRA 25 field-emission scanning electron microscope (FE-SEM) equipped with an EDAX NT EBSD detector at 20 kV and a nominal step size of $0.4 \mu\text{m}$. For each condition, one EBSD map of approximately $500 \times 500 \mu\text{m}^2$ was acquired. EBSD-derived statistics are therefore reported as mapped-area descriptors rather than full build-scale averages. The EBSD data were used to quantify phase fraction, grain morphology, crystallographic texture, phase-boundary character, and local orientation-spread metrics, including kernel average misorientation (KAM). Phase fractions and misorientation values were interpreted with consideration of indexing quality, step size, surface preparation, and the inability of EBSD to resolve nanoscale precipitates or highly deformed substructures below the scan resolution.

EBSD data were processed in MATLAB using MTEX 6.1.0. Ferrite and austenite were indexed as $\alpha\text{-Fe}$ and $\gamma\text{-Fe}$, respectively, using cubic $m\bar{3}m$ symmetry and lattice parameters of $2.870 \text{ \AA}$ and $3.650 \text{ \AA}$. Neighbor-orientation and confidence-index (CI) correlation cleanup were used to reduce isolated indexing noise, and points with $CI < 0.10$ were excluded before quantitative analysis. Grain reconstruction used a minimum misorientation threshold of 1° . Low-angle grain boundaries (LAGBs) were defined as boundaries with disorientation angles of $1^\circ\text{--}15^\circ$, whereas high-angle grain boundaries (HAGBs) were defined as boundaries with disorientation angles $\geq 15^\circ$. HAGB histograms were plotted over $15^\circ\text{--}65^\circ$. Ferrite–austenite interphase boundaries were analyzed separately from same-phase boundaries. KAM, grain reference orientation deviation (GROD), and grain orientation spread (GOS) were used as orientation-spread descriptors, not as direct measurements of total dislocation density or residual stress. Phase-resolved orientation distribution functions (ODFs) were calculated from indexed EBSD points using the MTEX `calcDensity` function with a 10° half-width and 10° angular resolution. ODF sections were plotted at $\varphi_2 = 0^\circ$, 45° , and 65° , with no specimen symmetry imposed. Because the austenite fraction was low in the AS and SR400–SR550 conditions, austenite ODFs for these conditions were interpreted qualitatively.

Same-phase near-coincident site lattice (near-CSL) boundaries were identified for selected Σ values from 3 to 31, namely $\Sigma = 3, 5, 7, 9, 11, 13, 17, 19, 21, 25, 27, 29$, and 31. These boundaries were classified using the Brandon criterion [96, 97]:

$$\Delta\theta_\Sigma = \frac{15^\circ}{\sqrt{\Sigma}}, \quad (3.2)$$

where $\Delta\theta_\Sigma$ is the permitted angular deviation for a near-CSL boundary of order Σ . Ferrite–austenite interphase boundaries were excluded from the near-CSL analysis. The reported near-CSL fractions are therefore misorientation-classified same-phase boundary fractions. They are not complete five-parameter grain-boundary character distributions. Combined same-phase CSL fractions were used only as descriptive boundary-network statistics. Orientation relationships (ORs) between ferrite and austenite were evaluated from interphase-boundary misorientations. The measured boundaries were compared with ideal Kurdjumov–Sachs (KS) and Nishiyama–Wassermann (NW) variants

generated from the α -Fe and γ -Fe crystal symmetries. For each interphase-boundary segment, the minimum angular deviation from the nearest KS or NW variant was calculated. Boundaries within 5° of either ideal relationship were classified as OR-compatible. This classification is tolerance-dependent and was used only as a comparative descriptor.

X-ray diffraction (XRD) was used for phase identification and comparative assessment of diffraction broadening in laser powder bed fusion (LPBF)-fabricated super duplex stainless steel (SDSS) 2507. Measurements were performed using a PANalytical X'Pert Pro diffractometer with Co-K α radiation ($\lambda = 0.1789$ nm), operated at 40 kV and 30 mA. Scans were acquired in Bragg–Brentano geometry over a 2θ range of 30° – 100° , using a step size of 0.05° and a counting time of 100 s per step. Phase identification was conducted using X'Pert HighScore Plus with Joint Committee on Powder Diffraction Standards (JCPDS) and Inorganic Crystal Structure Database (ICSD) reference data. Diffraction-broadening descriptors were evaluated separately for ferrite, denoted as α -Fe, and austenite, denoted as γ -Fe. The α -Fe analysis used the (110), (200), and (211) reflections, whereas the γ -Fe analysis used the (111), (200), and (220) reflections. The overlapping γ -Fe (111)/ α -Fe (110) region was fitted jointly, whereas the remaining reflections were fitted individually using pseudo-Voigt functions on local linear backgrounds. A phase result was retained only when all three target reflections satisfied a minimum signal-to-noise ratio of 5 and a fit-window coefficient of determination, R^2 , of at least 0.80. The AS γ -Fe reflection sets of standard SDSS 2507 and SDSS 2507 +3 wt.% Ni did not satisfy these criteria and were therefore excluded from normalized γ -Fe stress-relief trajectories. Unless explicitly reported as phase-resolved, the stress-relief broadening trajectories presented in this work are based on the α -Fe reflection set.

A classical Williamson–Hall broadening framework was used because it provides a transparent first-order basis for separating size-related and strain-related contributions to XRD peak broadening using the limited set of consistently fitted reflections available across the investigated alloy conditions [98, 99]. The classical form was selected instead of modified Williamson–Hall approaches because anisotropic strain treatment, reflection-dependent elastic or dislocation contrast factors, and reliable instrumental-broadening correction were not available for all alloy systems and conditions. The relation is written as

$$\beta \cos \theta = \frac{K\lambda}{D} + 4\varepsilon \sin \theta, \quad (3.3)$$

where β is the fitted full width at half maximum in radians, θ is the Bragg angle, K is the shape factor, λ is the X-ray wavelength, D is the apparent coherent-domain-size descriptor, and ε is the breadth-derived lattice-microstrain descriptor. In a Williamson–Hall plot, the intercept corresponds to $K\lambda/D$, whereas the slope corresponds to ε . Because several retained reflection sets contained only three usable peaks and showed limited linearity, the classical Williamson–Hall relation was used only as a descriptor-level broadening framework, not as rigorous regression-based quantification of absolute coherent-domain size or lattice strain.

For each retained phase and condition, the median fitted full width at half maximum (FWHM),

β_{med} , and median Bragg angle, θ_{med} , were used to calculate a Scherrer-type apparent crystallite-size descriptor [100, 99]:

$$D = \frac{K\lambda}{\beta_{\text{med}} \cos \theta_{\text{med}}}, \quad (3.4)$$

where D is reported in nm, $K = 0.9$, λ is the Co- $K\alpha$ wavelength, β_{med} is expressed in radians, and θ_{med} is one half of the fitted 2θ peak position. A size-related broadening descriptor was calculated as

$$\rho_D = \frac{1}{D^2}, \quad (3.5)$$

where ρ_D is expressed in m^{-2} after conversion from nm^{-2} . Because ρ_D is mathematically derived from D , it was used only as the size-related counterpart to D , not as an independent measure of microstructural change or absolute dislocation density. A breadth-derived lattice-microstrain descriptor was estimated as [99, 101]

$$\varepsilon = \frac{\beta_{\text{med}}}{4 \tan \theta_{\text{med}}}, \quad (3.6)$$

where ε is dimensionless. Peak-breadth values and Bragg angles were converted from degrees to radians before calculation. Instrumental-broadening correction was not applied because no standard-reference scan was available. Accordingly, D , ρ_D , and ε were used exclusively as comparative diffraction-broadening descriptors, not as direct measurements of residual stress, absolute coherent-domain size, absolute dislocation density, or precipitation state.

The XRD-derived descriptors were compared with scanning electron microscopy (SEM), electron backscatter diffraction (EBSD), transmission electron microscopy (TEM), hardness, and mechanical results. Texture, phase overlap, residual stress, elastic anisotropy, fitted-background selection, and the absence of instrumental-broadening correction can influence fitted peak breadths [99, 101, 102].

3.6 Transmission electron microscopy

Transmission electron microscopy (TEM) and scanning transmission electron microscopy (STEM) were used for nanoscale analysis of precipitates, cellular structures, and dislocation features. Site-specific lamellae were prepared by focused ion beam (FIB) milling and analyzed using a Titan 80–300 FEI scanning/transmission electron microscope (S/TEM) operated at 300 kV. Bright-field TEM, high-resolution transmission electron microscopy (HRTEM), high-angle annular dark-field STEM (HAADF-STEM), selected-area electron diffraction (SAED), and nanoscale energy-dispersive X-ray spectroscopy (EDS) were used where appropriate.

3.7 Mechanical testing

Mechanical testing was conducted to evaluate the macroscopic response of the LPBF-fabricated alloy systems after fabrication and post-build heat treatment. Rockwell A hardness was measured on both surface and cross-sectional regions using a Zwick ZHR 4150TK tester in accordance with ISO 6508. The HRA scale was used with a 60 kgf major load, and 11 measurements were

performed for each condition. Vickers microhardness was measured on polished cross-sections using a Future-Tech FM-ARS tester in accordance with ASTM E384. The measurements were performed using $HV_{0.1}$, corresponding to a 100 gf load, with a dwell time of 15 s. Eleven indents were collected for each condition. The hardness measurements were used to assess changes associated with LPBF processing, Ni modification, and post-build heat treatment.

Tensile testing was performed at room temperature using a Zwick Z100 universal testing machine in accordance with PN-EN ISO 6892-1. Flat sub-sized dog-bone specimens were fabricated in the horizontal orientation and loaded perpendicular to the build direction. The nominal specimen geometry was as follows: total length 89.7 mm, grip-to-grip length 81 mm, gauge length 32 mm, reduced-section width 5.7 mm, grip-section width 8.7 mm, thickness 4.35 mm, and transition radius 25 mm. Four specimens were tested for each condition. The reported tensile parameters include the 0.2% offset yield strength, ultimate tensile strength, and elongation to fracture.

Charpy impact testing was performed using rectangular unnotched specimens with nominal dimensions of 55 mm × 10 mm × 10 mm. The tests were conducted using a pendulum impact tester following PN-EN ISO 148-2 procedures, where applicable to the test geometry, with three specimens tested for each condition. Because the specimens were unnotched, the absorbed energies are interpreted comparatively within the present alloy and heat-treatment matrix and are not reported as standard Charpy V-notch toughness values. Fracture surfaces of tensile and impact specimens from the standard SDSS 2507 alloy were examined by SEM to identify fracture morphology, defect-assisted failure features, and the possible influence of post-build heat treatment on damage tolerance.

3.8 Corrosion testing

Electrochemical testing was conducted in naturally aerated 3.5 wt.% NaCl solution at 25 °C using a conventional three-electrode cell. The specimen cross-section served as the working electrode, Ag/AgCl as the reference electrode, and a platinum wire as the counter electrode. Unless otherwise stated, all potentials are reported versus Ag/AgCl. The exposed area and final surface finish were kept constant for all measurements. The experimental workflow comprised open-circuit potential (OCP) monitoring, cyclic potentiodynamic polarization (CPP), and electrochemical impedance spectroscopy (EIS), which provided the primary electrochemical basis for comparing the as-built and heat-treated alloy conditions. The data-driven workflow in Section 3.8.1 was used only as a supplementary descriptor-level comparison.

Three independently prepared specimens were tested for each heat-treatment condition. The OCP was monitored for 60 min before CPP testing. CPP was performed from $E_{OCP} - 100$ mV at 1 mV s^{-1} until the current density reached 1 mA cm^{-2} , followed by scan reversal. The extracted CPP parameters were E_{corr} , J_{corr} , E_{br} , and E_{rp} . Polarization resistance was estimated from the CPP response using the Stern–Geary relation [103, 104],

$$R_p = \frac{\beta_a \beta_c}{2.3 J_{corr} (\beta_a + \beta_c)}, \quad (3.7)$$

where R_p is the area-normalized polarization resistance ($\text{k}\Omega \cdot \text{cm}^2$), J_{corr} is the corrosion current density ($\mu\text{A cm}^{-2}$), and β_a and β_c are the anodic and cathodic Tafel slopes (mV), respectively. For the passive SDSS 2507 alloy systems studied here, R_p was used as a comparative electrochemical indicator, not as a standalone proof of localized-corrosion resistance.

EIS was conducted at OCP over 10^5 – 10^{-2} Hz using a 10 mV sinusoidal perturbation. The spectra were fitted using equivalent electrical circuits selected to represent the dominant impedance response. The fitted parameters included solution resistance, R_s , charge-transfer resistance, R_{ct} , and constant-phase-element parameters where applicable. Fit quality was recorded using the fitting objective, χ^2 , and the maximum absolute residuals in the real and imaginary impedance components, $\max|\Delta Z'|$ and $\max|\Delta Z''|$, respectively:

$$\Delta Z' = Z'_{\text{exp}} - Z'_{\text{fit}}, \quad \Delta Z'' = Z''_{\text{exp}} - Z''_{\text{fit}}. \quad (3.8)$$

The fitted EIS parameters were interpreted together with the OCP and CPP results rather than as independent corrosion-resistance metrics. Repeat electrochemical measurements were retained as replicate-level records. Where a representative trace was required for comparative plotting or tabulation, the trace with the highest R_p or R_{ct} was treated as the upper-bound response within the tested replicate set. This selection represents a best-response comparison, not a statistical mean response.

3.8.1 Data-driven corrosion screening workflow

The supplementary corrosion-screening workflow used descriptors extracted from cyclic potentiodynamic polarization (CPP) curves, together with composition, processing, heat-treatment, and environment variables, to examine whether descriptor-level trends were consistent with the measured electrochemical response. The CPP-derived descriptors included polarization-based quantities such as corrosion potential, corrosion current density, polarization resistance, breakdown potential, repassivation potential, and repassivation-related potential differences, where applicable. Electrochemical impedance spectroscopy (EIS) parameters were not used as model inputs or targets; they were used only as independent comparators when comparing CPP-based screening trends with measured impedance response. The workflow did not replace direct testing and was not used as an independent validation of localized-corrosion resistance. The workflow used an apparent repassivation-region descriptor, $R_{p,\text{app}}$, defined as

$$R_{p,\text{app}} = \left. \frac{E}{J} \right|_{E=E_{\text{rp}}} = \frac{E_{\text{rp}}}{J_{\text{rp}}}, \quad (3.9)$$

where E_{rp} is the repassivation potential measured versus Ag/AgCl and J_{rp} is the current density at E_{rp} . For E_{rp} in mV and J_{rp} in $\mu\text{A cm}^{-2}$, $R_{p,\text{app}}$ is expressed in $\text{k}\Omega \cdot \text{cm}^2$. Because it is calculated from a selected point on the CPP curve rather than from a local slope, it is treated only as an apparent descriptor, not as a mechanistic polarization resistance.

An additional CPP-derived descriptor was calculated to support the corrosion-screening workflow.

This descriptor was defined from the reverse scan at the repassivation potential, E_{rp} , where the corresponding current-density magnitude is denoted as J_{rp} . Because the CPP current-density coordinate is represented on a logarithmic scale, the plotted axis is expressed as $\log_{10}|J|$. The extraction point and notation used for this descriptor are shown schematically in Figure 3.1. The apparent descriptor was calculated as

$$R_{p,app} = \frac{E}{J} \bigg|_{E=E_{rp}} = \frac{E_{rp}}{J_{rp}}, \quad (3.10)$$

where E_{rp} is the repassivation potential obtained from the reverse CPP scan and J_{rp} is the current-density magnitude at that selected potential. The descriptor has units of $\Omega \cdot \text{cm}^2$ when E_{rp} and J_{rp} are expressed in V and Acm^{-2} , respectively; equivalently, it is expressed in $\text{k}\Omega \cdot \text{cm}^2$ when E_{rp} is given in mV and J_{rp} in μAcm^{-2} . In this work, $R_{p,app}$ is used only as an operational CPP-derived screening descriptor. It is not equivalent to the conventional polarization resistance, R_p , obtained from near-open-circuit polarization behaviour or from the Stern–Geary relation, and it is not interpreted as a direct mechanistic measure of corrosion resistance.

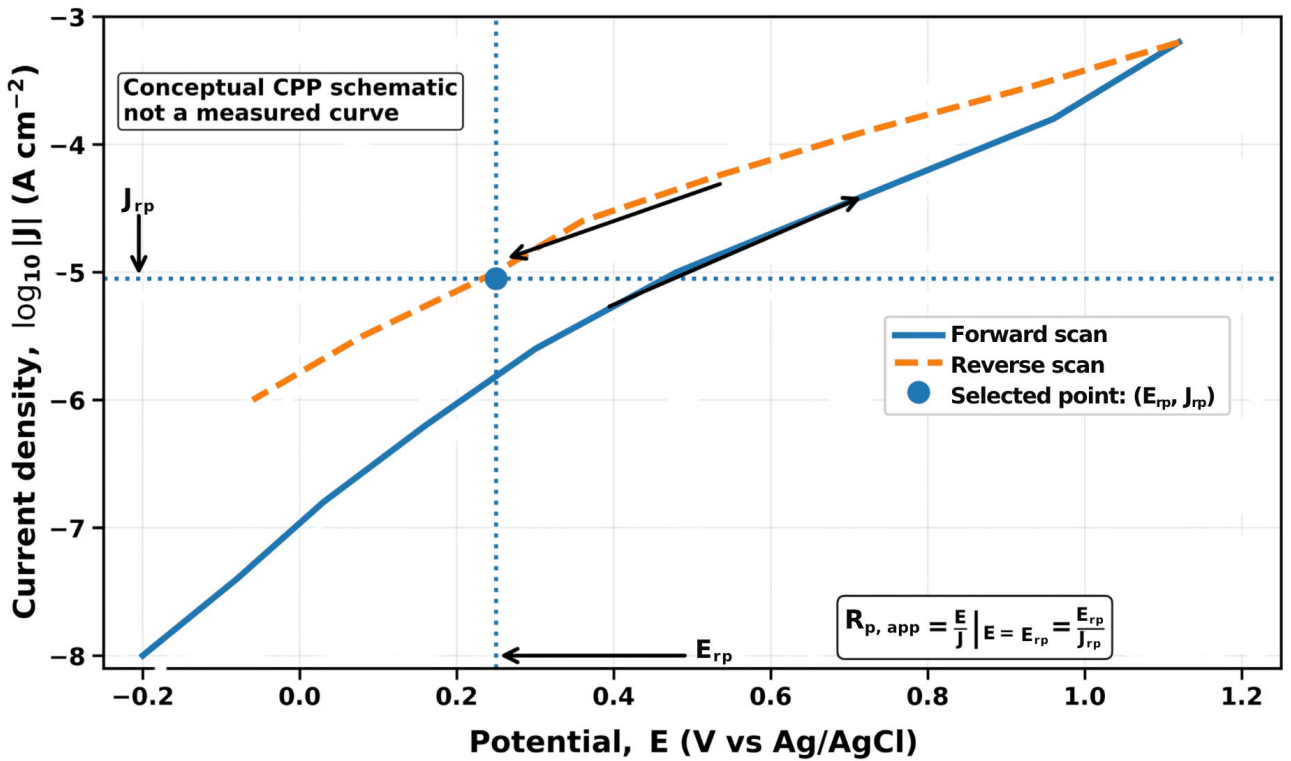


Figure 3.1: Conceptual extraction of the apparent CPP-derived descriptor, $R_{p,app}$, from a cyclic potentiodynamic polarization curve. The selected point is located on the reverse scan at the repassivation potential, E_{rp} , with corresponding current-density magnitude, J_{rp} . The curve is conceptual and is not a measured CPP response. The descriptor is used only as an operational screening descriptor and is not equivalent to conventional polarization resistance.

The model dataset contained 287 corrosion literature-derived potentiodynamic-polarization entries for stainless steels and related corrosion-resistant alloys, together with six SDSS 2507 +3 wt.%

Table 3.4: Descriptor traceability for the CPP-based corrosion-screening workflow.

Descriptor	Workflow status	Source	Interpretation boundary
Model inputs			
Ni, Cr, Mo	Composition inputs	Nominal or reported chemistry	Used as bulk-composition descriptors; local interpretation requires phase-specific and microstructural evidence.
PREN	Engineered composition input	Calculated from nominal chemistry	Used as a ranking descriptor; interpretation requires support from CPP, EIS, and microstructural evidence.
T_{HT}	Heat-treatment input	Heat-treatment matrix	Represents nominal thermal exposure; does not include heating rate, gradients, or local LPBF reheating.
t_{HT}	Heat-treatment input	Heat-treatment matrix	Represents nominal holding time; transformation response also depends on kinetics and prior microstructure.
Cooling route	Categorical processing input	Heat-treatment matrix	Represents cooling class; actual cooling-rate effects are interpreted through experimental evidence.
Electrolyte	Environment input	Experimental protocol and literature records	Used as a test-environment descriptor; literature records may differ in pH, aeration, temperature, chloride activity, and surface finish.
CPP-derived descriptors			
E_{rp}	Used for $R_{p,app}$	CPP reverse scan	Used as a repassivation descriptor within the applied CPP protocol.
$R_{p,app}$	Model target output	E_{rp}/J_{rp}	Apparent point descriptor interpreted within the CPP-screening workflow.
R_p	CPP comparison descriptor	CPP; Stern–Geary estimate	Used for comparative trends; values depend on Tafel-region selection in passive alloys.
E_{br}	CPP comparison descriptor	CPP forward scan	Used as a breakdown descriptor; sensitive to scan rate, surface finish, and pit stabilization.
Independent electrochemical comparator			
R_{ct}	EIS comparator outside the CPP model	EIS equivalent-circuit fitting	Used to compare impedance response with CPP-screening trends; circuit-fit interpretation is model-dependent.

Ni records from the present work. These six records had the same nominal composition but different processing conditions and were included in the training dataset. A tabular neural-network architecture with residual learning and nickel-sensitive weighting was used to estimate $R_{p,app}$ from CPP-relevant composition, processing, heat-treatment, and environment descriptors. Model assessment used cross-validation, with performance recorded using the coefficient of determination, mean absolute error, and root mean square error. Feature sensitivity was evaluated using tree-based feature-importance analysis and permutation scoring. Dataset construction, descriptor definitions, feature engineering, preprocessing, and applicability-domain controls are provided in Appendix A. Because the six SDSS 2507 +3 wt.% Ni records were included during training, their model response was treated as training-domain consistency rather than external validation. The standard SDSS 2507 and SDSS 2507 +6 wt.% Ni datasets were used to assess qualitative cross-alloy consistency outside the closest experimentally represented composition. Model-derived trends were therefore interpreted only as CPP-based descriptor-level screening support and were compared with direct OCP, CPP, and independent EIS evidence.

Chapter 4

Results

This chapter presents the experimental results for three LPBF-fabricated SDSS 2507 alloy systems: standard UNS S32750, +3 wt.% Ni, and +6 wt.% Ni. The results are arranged within a comparative framework to show the effects of alloy chemistry and post-build heat treatment across the alloy series. The chapter first establishes the feedstock condition and the porosity-based LPBF process window selection, and then presents the microstructural, mechanical, and corrosion responses in the as-built and post-processed states.

4.1 Feedstock morphology and particle size

Standard UNS S32750 powder and Ni powder were used as the starting powders for preparing the alloy series. The powder morphology and measured particle-size distributions of the LPBF feedstocks are shown in Figure 4.1. Quantitative particle-size information for the starting powders and LPBF feedstocks is summarized in Table 4.1. Presenting both the morphology and particle-size data documents the feedstock condition prior to LPBF processing and provides the basis for comparing the standard SDSS 2507, +3 wt.% Ni, and +6 wt.% Ni powder batches.

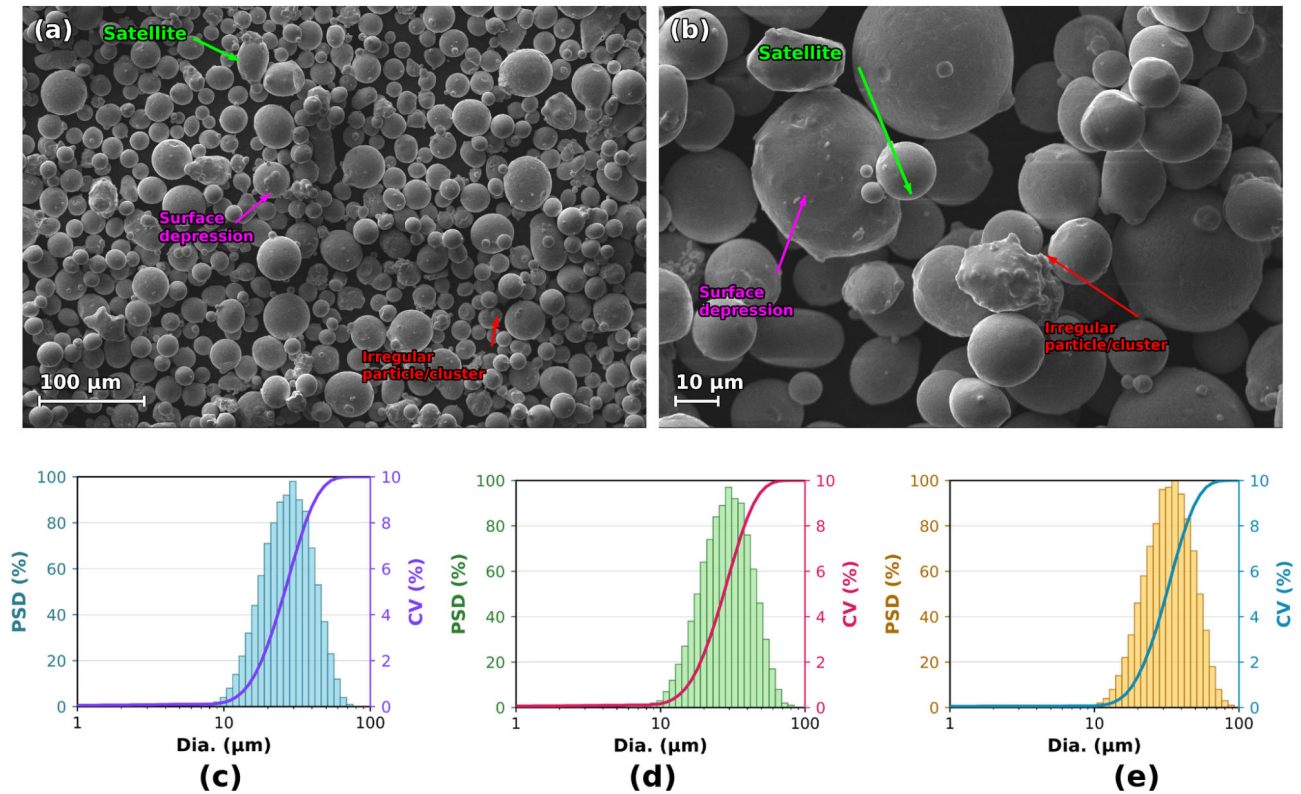


Figure 4.1: Powder morphology and particle-size distribution of the LPBF feedstocks: (a,b) representative SEM micrographs of the gas-atomized SDSS 2507 powder at low and higher magnification, respectively; (c) standard SDSS 2507, (d) SDSS 2507 +3 wt.% Ni, and (e) SDSS 2507 +6 wt.% Ni particle-size distribution histograms with cumulative volume curves.

The low-magnification SEM image in Figure 4.1(a) shows a broad population of predominantly spherical particles. The higher-magnification image in Figure 4.1(b) shows representative local morphological features, including attached satellite particles, surface depressions, and occasional irregular particles or particle clusters. These observations show that the SDSS 2507 powder feedstock is predominantly spherical, with localized deviations from ideal particle morphology.

The measured particle-size distributions of the standard SDSS 2507, SDSS 2507 +3 wt.% Ni, and SDSS 2507 +6 wt.% Ni feedstocks are shown in Figure 4.1(c–e), respectively. The standard SDSS 2507 and +3 wt.% Ni feedstocks exhibit closely comparable distributions, whereas the +6 wt.% Ni feedstock shows a modest shift toward coarser particle sizes. The cumulative volume curves in Figure 4.1(c–e) further show that the standard SDSS 2507 and +3 wt.% Ni feedstocks have nearly overlapping particle-size responses, while the +6 wt.% Ni curve is displaced slightly toward larger diameters. This particle-size difference is noted as a feedstock-level variable to consider when comparing the alloy systems.

For rows where full cumulative particle-size metrics were available, the span was calculated as

$$\text{Span} = \frac{D_{90} - D_{10}}{D_{50}}, \quad (4.1)$$

where D_{10} , D_{50} , and D_{90} represent the particle diameters below which 10%, 50%, and 90% of the measured powder volume occurs, respectively. The quantitative particle-size metrics in Table 4.1 are consistent with the trends shown in Figure 4.1(c–e). Relative to the standard SDSS 2507 feedstock, the +6 wt.% Ni feedstock shows increases of approximately 3.5 μm in D_{10} , 5.5 μm in D_{50} , and 7.8 μm in D_{90} . Despite this shift toward larger particle diameters, the calculated span remains similar across the three LPBF feedstocks, with values of 1.05, 1.05, and 1.00 for the standard SDSS 2507, +3 wt.% Ni, and +6 wt.% Ni feedstocks, respectively.

Table 4.1: Particle-size metrics for starting powders and LPBF feedstocks.

Feedstock / powder	Data source	Size class (μm)	D_{10} (μm)	D_{50} (μm)	D_{90} (μm)	Mean (μm)	Span
Standard SDSS 2507 starting powder	Manufacturer	15–53	19.40	32.40	52.10	NR	1.01
Ni starting powder	Manufacturer	22–45	NR	NR	NR	NC	NC
Standard SDSS 2507	Measured	N/A	14.95	26.07	42.28	28.94	1.05
SDSS 2507 +3 wt.% Ni	Measured	N/A	15.55	27.40	44.39	30.41	1.05
SDSS 2507 +6 wt.% Ni	Measured	N/A	18.48	31.54	50.12	34.76	1.00

Note: NR = not reported; NC = not calculated; N/A = not applicable. Ni powder class with sieve/Microtrac limits rather than full D_{10} , D_{50} , and D_{90} values.

Overall, the combined morphology and particle-size results show that the three LPBF feedstocks have comparable size-distribution breadths and predominantly spherical powder morphology.

The main measured difference among the LPBF feedstocks is the slightly coarser particle-size distribution of the +6 wt.% Ni feedstock.

4.2 LPBF screening and porosity

LPBF process-window screening was performed to identify a common low-porosity as-built processing condition for the standard SDSS 2507, SDSS 2507 +3 wt.% Ni, and SDSS 2507 +6 wt.% Ni alloy systems. The screening matrix is given in Table 3.2. The purpose of this step was not full LPBF process optimization, but selection of a reproducible parameter set with sufficiently low defect content for subsequent microstructural, mechanical, and electrochemical comparisons.

Figure 4.2(a,b) shows the defect fields obtained from the process-screening matrix. Lower- V_{ED} conditions contain a higher density of irregular cavity-like defects and partially fused regions, consistent with lack-of-fusion-type porosity. With increasing V_{ED} , the apparent porosity decreases, the frequency of large irregular defects is reduced, and the remaining defect population becomes less severe in the examined optical fields.

Figure 4.2(c) summarizes the porosity area fraction, ϕ , as a function of volumetric energy density, V_{ED} . The A–D parameter sets include standard-deviation error bars from repeated optical image-analysis fields, whereas the P1–P6 data are shown as a preheated subset. The porosity trend is described empirically by

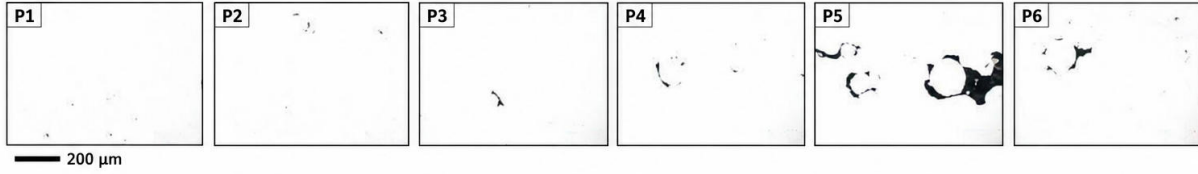
$$\phi = 5.25 \times 10^5 V_{ED}^{-2.61} + 0.22,$$

with $R^2 = 0.974$. This fit is used only as a process-window descriptor for the screened parameter space. Because the P-series and A–D series differ in preheating condition, hatch spacing, and parameter range, the fitted trend is not interpreted as a mechanistic separation of individual LPBF variables.

The horizontal guide lines at 1.0% and 0.5% porosity define practical lower-defect thresholds within the present optical image-analysis dataset. Their intersections with the empirical fit occur near 160 and 188 J mm^{−3}, respectively. These values are used as internal screening references rather than universal process limits, because two-dimensional optical porosity area fraction does not fully capture three-dimensional defect morphology, defect connectivity, or near-surface defect populations.

Based on the combined porosity response and practical process reproducibility, $P = 180$ W, $v = 300$ mm/s, $h = 120$ μm, and $t = 30$ μm were selected, corresponding to $V_{ED} = 166.7$ J mm^{−3}. This condition lies above the empirical 1.0% porosity threshold and was therefore adopted as the common as-built LPBF basis for all three alloy systems. Subsequent alloy comparisons are conditional on this porosity-screened processing basis and do not imply independent optimization of phase balance, residual stress, texture, corrosion response, or mechanical performance.

(a) P-series parameter set



(b) A-D parameter series

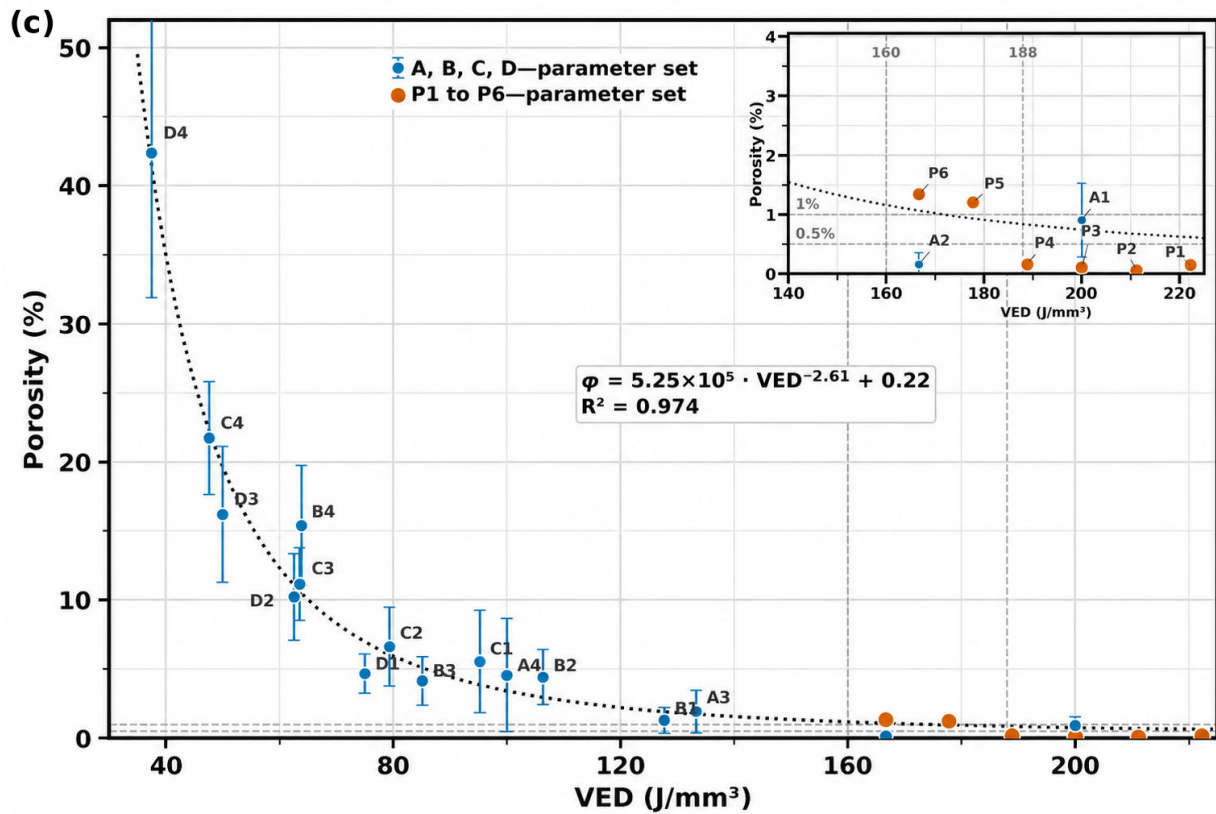
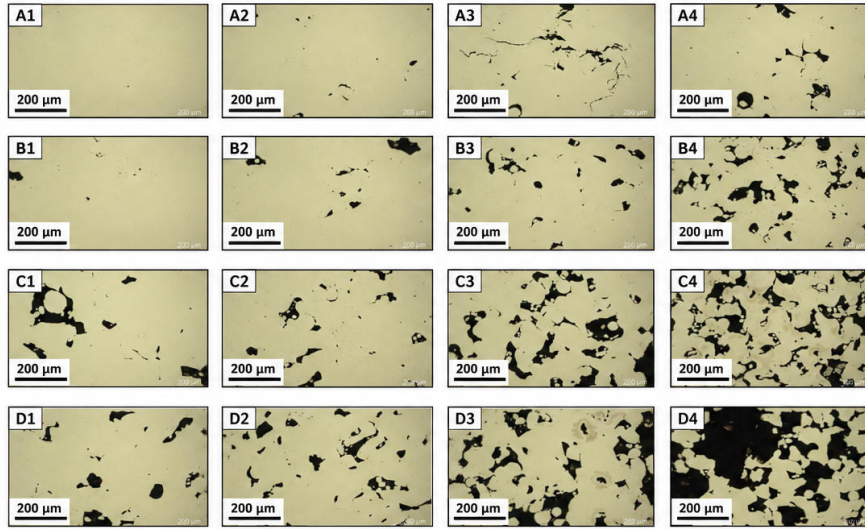


Figure 4.2: Porosity-based LPBF process screening for SDSS 2507. (a) Optical image-analysis fields for the P1–P6 preheated parameter set. (b) Optical image-analysis fields for the A–D parameter series. (c) Porosity area fraction as a function of volumetric energy density, with an empirical inverse-power fit and practical threshold lines.

4.3 Microstructural changes

4.3.1 SEM and XRD results

The scanning electron microscopy (SEM) micrographs in Figures 4.3–4.5 show representative microstructural morphologies of the standard, +3 wt.% Ni, and +6 wt.% Ni laser powder bed fusion (LPBF)-fabricated super duplex stainless steel (SDSS) 2507 alloys after fabrication and post-build heat treatment. The SEM observations are reported as morphology-based results, focusing on the spatial distribution and visible morphology of ferritic and austenitic features in the etched microstructures. Quantitative phase-balance and crystallographic descriptors are presented separately in Section 4.4.

In the standard LPBF-fabricated SDSS 2507 alloy, Figure 4.3 compares the AS, SR550, and SA1100 conditions. The AS microstructure, panels (a,b), contains an α -Fe-rich field with intergranular γ -Fe, triple-junction γ -Fe, grain-boundary austenite (GBA), and subgrain/cell-scale contrast. The visible γ -Fe features are discontinuous and occur primarily along ferritic boundaries and junctions. After SR550, panels (c,d), the ferrite-rich morphology remains evident, with intergranular γ -Fe and residual subgrain/cell traces. In contrast, SA1100, panels (e,f), shows a more developed duplex morphology containing Widmanstätten γ -Fe, interconnected γ -Fe, and retained α -Fe regions.

For the LPBF-fabricated SDSS 2507 +3 wt.% Ni alloy, Figure 4.4 shows the AS, SR400, SR450, SR500, SR550, and SA1100 conditions. Melt-pool banding is visible in AS, panel (a), together with α -Fe and limited SEM-visible γ -Fe features. The stress-relieved states, panels (b–e), retain an α -Fe-rich morphology with boundary-associated and intragranular γ -Fe features. Across SR400–SR550, ferritic packets remain the dominant visible feature, with localized austenitic regions rather than a fully rebalanced duplex morphology. The SR550 condition, panel (e), contains larger visible α -Fe packets. In SA1100, panel (f), boundary-associated and Widmanstätten γ -Fe features are more clearly developed.

For the LPBF-fabricated SDSS 2507 +6 wt.% Ni alloy, Figure 4.5 presents the AS, SR350, SR425, SR525, and SA1150 conditions. The AS microstructure, panel (a), contains triple-junction, discrete, and Widmanstätten γ -Fe features within a mixed SEM-visible α -Fe/ γ -Fe morphology. The stress-relieved states, panels (b–d), retain mixed α -Fe and γ -Fe morphologies, including Widmanstätten, intergranular, discrete, and locally network-like γ -Fe features. Across the non-solution-annealed +6 wt.% Ni conditions, SEM-visible γ -Fe features are distributed more extensively than in the corresponding standard and +3 wt.% Ni condition-specific micrographs; quantitative cross-alloy phase comparison is based on the EBSD results presented in Section 4.4. After SA1150, panel (e), larger α -Fe regions coexist with network-like γ -Fe.

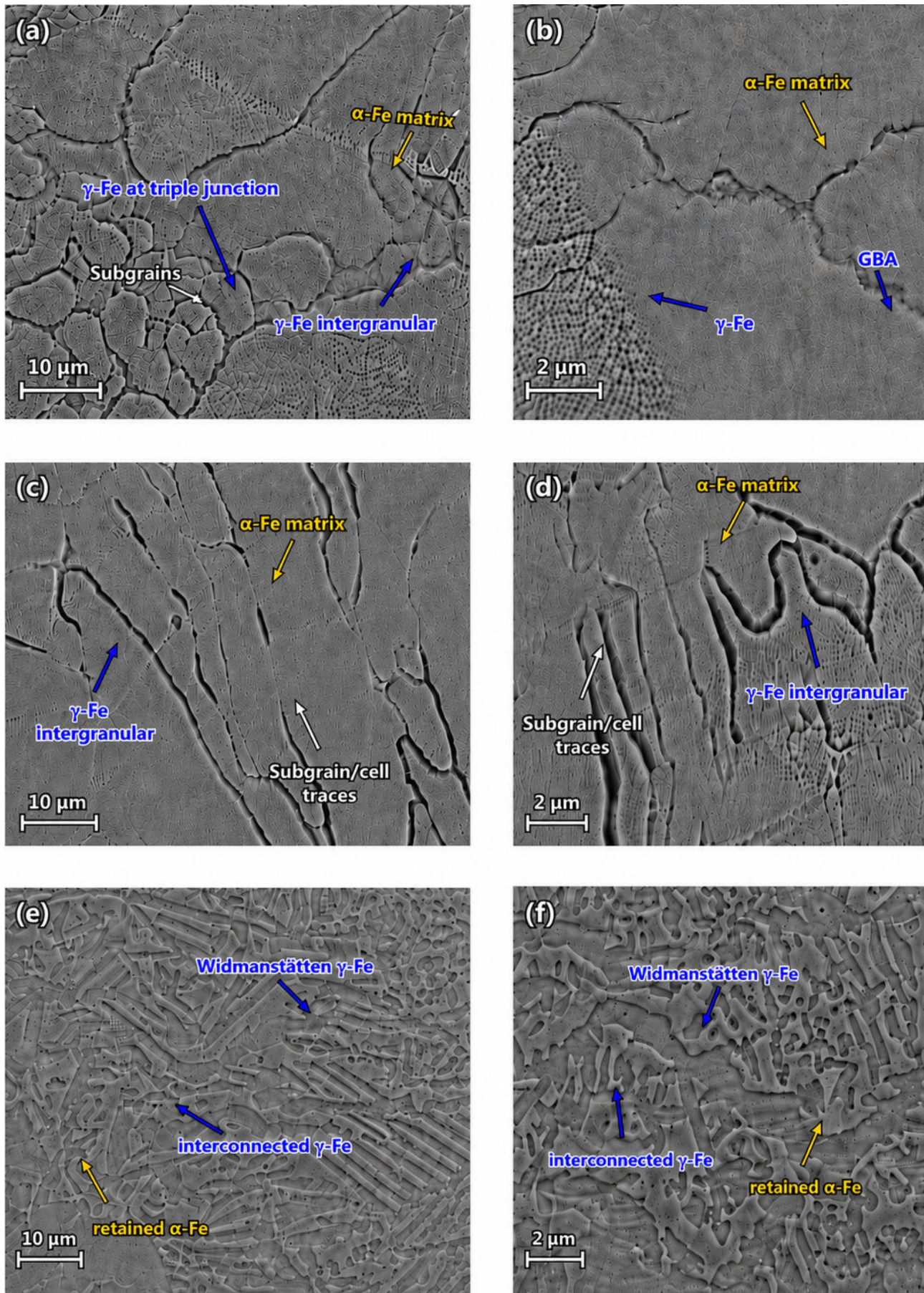


Figure 4.3: Representative SEM micrographs of XY-plane sections of standard LPBF-fabricated SDSS 2507 in the (a,b) AS, (c,d) SR550, and (e,f) SA1100 conditions.

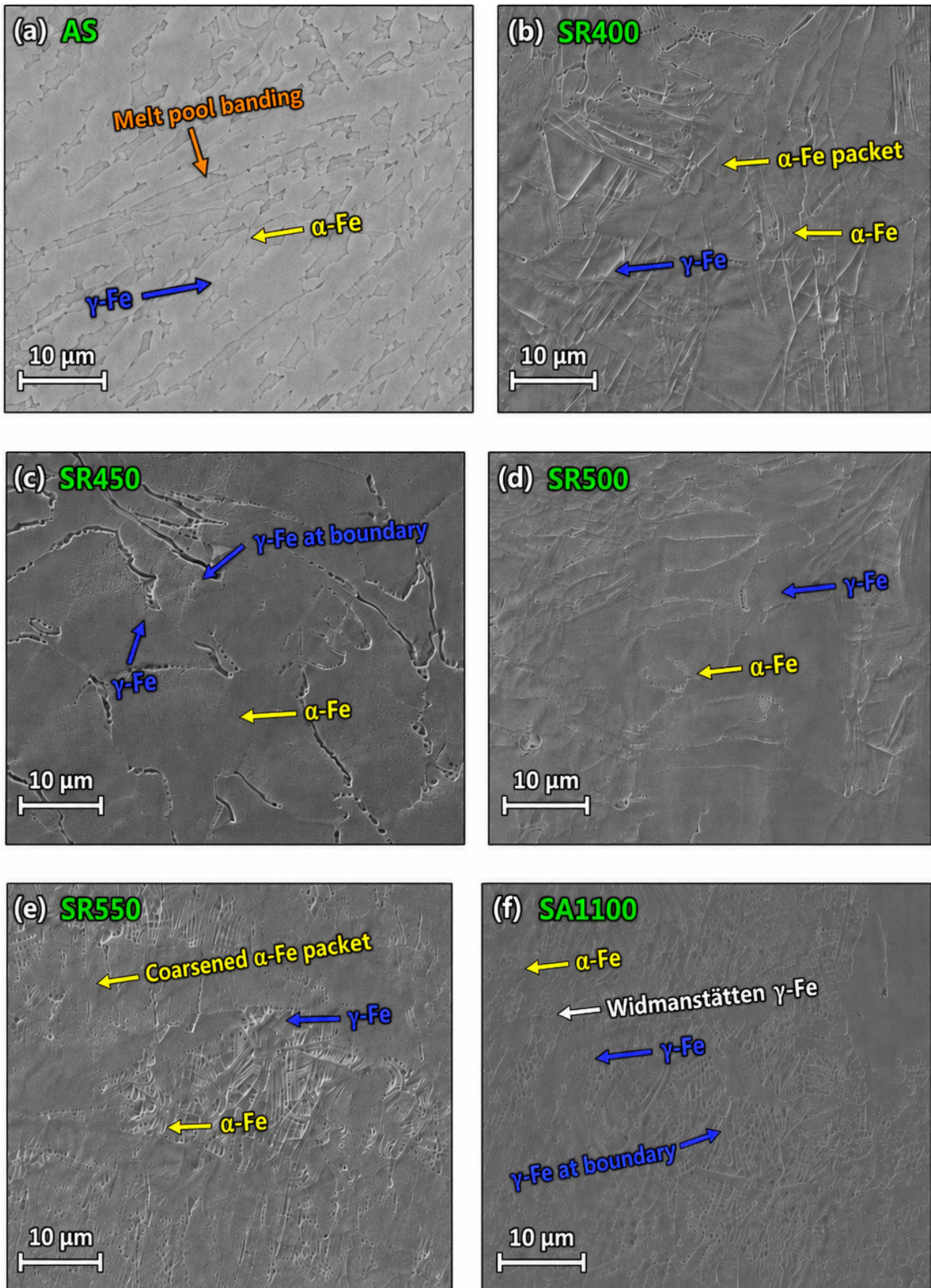


Figure 4.4: Representative SEM micrographs of XZ-plane sections of LPBF-fabricated SDSS 2507 +3 wt.% Ni in the (a) AS, (b) SR400, (c) SR450, (d) SR500, (e) SR550, and (f) SA1100 conditions.

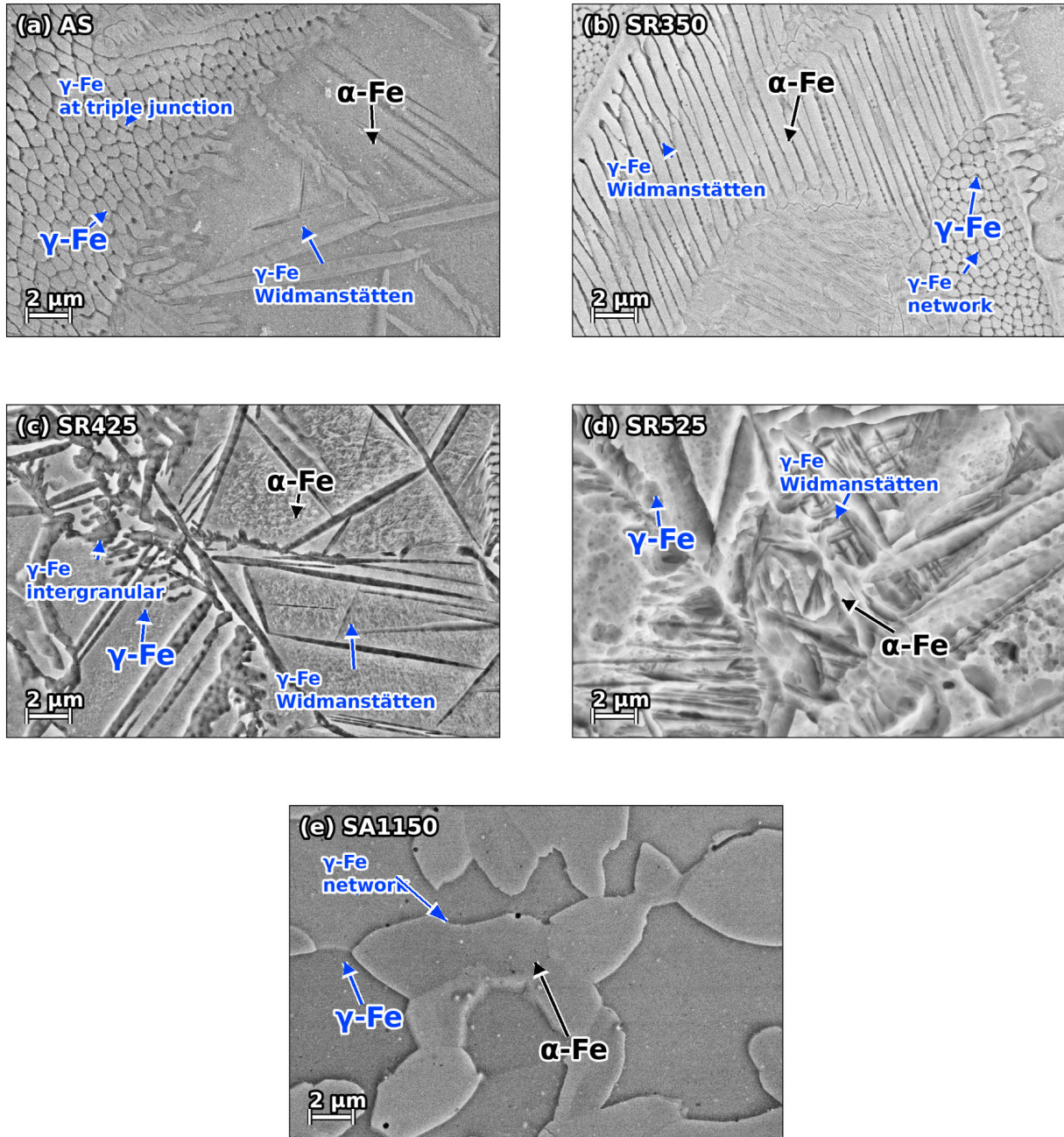


Figure 4.5: Representative SEM micrographs of XZ-plane sections of LPBF-fabricated SDSS 2507 +6 wt.% Ni in the (a) AS, (b) SR350, (c) SR425, (d) SR525, and (e) SA1150 conditions.

Following scanning electron microscopy (SEM), X-ray diffraction (XRD) was used to establish the ferrite–austenite reflection context and to compare diffraction-broadening changes after post-processing. The investigated materials were laser powder bed fusion (LPBF)-fabricated super duplex stainless steel (SDSS) 2507 in its standard composition and with 3 and 6 weight percent (wt.%) nickel (Ni) additions, hereafter referred to as the standard, +3 wt.% Ni, and +6 wt.% Ni alloy systems. The XRD results are used for phase identification and comparative peak-broadening assessment; quantitative phase-constitution comparison is based on the electron backscatter

diffraction (EBSD) results in Section 4.4.

Ferrite (α -Fe) and austenite (γ -Fe) reflections were identified in the investigated alloy systems. For descriptor extraction, the ferritic phase was tracked using the α -Fe {110}, {200}, and {211} reflection families, while the austenitic phase was tracked using the γ -Fe {111}, {200}, {220}, {311}, and {222} reflection families where visible. Crystallographically equivalent cubic indices, such as (110)/(011), (200)/(002), and (211)/(112), were treated as equivalent reflection families. Relative reflection intensities were interpreted qualitatively because they can be affected by texture, phase fraction, absorption, and peak overlap.

Figure 4.6 compares the XRD patterns of the standard and +3 wt.% Ni alloy systems. For the standard alloy, the XRD-investigated conditions were the as-built condition (AS), stress-relieved at 300 °C for 5 h with furnace cooling (SR300), stress-relieved at 400 °C for 1 h with furnace cooling (SR400), stress-relieved at 550 °C for 5 h with furnace cooling (SR550), and solution-annealed at 1100 °C for 15 min followed by water quenching (SA1100). For the +3 wt.% Ni alloy, the XRD-investigated conditions were AS, stress-relieved at 400, 450, 500, and 550 °C for 1 h with furnace cooling (SR400, SR450, SR500, and SR550, respectively), and SA1100.

In both alloy systems, the AS and stress-relieved conditions were dominated by α -Fe reflections, showing that stress relief did not produce a strong XRD-visible austenite response. In contrast, SA1100 showed clearly resolved γ -Fe reflections together with retained α -Fe reflections, consistent with stronger ferrite-to-austenite transformation after solution annealing. The standard and +3 wt.% Ni XRD patterns therefore establish the phase-reflection context for the subsequent α -Fe broadening analysis and for comparison with the EBSD-mapped phase fractions.

For the standard alloy, the AS, SR300, SR400, and SR550 patterns retained a ferrite-dominated diffraction character, with the main α -Fe reflections remaining the most prominent features across the scanned range. The stress-relieved conditions did not show a marked increase in the visible γ -Fe peak response relative to AS. By comparison, SA1100 showed a clear increase in identifiable γ -Fe reflections while retaining α -Fe peaks, confirming that the solution-annealed standard alloy remained duplex rather than fully austenitic.

For the +3 wt.% Ni alloy, the AS and stress-relieved conditions similarly showed dominant α -Fe reflections, indicating that the investigated stress-relief treatments did not generate strong XRD-visible phase rebalancing. The SA1100 condition showed the clearest γ -Fe reflection response within the +3 wt.% Ni series, consistent with the EBSD observation that this condition contained the highest austenite fraction among the +3 wt.% Ni conditions.

Across the standard and +3 wt.% Ni alloy systems, the XRD patterns show that low-temperature stress relief preserved the ferrite-rich diffraction character of the LPBF-inherited state, whereas solution annealing produced the most evident diffraction-level austenite response. Because texture and peak overlap affect relative peak intensities, these observations are treated as qualitative phase-identification results. The following peak-broadening analysis is therefore restricted to comparative changes in the selected α -Fe reflections rather than quantitative phase-fraction determination.

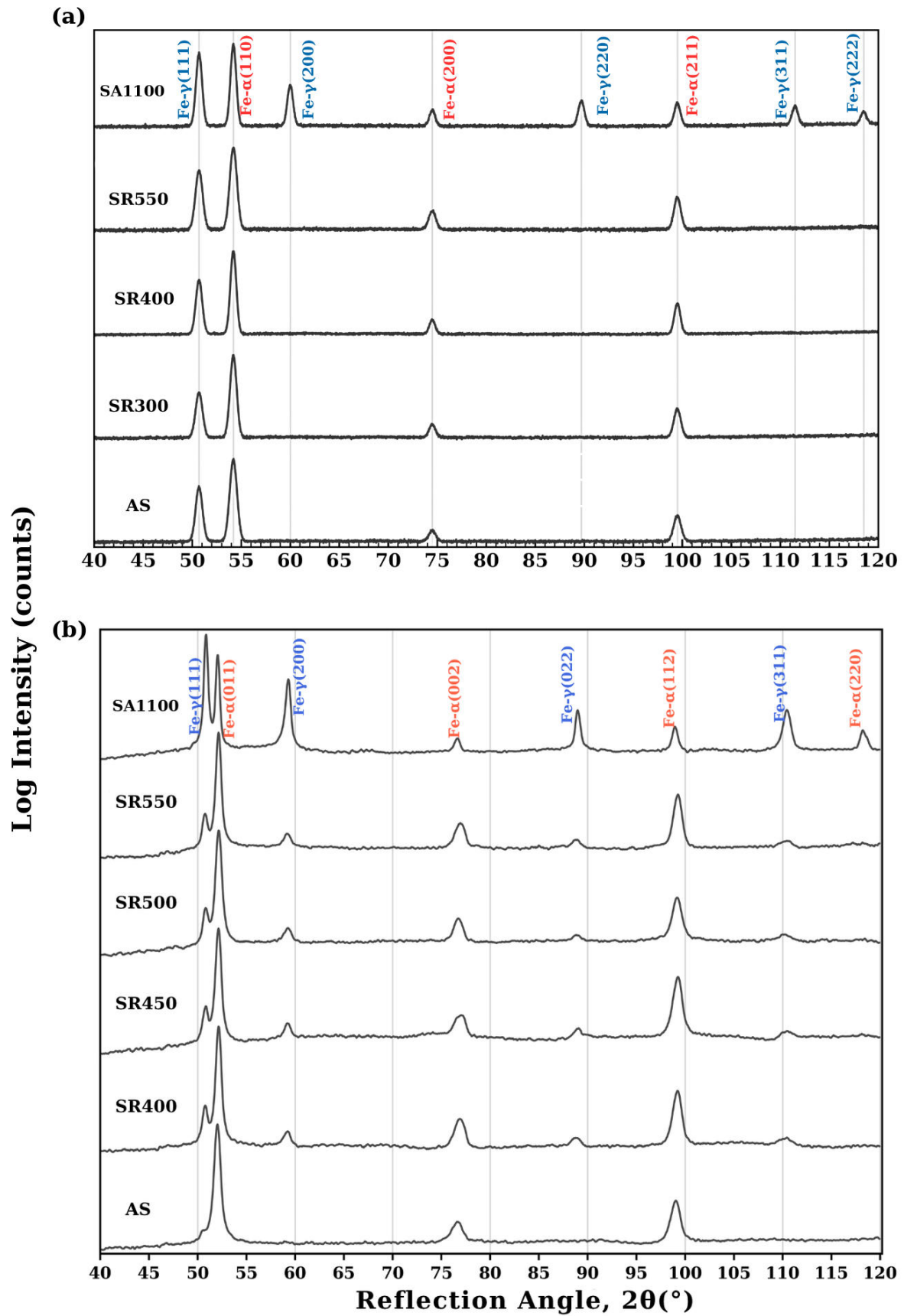


Figure 4.6: XRD patterns of LPBF-fabricated SDSS 2507: **(a)** standard alloy in the AS, SR300, SR400, SR550, and SA1100 conditions, and **(b)** +3 wt.% Ni alloy in the AS, SR400, SR450, SR500, SR550, and SA1100 conditions. Ferritic and austenitic reflections are labelled as α -Fe and γ -Fe, respectively.

For the +6 wt.% Ni alloy system, the corresponding stacked XRD patterns are shown in Figure 4.7. The XRD-investigated conditions were AS, stress-relieved at 350 °C for 2 h with furnace cooling (SR350), stress-relieved at 425 and 525 °C for 1 h with furnace cooling (SR425 and SR525, respectively), and solution-annealed at 1150 °C for 10 min followed by water quenching (SA1150). These patterns provide the ferrite–austenite reflection context for the +6 wt.% Ni heat-treatment sequence before the diffraction-broadening descriptors are compared across the investigated alloy systems.

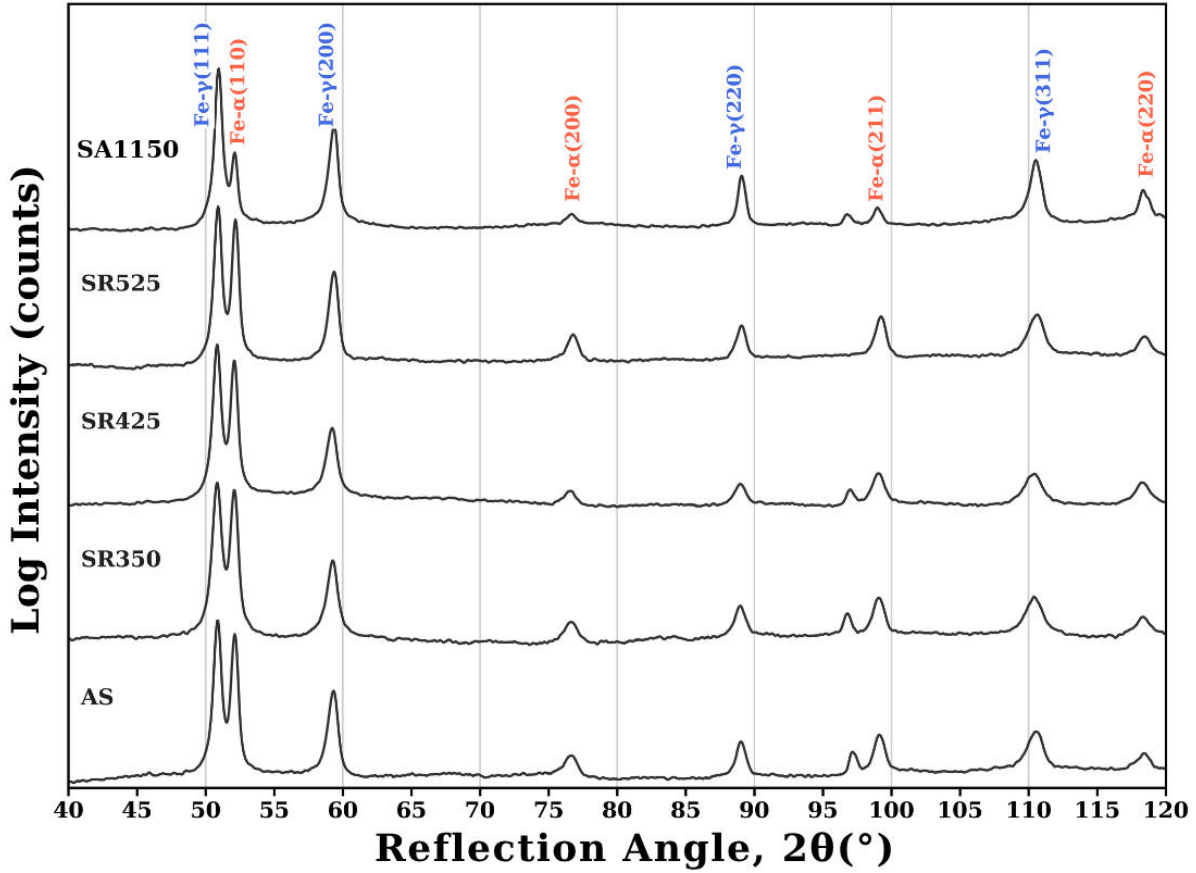


Figure 4.7: XRD patterns of the L-PBF-fabricated SDSS 2507 +6 wt.% Ni alloy in the AS, SR350, SR425, SR525, and SA1150 conditions. Ferritic and austenitic reflections are labelled as α -Fe and γ -Fe, respectively.

The α -Fe-reflection-derived XRD parameters comprise fitted peak positions and breadths, the median peak breadth β_{med} , the apparent coherent-domain size D , the coherent-domain-size-based broadening descriptor ρ_D , and the lattice microstrain descriptor ε for the three alloy systems (Table 4.2). These descriptors are used as comparative XRD-derived quantities for tracking the ferritic diffraction-broadening response after stress relief and solution annealing. Higher β_{med} , ρ_D , and ε values indicate broader ferritic reflections, whereas larger D values indicate narrower ferritic peak breadths within the assumptions of the applied line-broadening treatment. The comparison is therefore made primarily within each alloy system and heat-treatment sequence, rather than as an absolute cross-alloy measurement. The descriptors are not treated as direct measurements of absolute coherent-domain size, dislocation density, residual stress, or precipitation state.

Table 4.2: α -Fe-reflection-derived XRD parameters for the standard, +3 wt.% Ni, and +6 wt.% Ni SDSS 2507 alloy systems. Median aggregation was applied to the fitted α -Fe reflection set.

Alloy system	Condition	Parameter	α -Fe (110)	α -Fe (200)	α -Fe (211)	β_{med} ($^{\circ}$)	D (nm)	ρ_{D} (10^{15} m^{-2})	ε (10^{-3})	
Standard SDSS 2507										
	AS	2θ ($^{\circ}$)	52.10	76.64	99.03	0.955	12.31	6.60	5.27	
		β ($^{\circ}$)	0.51	1.24	0.96					
	SR300	2θ ($^{\circ}$)	52.05	76.51	98.95	0.870	13.50	5.49	4.82	
		β ($^{\circ}$)	0.49	1.20	0.87					
	SR400	2θ ($^{\circ}$)	52.16	76.81	99.11	0.888	13.26	5.69	4.89	
		β ($^{\circ}$)	0.51	1.08	0.89					
	SR550	2θ ($^{\circ}$)	52.17	76.76	99.12	0.823	14.31	4.89	4.53	
		β ($^{\circ}$)	0.45	1.01	0.82					
	SA1100	2θ ($^{\circ}$)	52.05	76.58	98.92	0.754	15.59	4.11	4.17	
		β ($^{\circ}$)	0.38	0.75	0.76					
	SDSS 2507 +3 wt.% Ni									
		AS	2θ ($^{\circ}$)	52.03	76.67	99.03	0.953	12.34	6.57	5.26
			β ($^{\circ}$)	0.48	1.39	0.95				
		SR400	2θ ($^{\circ}$)	52.15	76.94	99.21	0.920	12.81	6.10	5.05
β ($^{\circ}$)			0.44	1.41	0.92					
SR450		2θ ($^{\circ}$)	52.14	77.11	99.25	0.945	12.49	6.41	5.17	
		β ($^{\circ}$)	0.47	1.75	0.94					
SR500		2θ ($^{\circ}$)	52.16	76.76	99.18	0.986	11.93	7.03	5.43	
		β ($^{\circ}$)	0.49	0.99	0.99					
SR550		2θ ($^{\circ}$)	52.15	76.99	99.25	0.874	13.49	5.50	4.80	
		β ($^{\circ}$)	0.40	1.35	0.87					
SA1100		2θ ($^{\circ}$)	52.06	76.63	98.97	0.612	19.20	2.71	3.38	
		β ($^{\circ}$)	0.31	0.62	0.61					
SDSS 2507 +6 wt.% Ni										
		AS	2θ ($^{\circ}$)	52.12	76.66	99.13	0.817	14.39	4.83	4.51
	β ($^{\circ}$)		0.44	0.95	0.82					
	SR350	2θ ($^{\circ}$)	52.09	76.64	99.08	0.865	13.60	5.41	4.77	
		β ($^{\circ}$)	0.48	1.07	0.86					
	SR425	2θ ($^{\circ}$)	52.09	76.58	99.05	0.913	12.88	6.03	5.04	
		β ($^{\circ}$)	0.42	0.91	0.92					
	SR525	2θ ($^{\circ}$)	52.18	76.79	99.23	0.841	14.00	5.10	4.63	
		β ($^{\circ}$)	0.41	0.85	0.84					
	SA1150	2θ ($^{\circ}$)	52.12	76.67	98.99	0.669	17.59	3.23	3.69	
		β ($^{\circ}$)	0.38	0.79	0.67					

Figure 4.8 presents composition-specific normalized changes in the α -Fe-derived XRD descriptors β_{med} , D , and ε relative to the corresponding AS condition. The comparison is made within each alloy system because the AS baselines and stress-relief schedules differ among the standard, +3 wt.% Ni, and +6 wt.% Ni alloys. The descriptor ρ_D is retained in Table 4.2 and is not plotted separately because $\rho_D = 1/D^2$. The standard alloy showed lower β_{med} and ε , together with higher D , after stress relief. The +3 wt.% Ni alloy showed a non-monotonic condition-dependent response, with the highest stress-relief β_{med} , ρ_D , and ε in SR500 and the largest stress-relief D in SR550. The +6 wt.% Ni alloy showed its largest broadening-related change at SR425, where β_{med} and ε reached their maximum normalized increases and D reached its largest normalized decrease. This α -Fe-only comparison summarizes the ferritic broadening response during stress relief, while the subsequent phase-resolved analysis assesses whether the austenitic reflections followed the same post-processing trend.

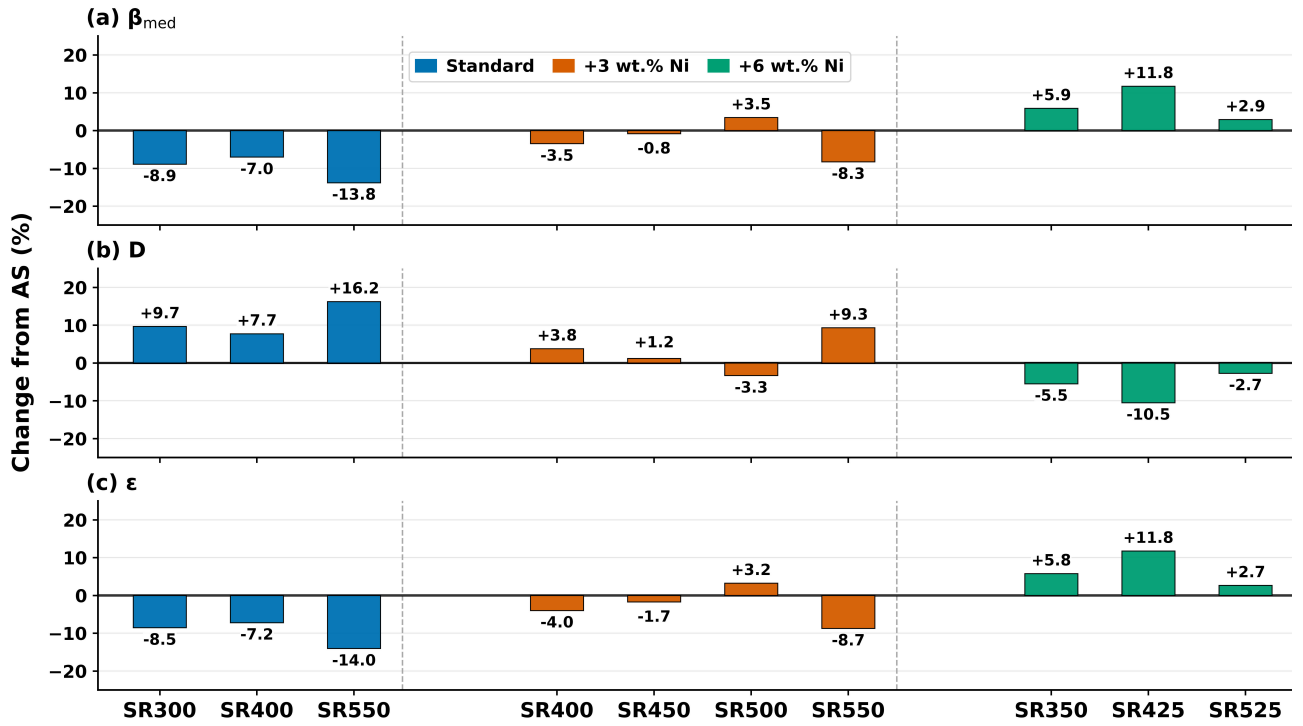


Figure 4.8: Composition-specific normalized changes in α -Fe-reflection-derived XRD descriptors after stress relief in LPBF-fabricated SDSS 2507: (a) β_{med} , (b) D , and (c) ε . Values are expressed as percentage changes relative to the corresponding AS condition within each alloy system. Bars compare separate heat-treatment conditions; vertical separators distinguish alloy compositions, and no connection is made between different stress-relief states.

Phase-resolved XRD descriptors were calculated separately from the α -Fe and γ -Fe reflection sets to compare the ferritic and austenitic diffraction-broadening response (Table 4.3). For both phases, β_{med} , D , ρ_D , and ε were evaluated using the same fitting procedure. The descriptors D , ρ_D , and ε are comparative quantities derived from fitted peak breadths and are not interpreted as direct measurements of absolute coherent-domain size, dislocation density, residual stress, or precipitation state. Because $\rho_D = 1/D^2$, it is reported as the size-related counterpart to D ,

rather than as independent evidence of broadening change. The phase-resolved sequence was most complete for the +6 wt.% Ni alloy because both α -Fe and γ -Fe reflections were sufficiently resolved across the investigated conditions. For the standard and +3 wt.% Ni alloys, phase-resolved comparison was retained for the solution-annealed states, where the austenitic reflections were sufficiently defined for fitting.

Figure 4.9 shows the normalized phase-specific XRD descriptor changes for the +6 wt.% Ni alloy relative to the corresponding phase-specific AS values. From AS to SR425, both phases showed increased broadening-related descriptors and decreased apparent D . For α -Fe, β_{med} increased from 0.817° in AS to 0.865° in SR350 and 0.913° in SR425, while D decreased from 14.39 to 13.60 and then to 12.88 nm. The corresponding ρ_D increased from 4.83×10^{15} to $6.03 \times 10^{15} \text{ m}^{-2}$, and ε increased from 4.51×10^{-3} to 5.04×10^{-3} . For γ -Fe, β_{med} increased from 0.698° in AS to 0.730° in SR350 and 0.740° in SR425, while D decreased from 15.21 to 14.55 and then to 14.34 nm. The corresponding ρ_D increased from 4.32×10^{15} to $4.86 \times 10^{15} \text{ m}^{-2}$, and ε increased from 5.35×10^{-3} to 5.68×10^{-3} .

SR525 shifted both phases toward lower broadening than SR425. In α -Fe, β_{med} , ρ_D , and ε decreased to 0.841° , $5.10 \times 10^{15} \text{ m}^{-2}$, and 4.63×10^{-3} , respectively, while D increased to 14.00 nm. In γ -Fe, the corresponding values decreased to 0.658° , $3.84 \times 10^{15} \text{ m}^{-2}$, and 5.04×10^{-3} , while D increased to 16.14 nm. SA1150 showed the lowest β_{med} , lowest ρ_D , lowest ε , and largest D in both phases within the +6 wt.% Ni sequence. Across the complete sequence, the γ -Fe reflections retained lower β_{med} , lower ρ_D , and larger apparent D than the corresponding α -Fe reflections, whereas the γ -Fe microstrain descriptor remained slightly higher than the α -Fe value. Normalized γ -Fe stress-relief changes are not reported for the standard and +3 wt.% Ni alloys because their AS γ -Fe reflection sets did not meet the same fitting-quality criteria.

Table 4.3: Phase-resolved XRD-derived broadening descriptors for the complete SDSS 2507 +6 wt.% Ni heat-treatment sequence and for the solution-annealed conditions across the investigated alloy systems. The SA1150 row in the +6 wt.% Ni sequence also represents the +6 wt.% Ni solution-annealed condition.

Alloy system	Condition	α -Fe				γ -Fe			
		β_{med}	D	ρ_{D}	ε	β_{med}	D	ρ_{D}	ε
		($^{\circ}$)	(nm)	(10^{15} m^{-2})	(10^{-3})	($^{\circ}$)	(nm)	(10^{15} m^{-2})	(10^{-3})
Complete SDSS 2507 +6 wt.% Ni heat-treatment sequence									
SDSS 2507 +6 wt.% Ni	AS	0.817	14.39	4.83	4.51	0.698	15.21	4.32	5.35
	SR350	0.865	13.60	5.41	4.77	0.730	14.55	4.73	5.60
	SR425	0.913	12.88	6.03	5.04	0.740	14.34	4.86	5.68
	SR525	0.841	14.00	5.10	4.63	0.658	16.14	3.84	5.04
	SA1150	0.669	17.59	3.23	3.69	0.552	19.24	2.70	4.23
Solution-annealed comparison across alloy systems									
Standard SDSS 2507	SA1100	0.754	15.59	4.11	4.17	0.566	18.73	2.85	4.35
SDSS 2507 +3 wt.% Ni	SA1100	0.612	19.20	2.71	3.38	0.446	23.82	1.76	3.42

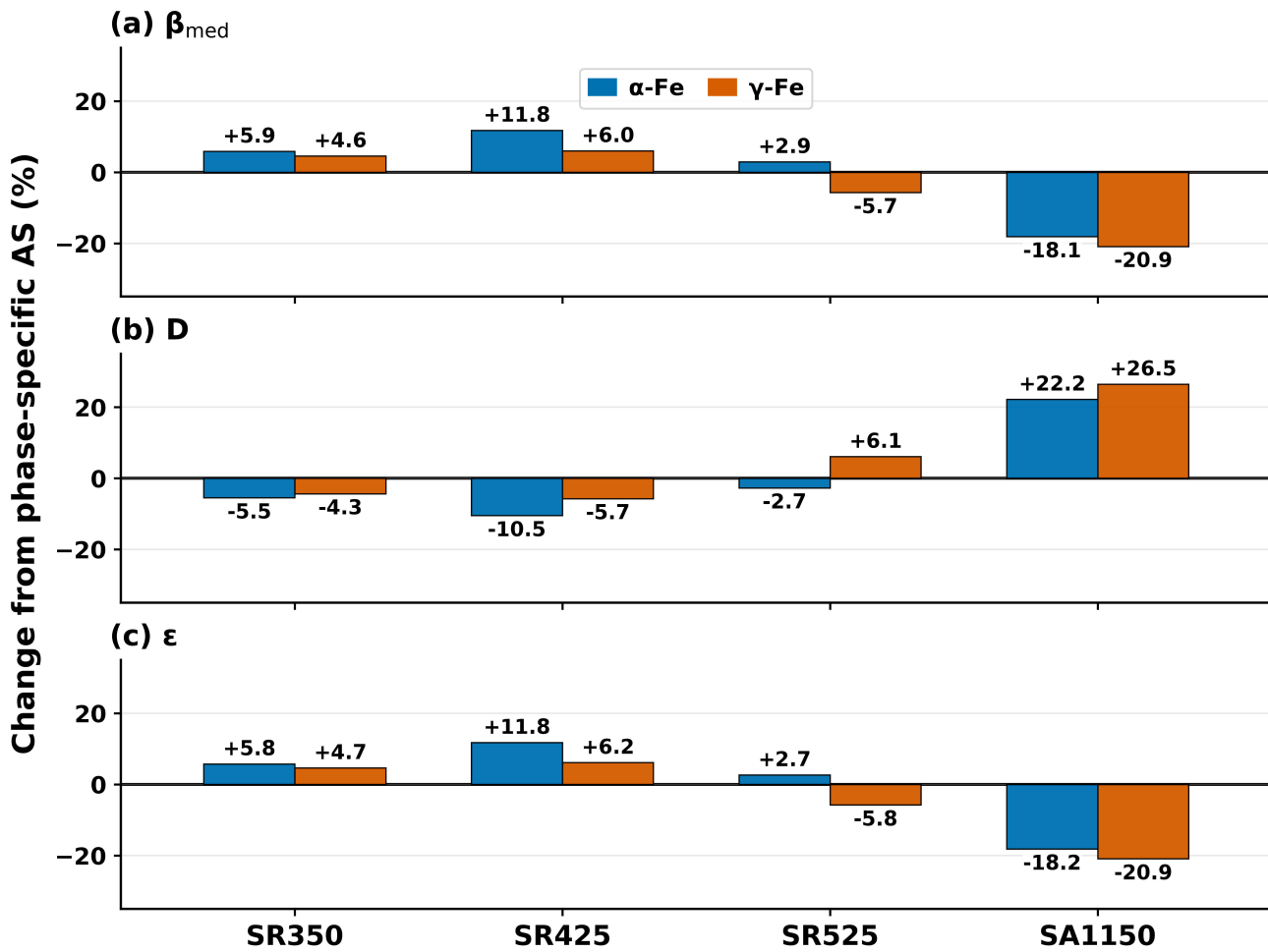


Figure 4.9: Phase-specific normalized XRD-derived descriptor changes for the LPBF-fabricated SDSS 2507 +6 wt.% Ni alloy: (a) β_{med} , (b) D , and (c) ϵ . Values are expressed as percentage changes relative to the corresponding phase-specific AS value. Bars compare separate heat-treatment conditions and do not imply continuous evolution between states.

The XRD-derived descriptors and EBSD kernel average misorientation (KAM) results showed related but scale-dependent changes after heat treatment. The XRD results captured phase-specific diffraction-broadening changes averaged over the irradiated volume, whereas KAM maps showed local orientation-gradient changes within the EBSD measurement field. In the stress-relieved and solution-annealed conditions, reductions in XRD peak broadening generally corresponded to lower local orientation heterogeneity, while increased broadening was associated with higher KAM-based orientation gradients.

These results are treated as comparative descriptors of the heat-treatment response, not as direct measurements of residual stress, absolute dislocation density, or a specific precipitation state. XRD therefore supports the phase-specific broadening assessment, while KAM supports the local misorientation assessment. Their relationship with phase constitution, hardness, tensile strength, ductility, impact response, and laboratory electrochemical behaviour is considered in the integrated discussion in Chapter 5.

4.4 Crystallographic changes

4.4.1 Phase distribution and grain morphology

Figure 4.10 shows the electron backscatter diffraction (EBSD) phase and inverse pole figure (IPF) maps of the standard laser powder bed fusion (LPBF)-fabricated super duplex stainless steel (SDSS) 2507 alloy in the as-built (AS), stress-relieved at 300 °C (SR300), stress-relieved at 550 °C (SR550), and solution-annealed at 1100 °C (SA1100) conditions. The AS map contains 96.4% α -Fe and 3.6% γ -Fe, with a mapped-area grain-size descriptor of 3.05 μm . SR300 and SR550 remain strongly α -Fe dominated, with 99.5 and 99.4% α -Fe, respectively. The corresponding γ -Fe fractions are 0.5 and 0.6%, and the mapped-area grain-size descriptors are 8.99 and 0.31 μm , respectively. In SA1100, the phase distribution is near-balanced, with 50.9% α -Fe and 49.1% γ -Fe, and the mapped-area grain-size descriptor is 3.51 μm . The IPF maps show elongated ferritic grains in AS, SR300, and SR550, while SA1100 shows a mixed α/γ morphology.

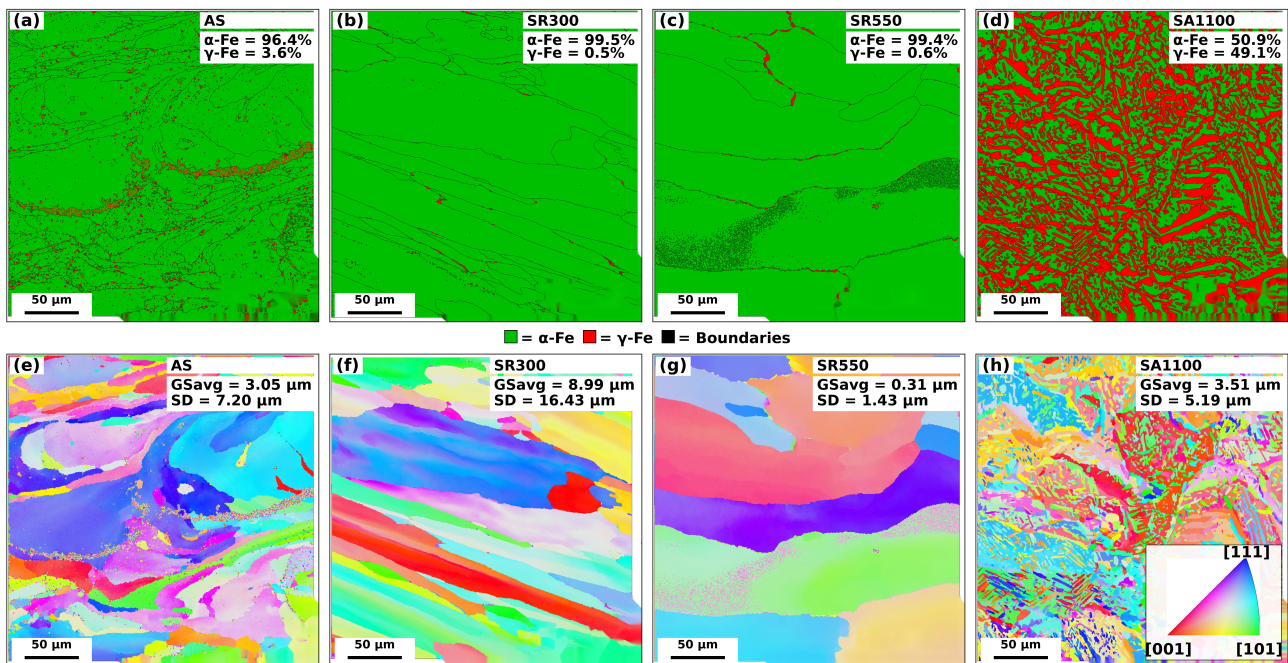


Figure 4.10: EBSD phase maps and corresponding IPF maps of standard LPBF-fabricated SDSS 2507 in the AS, SR300, SR550, and SA1100 conditions. Panels (a–d) show phase maps, and panels (e–h) show the corresponding IPF maps with mapped-area grain-size descriptors.

For the +3 wt.% Ni alloy, the electron backscatter diffraction (EBSD) phase and inverse pole figure (IPF) maps shown in Figure 4.11 include the as-built (AS), stress-relieved at 400 °C for 1 h (SR400), stress-relieved at 450 °C for 1 h (SR450), stress-relieved at 500 °C for 1 h (SR500), stress-relieved at 550 °C for 1 h (SR550), and solution-annealed at 1100 °C for 15 min (SA1100) conditions. The AS map contains 94.7% α -Fe and 5.3% γ -Fe, with a mapped-area grain-size descriptor of 7.74 μm . The stress-relieved maps contain 94.4–95.0% α -Fe and 5.0–5.6% γ -Fe. Their mapped-area grain-size descriptors are 5.49, 7.49, 5.61, and 6.50 μm for SR400, SR450, SR500, and SR550, respectively. The SA1100 map contains 30.6% α -Fe and 69.4% γ -Fe, with the

smallest mapped-area grain-size descriptor of 3.64 μm . The IPF maps show elongated and locally banded ferritic grain morphologies in AS and the stress-relieved conditions, while SA1100 shows a finer mixed α/γ morphology.

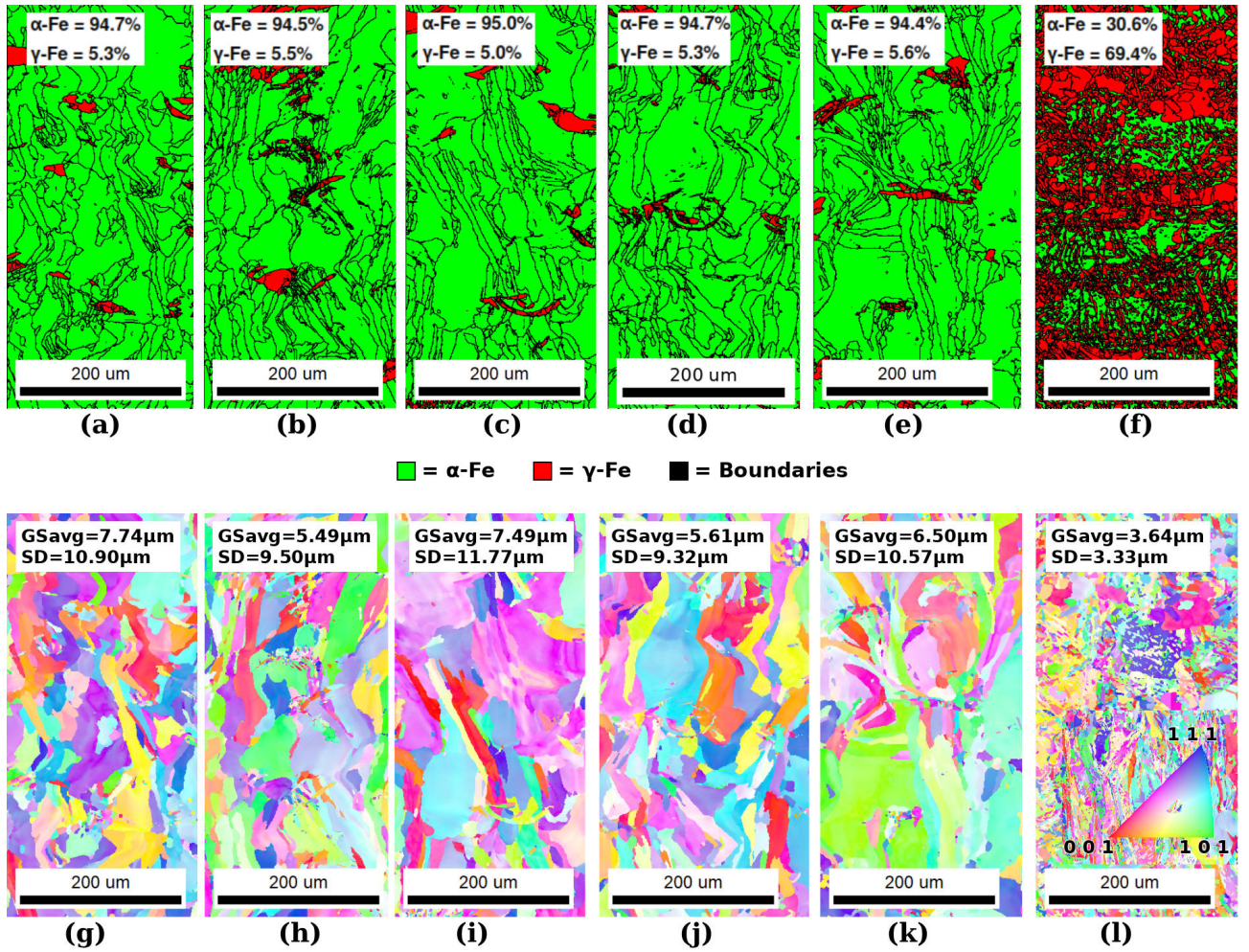


Figure 4.11: EBSD phase maps and corresponding IPF maps of LPBF-fabricated SDSS 2507 +3 wt.% Ni in the AS, SR400, SR450, SR500, SR550, and SA1100 conditions. Panels (a–f) show phase maps, and panels (g–l) show the corresponding IPF maps with mapped-area grain-size descriptors.

For the SDSS 2507 +6 wt.% Ni alloy, the electron backscatter diffraction (EBSD) phase maps in Figure 4.12(a–e) and the corresponding inverse pole figure (IPF) maps in Figure 4.12(f–j) include the as-built (AS), stress-relieved at 350 $^{\circ}\text{C}$ for 2 h (SR350), stress-relieved at 425 $^{\circ}\text{C}$ for 1 h (SR425), stress-relieved at 525 $^{\circ}\text{C}$ for 1 h (SR525), and solution-annealed at 1150 $^{\circ}\text{C}$ for 10 min (SA1150) conditions. The AS map contains 47.9% α -Fe and 52.1% γ -Fe, with a mapped-area grain-size descriptor of 3.59 μm . The SR350 and SR425 maps both contain 45.8% α -Fe and 54.2% γ -Fe, with mapped-area grain-size descriptors of 3.78 and 4.08 μm , respectively. The SR525 map contains 39.2% α -Fe and 60.8% γ -Fe, with the smallest mapped-area grain-size descriptor of 3.18 μm . The SA1150 map contains 10.9% α -Fe and 89.1% γ -Fe, with a mapped-area grain-size descriptor of 4.01 μm . The IPF maps show a fine mixed α/γ morphology across all conditions, with SA1150 showing the highest mapped γ -Fe fraction.

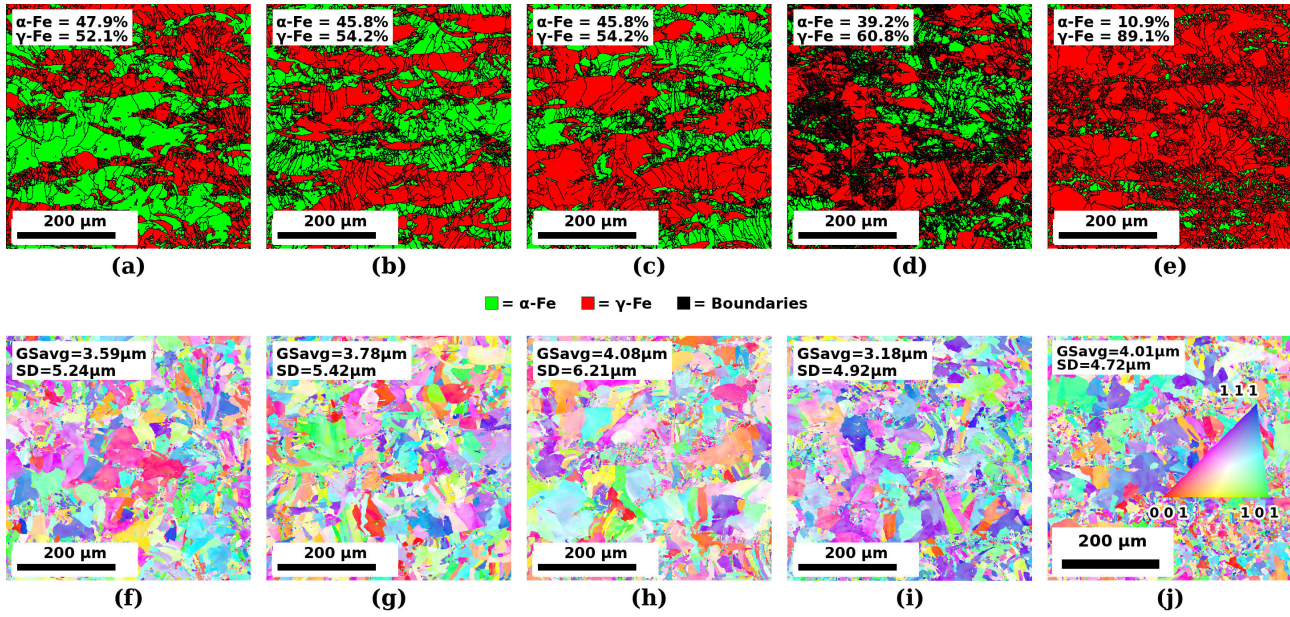


Figure 4.12: EBSD phase maps and corresponding IPF maps of LPBF-fabricated SDSS 2507 +6 wt.% Ni in the AS, SR350, SR425, SR525, and SA1150 conditions. Panels (a–e) show phase maps with α -Fe, γ -Fe, and boundaries; panels (f–j) show the corresponding IPF maps with mapped-area grain-size descriptors. The IPF colour key is shown in panel (j).

4.4.2 Boundary network and near-CSL boundaries

For the standard alloy, the boundary character maps, boundary fraction summary, and misorientation angle distributions in Figure 4.13 cover the as-built (AS), stress-relieved at 300 °C (SR300), stress-relieved at 550 °C (SR550), and solution-annealed at 1100 °C (SA1100) conditions. Boundaries are classified as low-angle grain boundaries (LAGBs), same-phase high-angle grain boundaries (HAGBs), and α/γ interphase boundaries. Near coincident site lattice (near-CSL) boundaries are reported separately. The AS condition contains a mixed boundary population, with contributions from LAGBs, HAGBs, and interphase boundaries.

SR300 and SR550 are dominated by same-phase ferritic boundary networks. Compared with SR300, SR550 shows a larger same-phase HAGB fraction and a stronger high-angle boundary contribution. Interphase boundary contributions remain limited in both stress-relieved states, consistent with the ferrite-rich phase maps in Figure 4.10. In contrast, SA1100 shows the largest α/γ interphase boundary fraction in the standard alloy and accompanies the near-balanced phase distribution. The misorientation-angle distribution for SA1100 also differs from the AS and stress-relieved states, with a larger contribution from interphase and same-phase high-angle boundaries. Overall, Figure 4.13 shows that the standard alloy changes from a ferrite-dominated boundary network in AS, SR300, and SR550 to one with a larger interphase boundary contribution after SA1100. The near-CSL boundary results are reported as boundary population descriptors rather than as evidence for a boundary engineering mechanism.

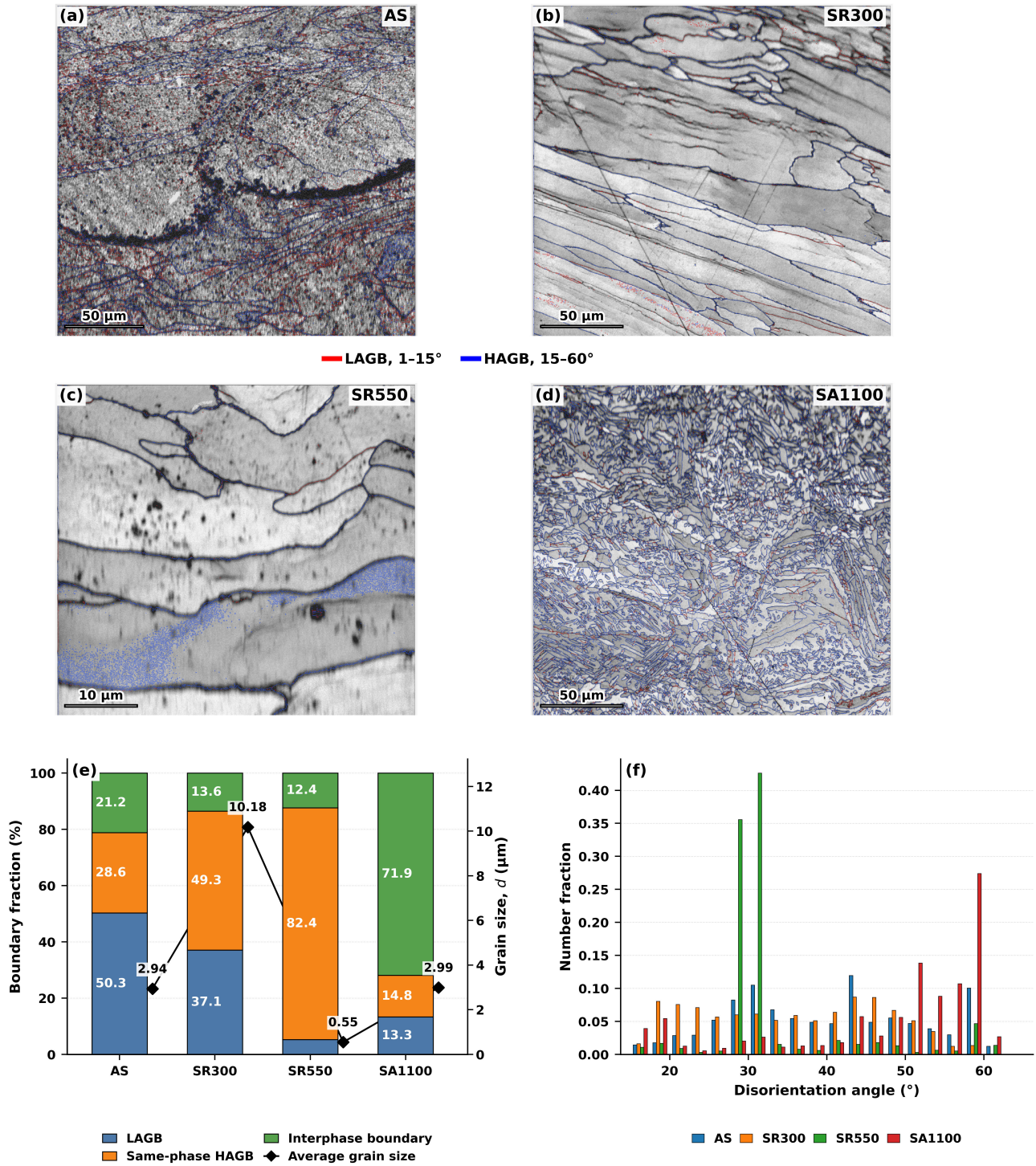


Figure 4.13: Boundary-character maps and boundary-fraction analysis of standard LPBF-fabricated SDSS 2507 in the AS, SR300, SR550, and SA1100 conditions. Panels (a–d) show LAGB/HAGB maps, panel (e) summarizes LAGB, same-phase HAGB, interphase-boundary fractions, and mapped-area grain-size descriptors, and panel (f) shows the corresponding misorientation-angle distributions.

The +3 wt.% Ni alloy boundary-network results in Figure 4.14 include the as-built (AS), stress-relieved at 400 °C for 1 h (SR400), stress-relieved at 450 °C for 1 h (SR450), stress-relieved at 500 °C for 1 h (SR500), stress-relieved at 550 °C for 1 h (SR550), and solution-annealed at 1100 °C for 15 min (SA1100) conditions. HAGB fractions are higher than LAGB fractions in all conditions.

The AS condition has a HAGB fraction of 78.8%. The HAGB fractions for SR400, SR450, SR500, SR550, and SA1100 range from 76.0 to 84.3%. The mapped-area grain-size descriptors are the same as those reported in Figure 4.11. The HAGB disorientation-angle distributions show main populations near 43–46°, with a sharper population in SA1100.

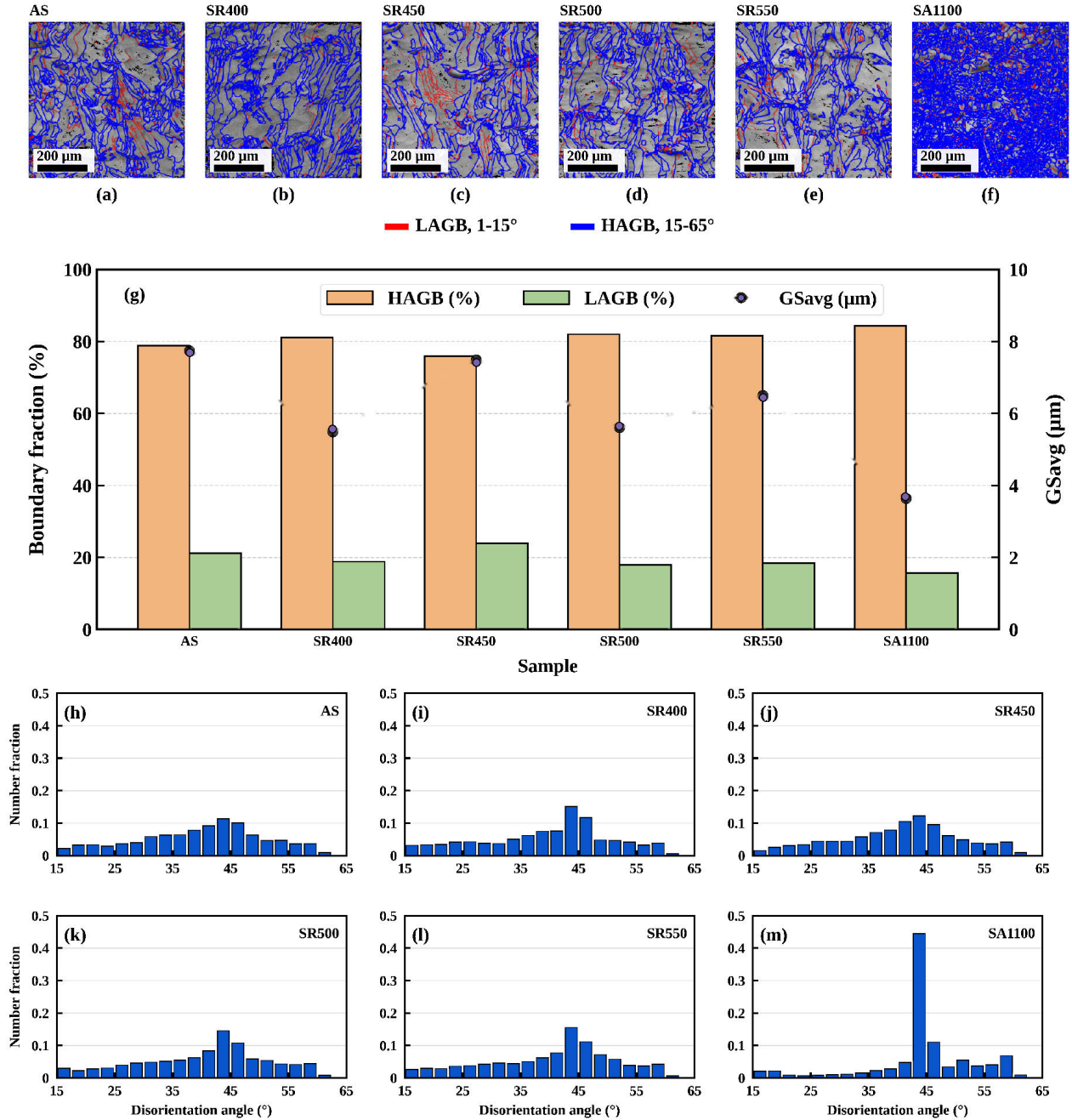


Figure 4.14: Boundary-network analysis of LPBF-fabricated SDSS 2507 +3 wt.% Ni in the AS, SR400, SR450, SR500, SR550, and SA1100 conditions. Panels (a–f) show maps overlaid with LAGBs and HAGBs, panel (g) summarizes LAGB/HAGB fractions and mapped-area grain-size descriptors, and panels (h–m) show HAGB disorientation-angle distributions.

The boundary-network response of the SDSS 2507 +6 wt.% Ni alloy was evaluated for the as-built (AS), stress-relieved at 350 °C for 2 h (SR350), stress-relieved at 425 °C for 1 h (SR425), stress-relieved at 525 °C for 1 h (SR525), and solution-annealed at 1150 °C for 10 min (SA1150) conditions. The boundary maps distinguish low-angle grain boundaries (LAGBs, 1–15°) from high-angle grain boundaries (HAGBs, 15–65°), as presented in Figure 4.15(a–e). HAGBs exceed LAGBs in all conditions, with AS and SR350 showing HAGB fractions of approximately 85% and 83%, respectively. SR425 and SA1150 show HAGB fractions of approximately 79% and 78%, whereas SR525 shows the lowest HAGB fraction, approximately 46% (Figure 4.15(f)). The mapped-area grain-size descriptors are 3.59, 3.78, 4.08, 3.18, and 4.01 μm for AS, SR350, SR425, SR525, and SA1150, respectively. The HAGB disorientation-angle distributions contain dominant populations near 43–46° across the investigated conditions (Figure 4.15(g–k)).

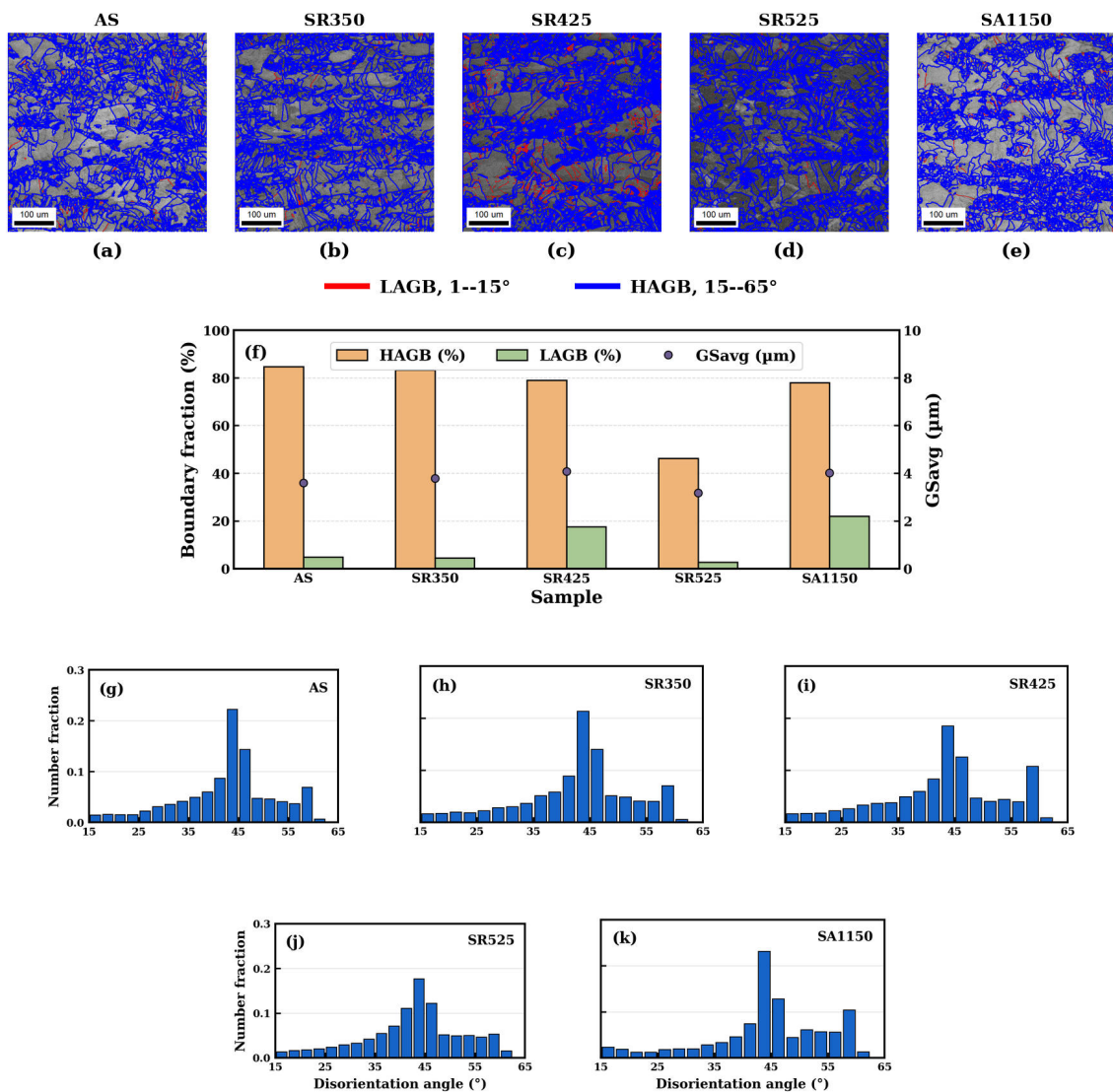


Figure 4.15: Boundary-network analysis of LPBF-fabricated SDSS 2507 +6 wt.% Ni in the AS, SR350, SR425, SR525, and SA1150 conditions. Panels (a–e) show maps overlaid with LAGBs and HAGBs, panel (f) summarizes LAGB/HAGB fractions and mapped-area grain-size descriptors, and panels (g–k) show HAGB disorientation-angle distributions.

Same-phase near-coincident site lattice (near-CSL) boundary distributions for the standard and +3 wt.% Ni alloys are shown in Figures 4.16 and 4.17, respectively. For the standard alloy, the near-CSL maps include AS, SR300, SR550, and SA1100. For the +3 wt.% Ni alloy, the near-CSL maps include AS, SR400, SR450, SR500, SR550, and SA1100. In the +3 wt.% Ni alloy, the $\Sigma 3$ boundary-length fraction is 2.59% in AS and 11.68% in SA1100, while the total near-CSL fraction is 10.01% in AS and 19.65% in SA1100. These values are same-phase near-CSL boundary-length fractions and do not represent complete five-parameter grain-boundary character distributions.

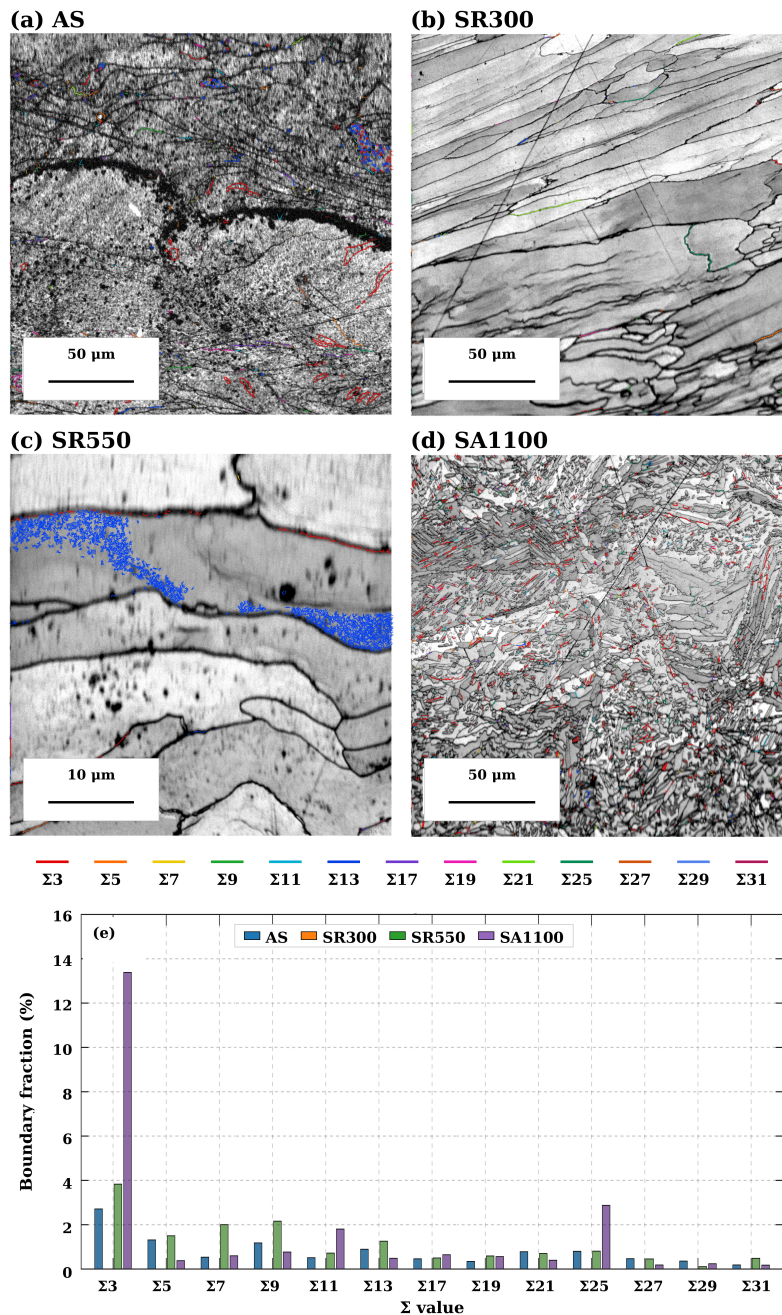


Figure 4.16: Same-phase near-CSL boundary distribution of standard LPBF-fabricated SDSS 2507 in the AS, SR300, SR550, and SA1100 conditions. Panels (a–d) show near-CSL boundary maps, and panel (e) shows the corresponding boundary-length fractions.

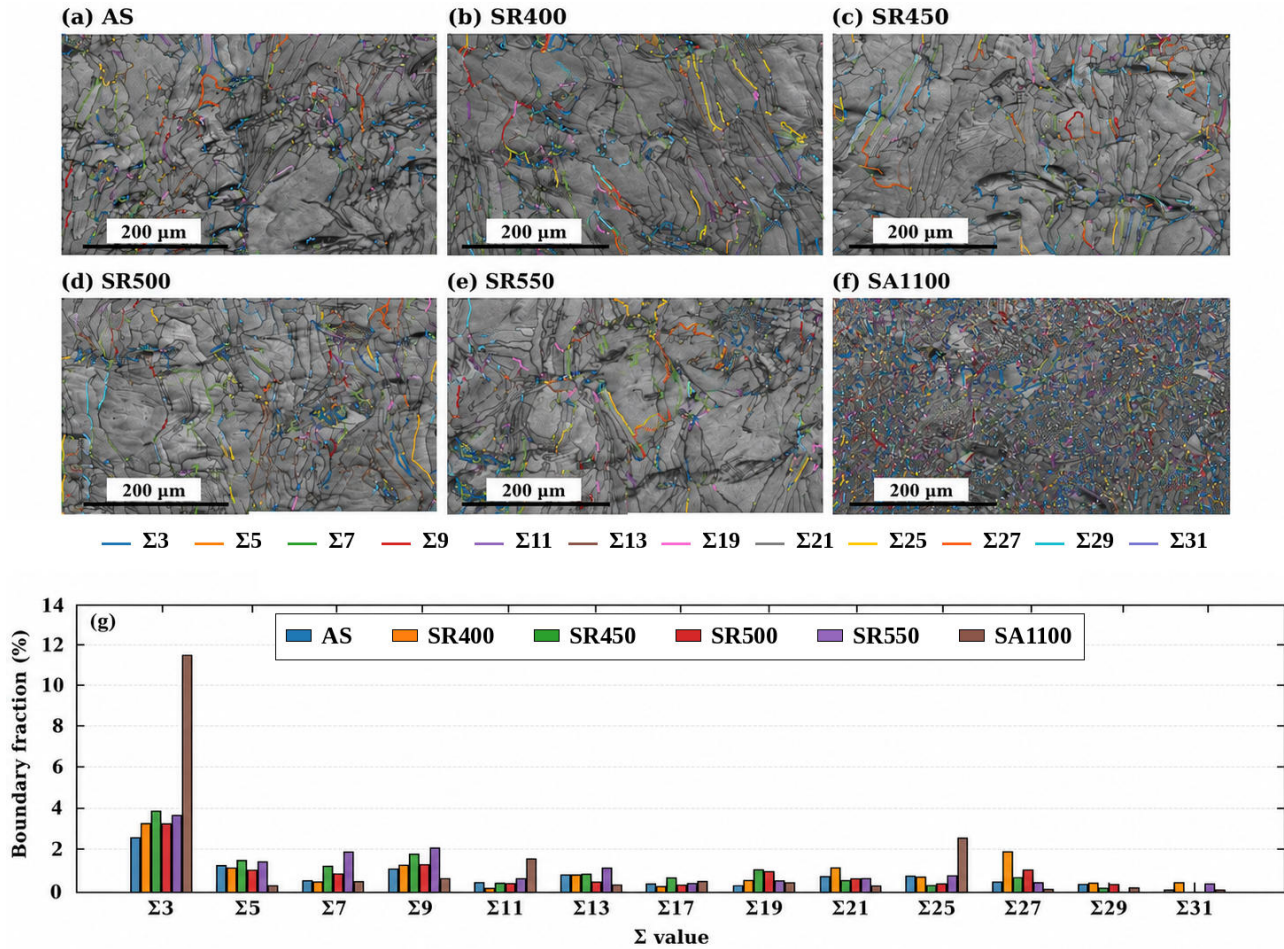


Figure 4.17: Same-phase near-CSL boundary analysis of LPBF-fabricated SDSS 2507 +3 wt.% Ni in the AS, SR400, SR450, SR500, SR550, and SA1100 conditions. Panels (a–f) show near-CSL boundary maps, and panel (g) shows the corresponding boundary-length fractions.

4.4.3 Ferrite–austenite interphase orientation relationships

The α/γ interphase-boundary segments classified as Kurdjumov–Sachs (KS)- or Nishiyama–Wassermann (NW)-compatible are compared in Figure 4.18 for the standard and +3 wt.% Ni laser powder bed fusion (LPBF)-fabricated super duplex stainless steel (SDSS) 2507 alloys. For the standard alloy, the maps include the as-built (AS), stress-relieved at 300 °C (SR300), stress-relieved at 550 °C (SR550), and solution-annealed at 1100 °C (SA1100) conditions. For the +3 wt.% Ni alloy, the maps include the as-built (AS), stress-relieved at 400 °C for 1 h (SR400), stress-relieved at 450 °C for 1 h (SR450), stress-relieved at 500 °C for 1 h (SR500), stress-relieved at 550 °C for 1 h (SR550), and solution-annealed at 1100 °C for 15 min (SA1100) conditions. In the standard alloy, AS, SR300, and SR550 show sparse KS/NW-compatible α/γ boundary segments, while SA1100 shows a larger mapped segment population. In the +3 wt.% Ni alloy, AS and the stress-relieved states show discontinuous KS/NW-compatible α/γ boundary segments across the mapped areas. SA1100 shows the largest mapped population of KS/NW-compatible segments in this alloy series. These maps are reported as tolerance-based interphase-boundary descriptors.

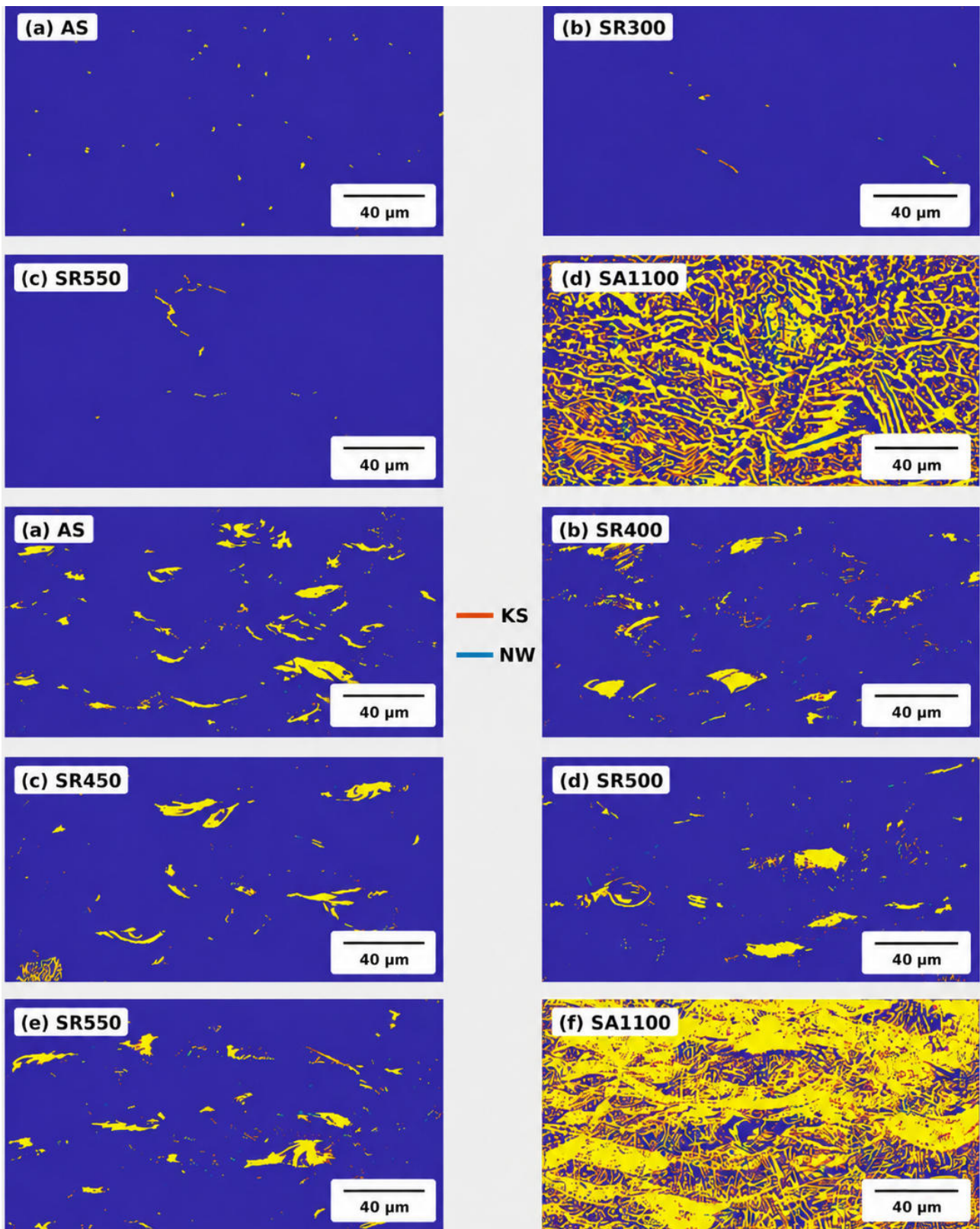


Figure 4.18: α/γ interphase-boundary orientation-relationship maps of LPBF-fabricated SDSS 2507 alloys. The upper block shows the standard alloy in AS, SR300, SR550, and SA1100. The lower block shows the +3 wt.% Ni alloy in AS, SR400, SR450, SR500, SR550, and SA1100. Orange denotes KS-compatible segments, and blue denotes NW-compatible segments.

For the standard alloy, the corresponding α/γ interphase-boundary misorientation distributions and minimum angular deviations from the nearest ideal KS or NW variant are summarized in Figure 4.19. The AS condition shows a sparse and scattered α/γ boundary-segment population. SR300 and SR550 show broader misorientation populations, mainly within the 35–55° range. The SA1100 distribution is narrower and centred near 43–46°. The minimum-deviation distributions vary with condition, with SA1100 showing the largest low-deviation population, mainly below approximately 5°.

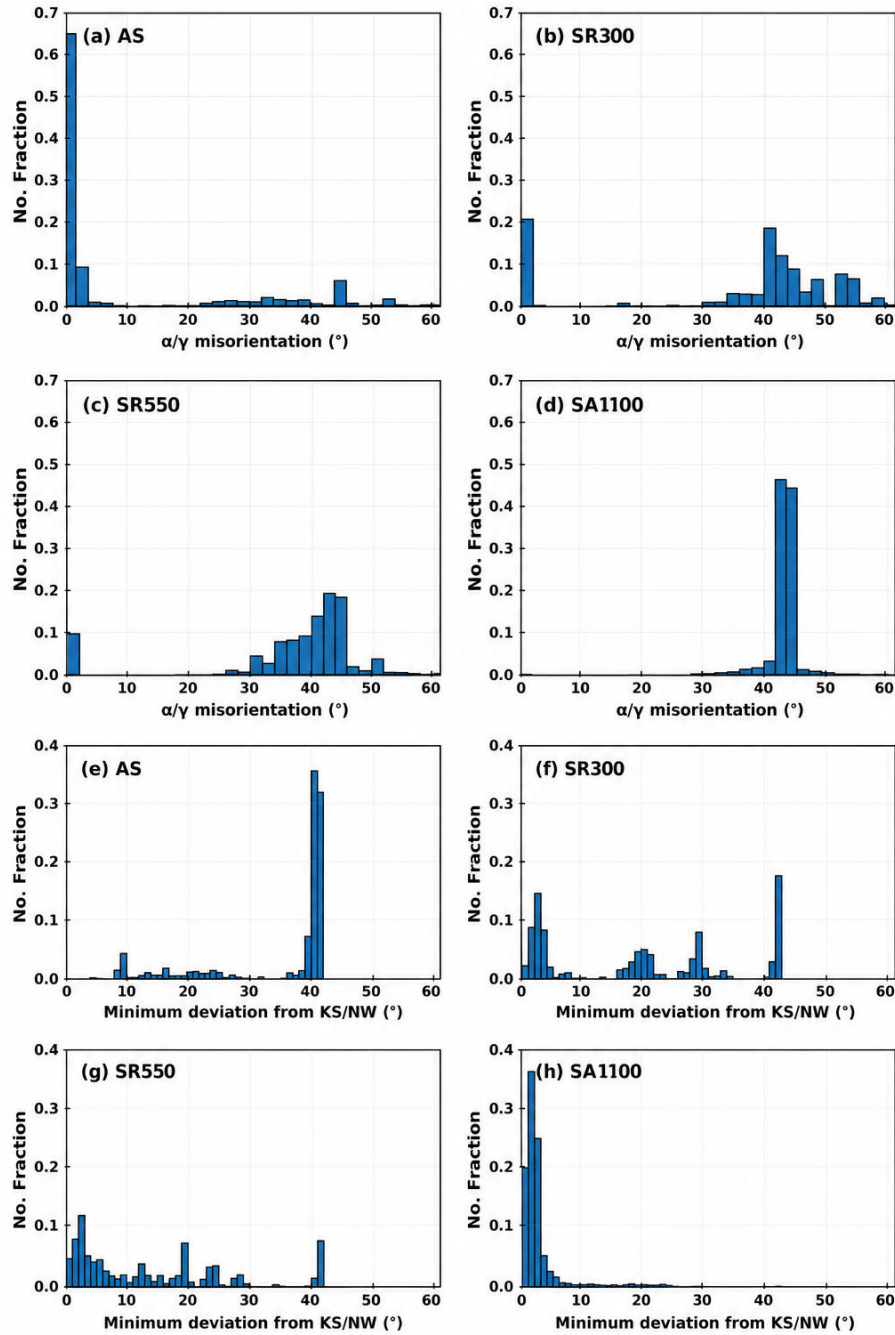


Figure 4.19: α/γ interphase-boundary misorientation and KS/NW minimum-deviation distributions for standard LPBF-fabricated SDSS 2507 in the AS, SR300, SR550, and SA1100 conditions. Panels (a–d) show α/γ misorientation distributions, and panels (e–h) show minimum angular deviations from the nearest KS or NW variant.

The +3 wt.% Ni alloy α/γ interphase-boundary misorientation and minimum-deviation distributions are shown in Figure 4.20. The AS and stress-relieved states show α/γ misorientation distributions mainly within the 35–50° range, with local maxima near 43–46°. SA1100 shows the narrowest distribution within this angular range. The minimum-deviation distributions show mixed low- and higher-deviation populations in AS and the stress-relieved states, while SA1100 shows the largest low-deviation population. These histograms are reported as mapped-area, tolerance-based interphase-boundary descriptors.

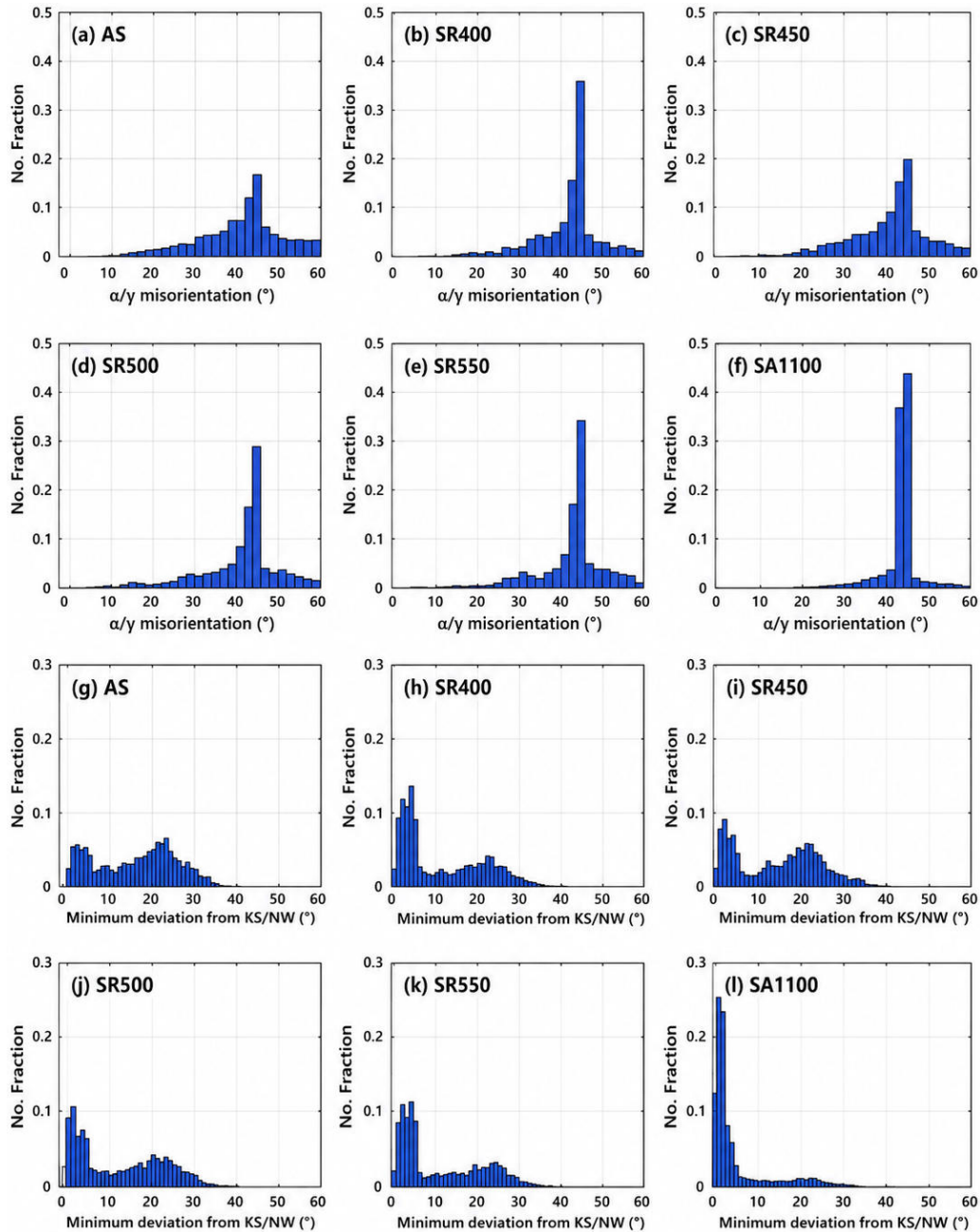


Figure 4.20: α/γ interphase-boundary misorientation and KS/NW minimum-deviation distributions for LPBF-fabricated SDSS 2507 +3 wt.% Ni in the AS, SR400, SR450, SR500, SR550, and SA1100 conditions. Panels (a–f) show α/γ misorientation distributions, and panels (g–l) show minimum angular deviations from the nearest KS or NW variant.

4.4.4 KAM-based local orientation heterogeneity

The kernel average misorientation (KAM) maps, and number-fraction distributions in Figure 4.21 show the standard LPBF-fabricated SDSS 2507 alloy in the AS, SR300, SR550, and SA1100 conditions. AS contains localized higher-KAM regions mainly along mapped boundaries and banded features. SR300 and SR550 show predominantly low-KAM fields with narrow linear higher-KAM regions, whereas SA1100 shows more spatially distributed KAM contrast. The calculated pixel-mean KAM values are 0.733° , 0.426° , 0.327° , and 0.444° for AS, SR300, SR550, and SA1100, respectively. The corresponding fractions of indexed pixels below 0.5° are approximately 23.4%, 78.6%, 92.0%, and 74.1%.

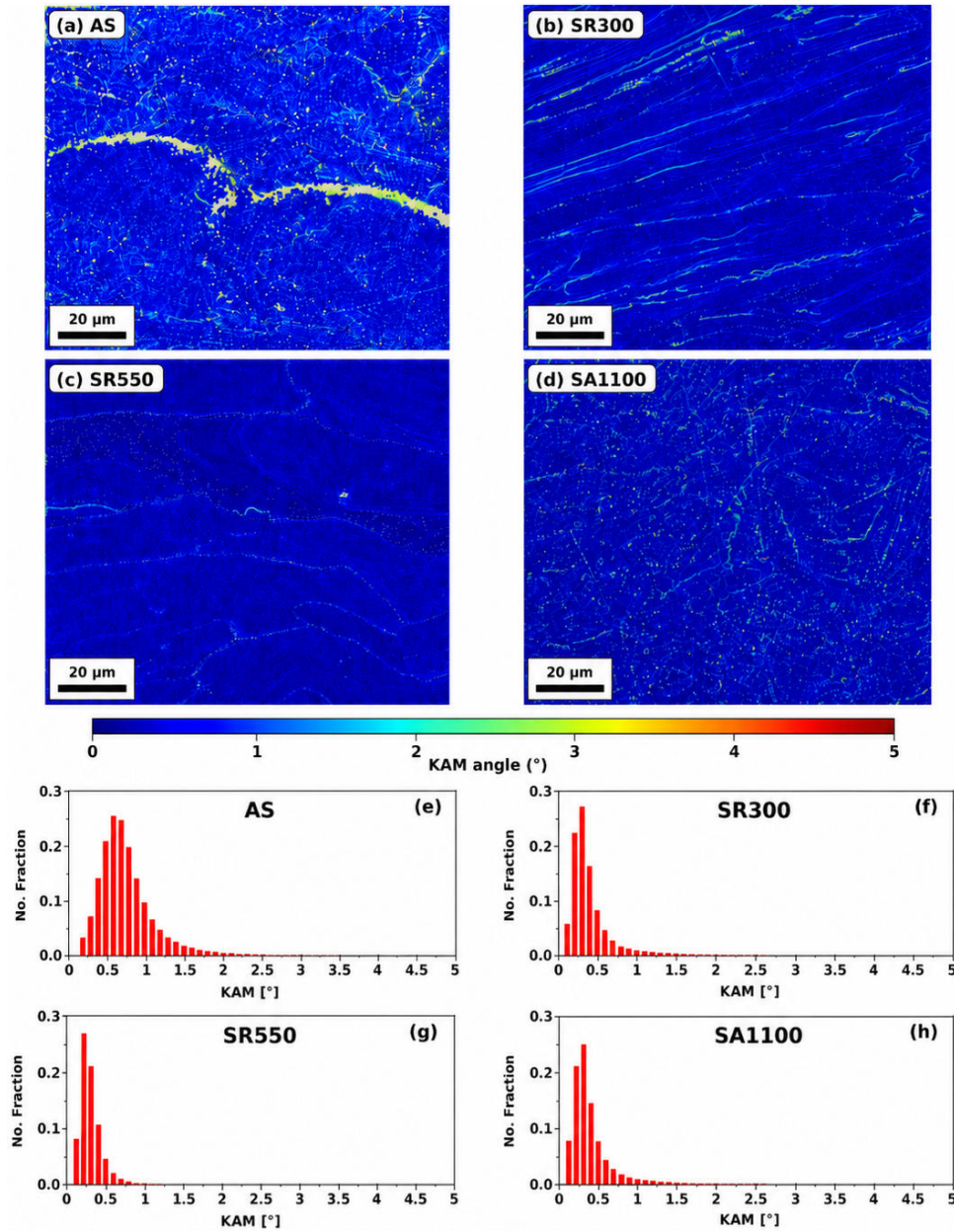


Figure 4.21: KAM maps and number-fraction distributions of standard LPBF-fabricated SDSS 2507 in the AS, SR300, SR550, and SA1100 conditions. Panels (a–d) show KAM maps, and panels (e–h) show the corresponding number-fraction distributions.

For the SDSS 2507 +3 wt.% Ni alloy, the KAM maps and number-fraction distributions in Figure 4.22 cover AS, SR400, SR450, SR500, SR550, and SA1100. Low KAM values occupy most of the mapped areas in all conditions. The calculated pixel-mean KAM values are 0.159° , 0.169° , 0.186° , 0.164° , 0.153° , and 0.176° , respectively. The corresponding fractions of indexed pixels below 0.5° are approximately 97.5%, 97.3%, 96.4%, 97.3%, 97.8%, and 96.2%. SR450 gives the highest pixel-mean KAM, whereas SR550 gives the lowest.

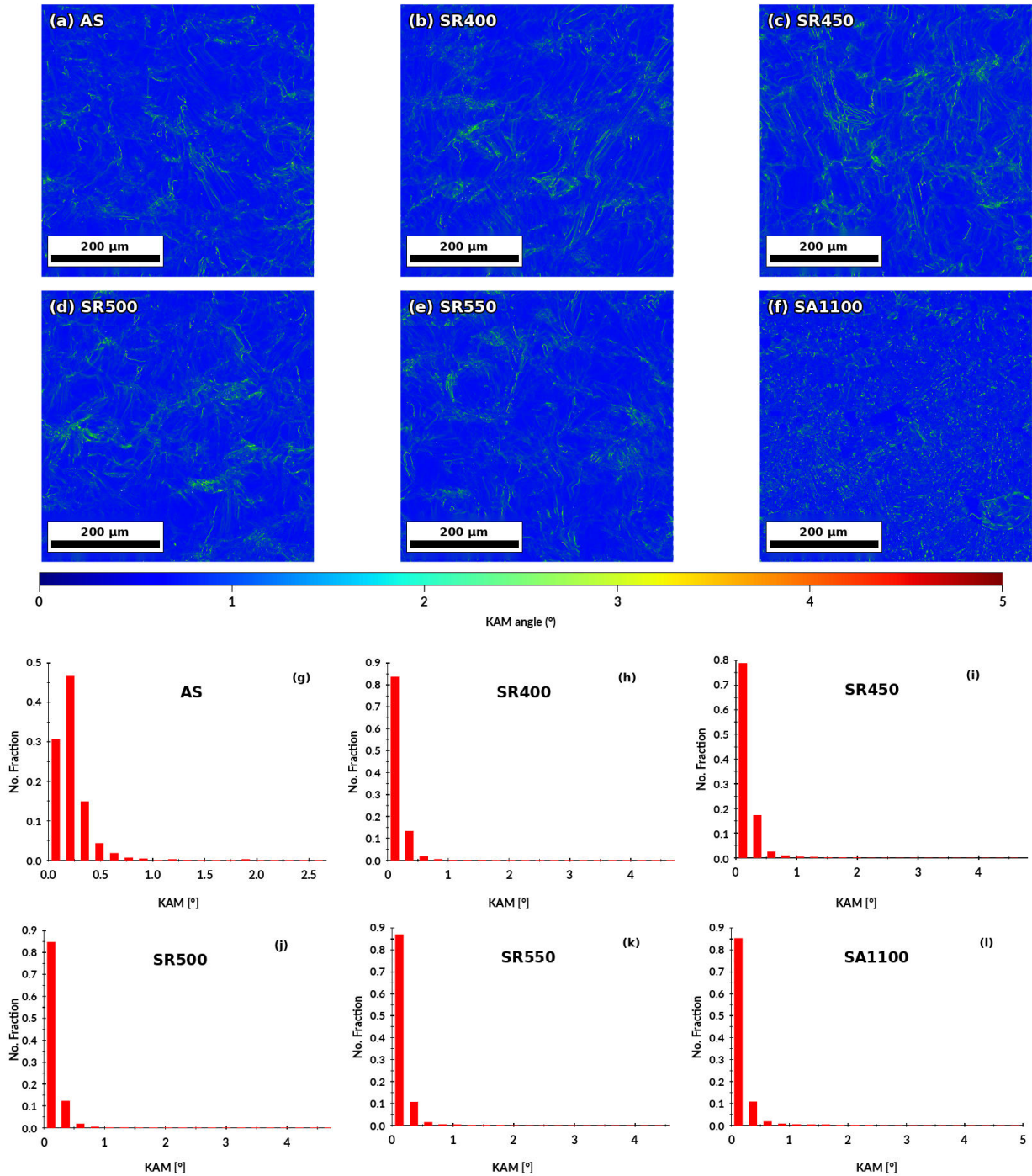


Figure 4.22: KAM maps and number-fraction distributions of LPBF-fabricated SDSS 2507 +3 wt.% Ni in the AS, SR400, SR450, SR500, SR550, and SA1100 conditions. Panels (a–f) show KAM maps, and panels (g–l) show the corresponding number-fraction distributions.

For the SDSS 2507 +6 wt.% Ni alloy, the KAM maps and number-fraction distributions in Figure 4.23 cover AS, SR350, SR425, SR525, and SA1150. The fractions of indexed pixels below 0.5° are approximately 95%, 94%, 92%, 60%, and 95%, respectively. The pixel-mean KAM values are 0.226° , 0.235° , 0.262° , 0.539° , and 0.214° , respectively. SR525 shows the broadest KAM distribution and the highest pixel-mean KAM value.

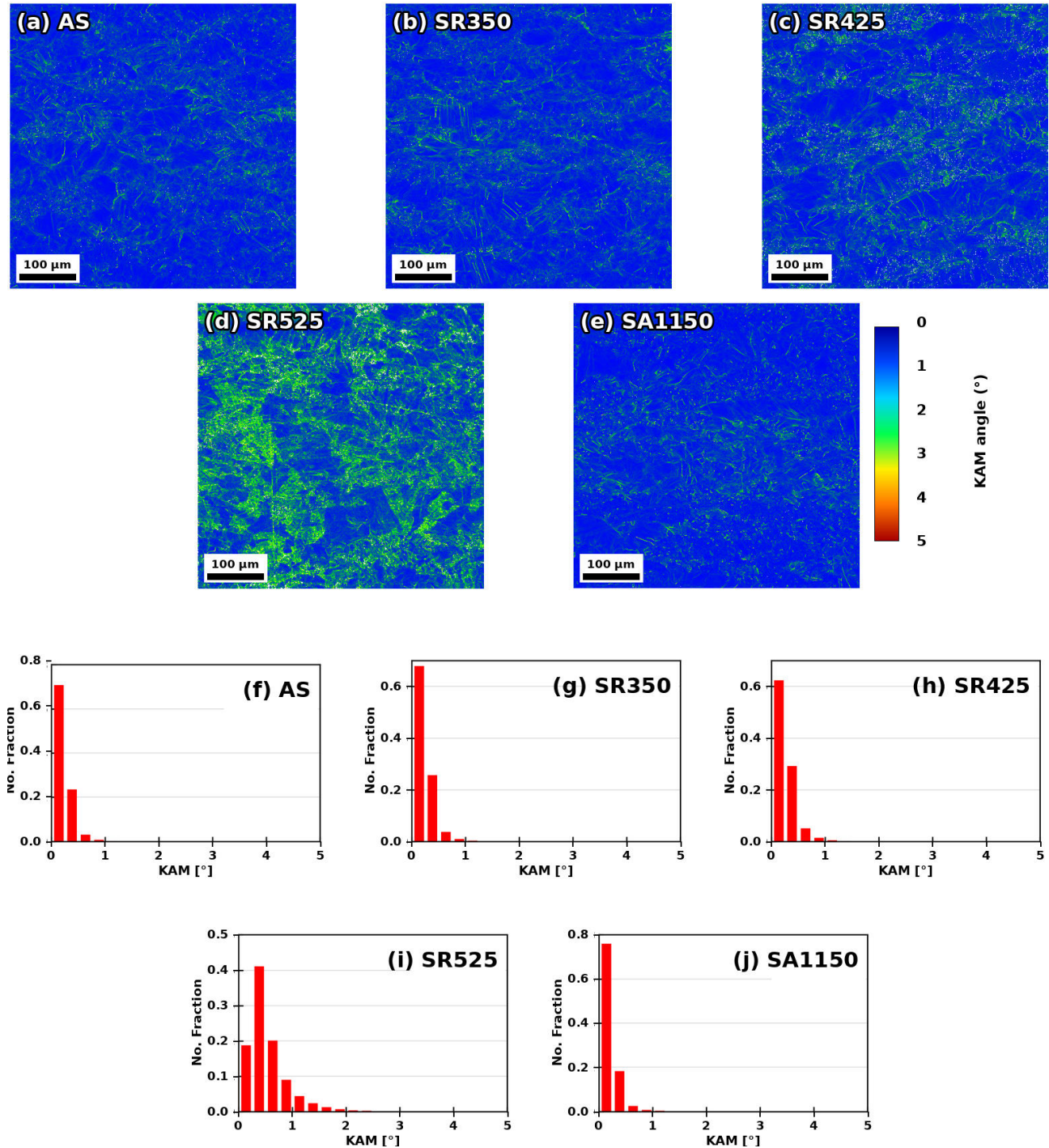


Figure 4.23: KAM maps and number-fraction distributions of LPBF-fabricated SDSS 2507 +6 wt.% Ni in the AS, SR350, SR425, SR525, and SA1150 conditions. Panels (a–e) show KAM maps using a common 0–5° colour scale, and panels (f–j) show the corresponding KAM number-fraction distributions.

4.4.5 GROD/GOS-based intragranular orientation heterogeneity

Figures 4.24–4.27 present the grain reference orientation deviation (GROD) and grain orientation spread (GOS) results for the standard and +3 wt.% Ni laser powder bed fusion (LPBF)-fabricated super duplex stainless steel (SDSS) 2507 alloys. The maps show the spatial distribution of intragranular orientation-spread values, and the cumulative and number-fraction plots summarize the corresponding angular distributions. For the standard alloy, Figure 4.24 shows the GROD maps and cumulative distributions for the as-built (AS), stress-relieved at 300 °C (SR300), stress-relieved at 550 °C (SR550), and solution-annealed at 1100 °C (SA1100) conditions. The maps show different GROD contrast levels across the four conditions. The cumulative distributions indicate that most measured GROD values lie in the lower angular range, with the curves approaching unity below approximately 10°. The corresponding GOS maps and histograms in Figure 4.25 show mainly low GOS values, with most grains below approximately 2°.

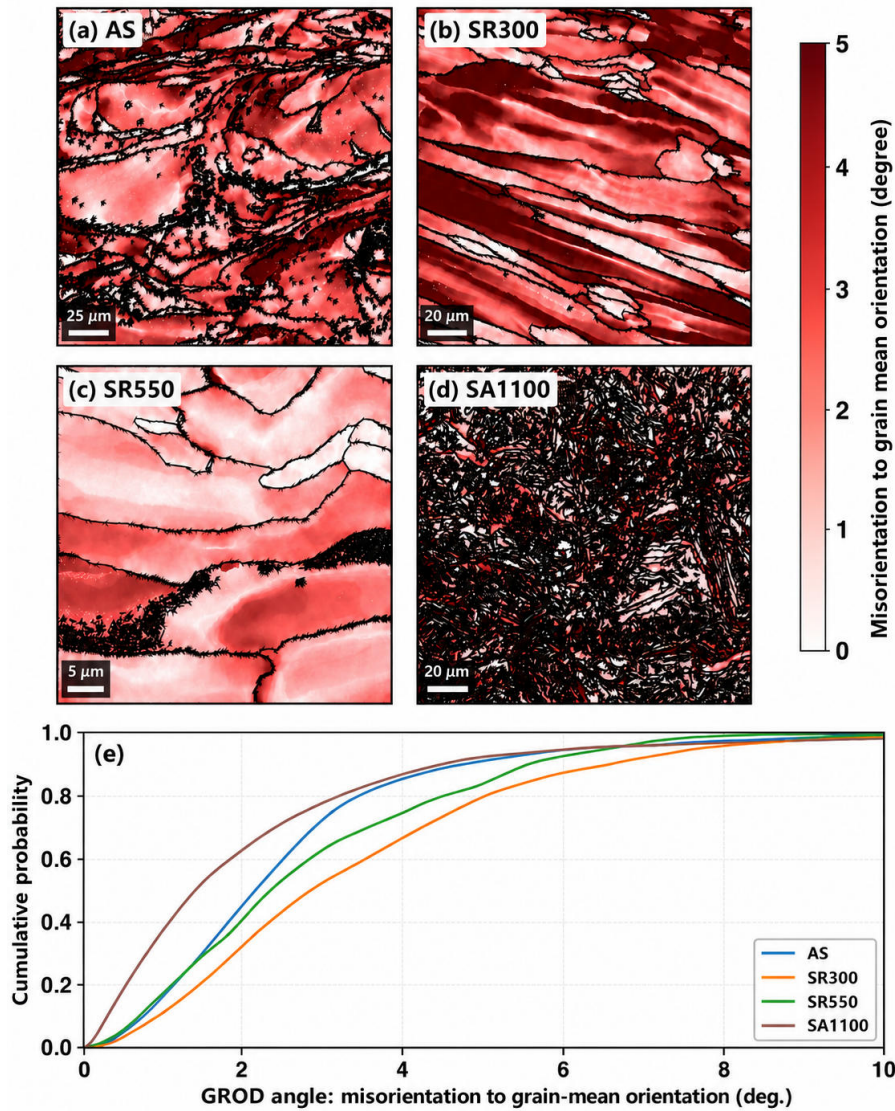


Figure 4.24: GROD angle maps and cumulative distributions of standard LPBF-fabricated SDSS 2507 in the AS, SR300, SR550, and SA1100 conditions. Panels (a–d) show GROD maps, and panel (e) shows the cumulative distributions.

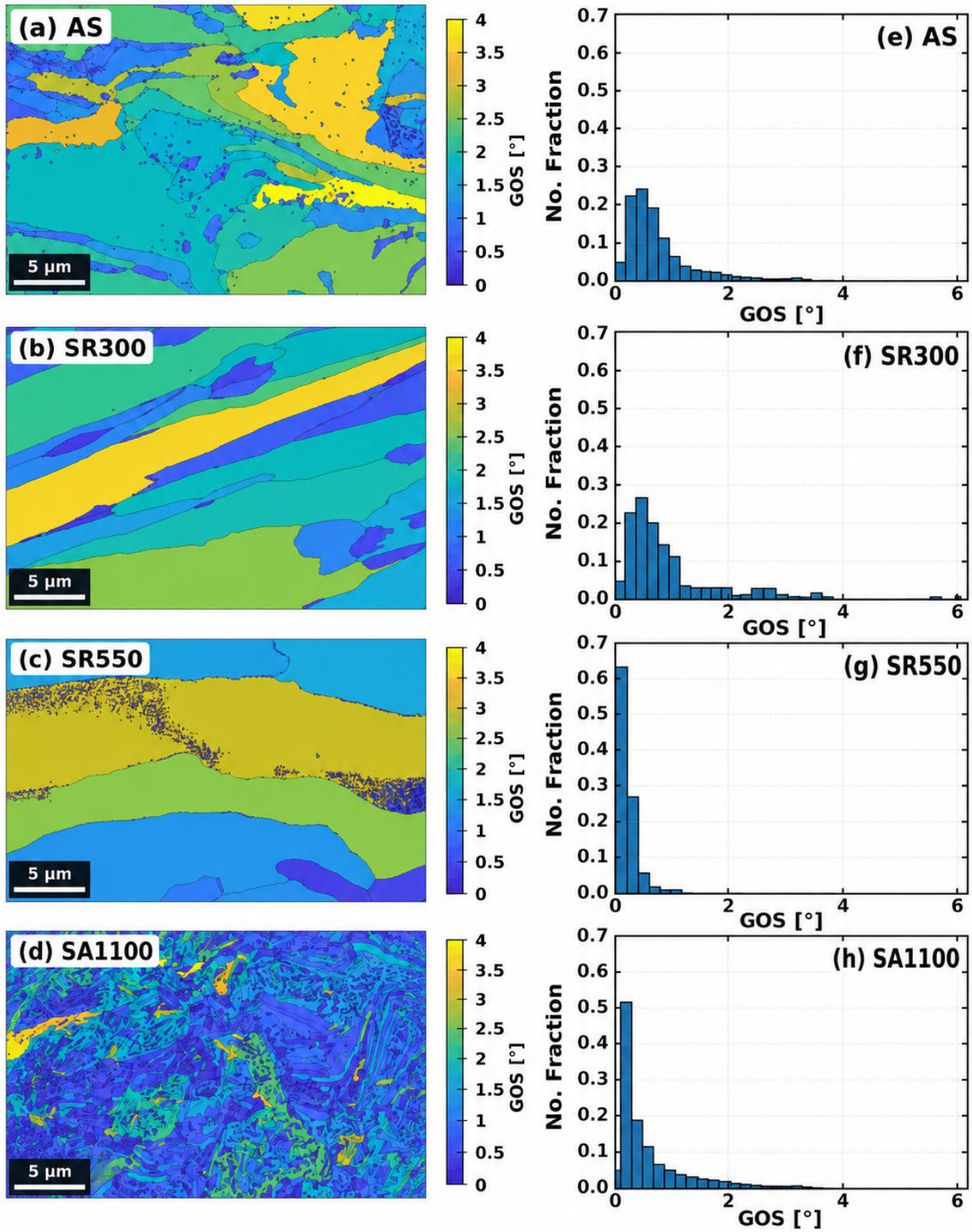


Figure 4.25: GOS maps and number-fraction distributions of standard LPBF-fabricated SDSS 2507 in the AS, SR300, SR550, and SA1100 conditions. Panels (a–d) show GOS maps, and panels (e–h) show the corresponding distributions.

For the +3 wt.% Ni alloy, Figure 4.26 shows the GROD maps and cumulative distributions for AS, stress-relieved at 400 °C for 1 h (SR400), stress-relieved at 450 °C for 1 h (SR450), stress-relieved at 500 °C for 1 h (SR500), stress-relieved at 550 °C for 1 h (SR550), and solution-annealed at 1100 °C for 15 min (SA1100). At approximately 2°, the cumulative probability is approximately 0.73 for SR550, 0.71 for SA1100, and 0.44 for SR500, with the remaining conditions between these values. The GOS maps and number-fraction distributions in Figure 4.27 show mainly low angular values in all conditions. Most grains are below approximately 2° in AS and the stress-relieved states, while the SA1100 distribution is mainly below approximately 1°. The GROD and GOS results are reported as mapped-area descriptors of intragranular orientation spread. They are not treated here as direct measurements of residual stress or dislocation density.

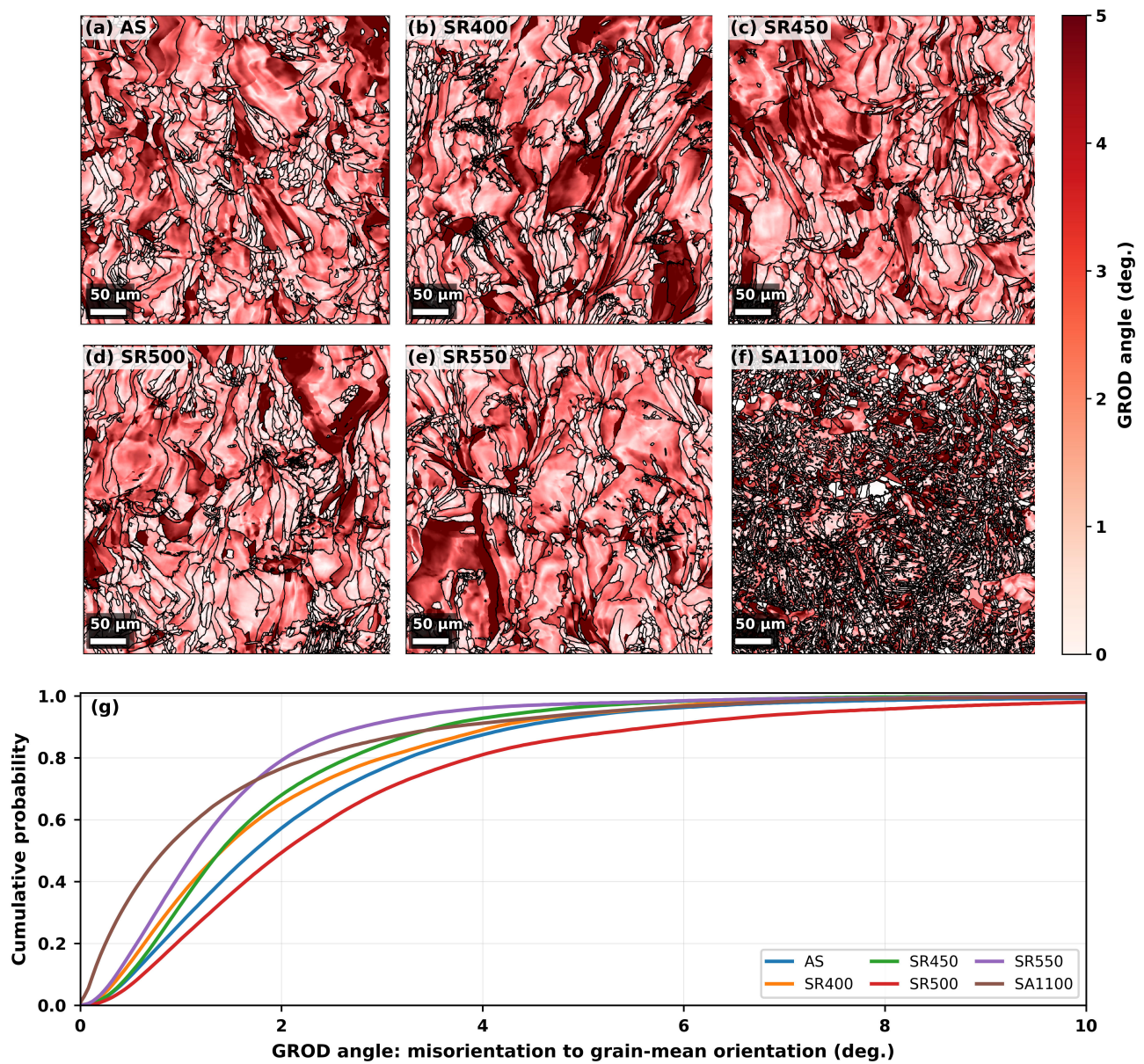


Figure 4.26: GROD angle maps and cumulative distributions of LPBF-fabricated SDSS 2507 +3 wt.% Ni in the AS, SR400, SR450, SR500, SR550, and SA1100 conditions. Panels (a–f) show GROD maps, and panel (g) shows the cumulative distributions.

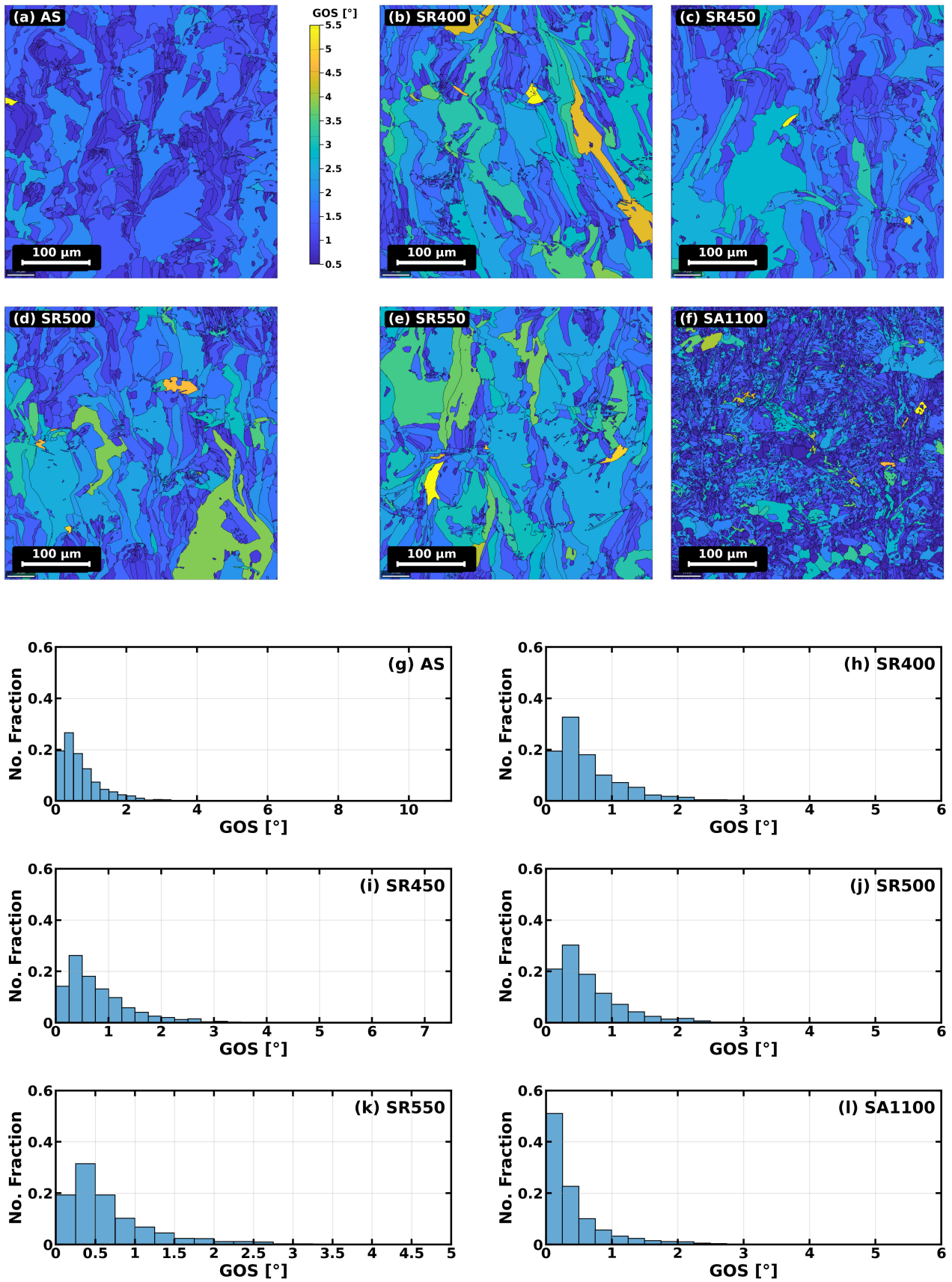


Figure 4.27: GOS maps and number-fraction distributions of LPBF-fabricated SDSS 2507 +3 wt.% Ni in the AS, SR400, SR450, SR500, SR550, and SA1100 conditions. Panels (a–f) show GOS maps, and panels (g–l) show the corresponding distributions.

4.4.6 Phase-resolved texture descriptors

Phase-resolved orientation distribution function (ODF) sections were used to report the mapped α -Fe and γ -Fe texture intensities in the standard and +3 wt.% Ni laser powder bed fusion (LPBF)-fabricated super duplex stainless steel (SDSS) 2507 alloys. The ODF sections are plotted at $\varphi_2 = 0^\circ, 45^\circ, \text{ and } 65^\circ$ for both phases. For the standard alloy shown in Figure 4.28, the mapped conditions are as-built (AS), stress-relieved at 300 °C (SR300), stress-relieved at 550 °C (SR550), and solution-annealed at 1100 °C (SA1100). The ferrite ODF sections show different intensity distributions among the mapped conditions. In SA1100, both α -Fe and γ -Fe ODF sections are presented because both phases are sufficiently represented in the mapped area.

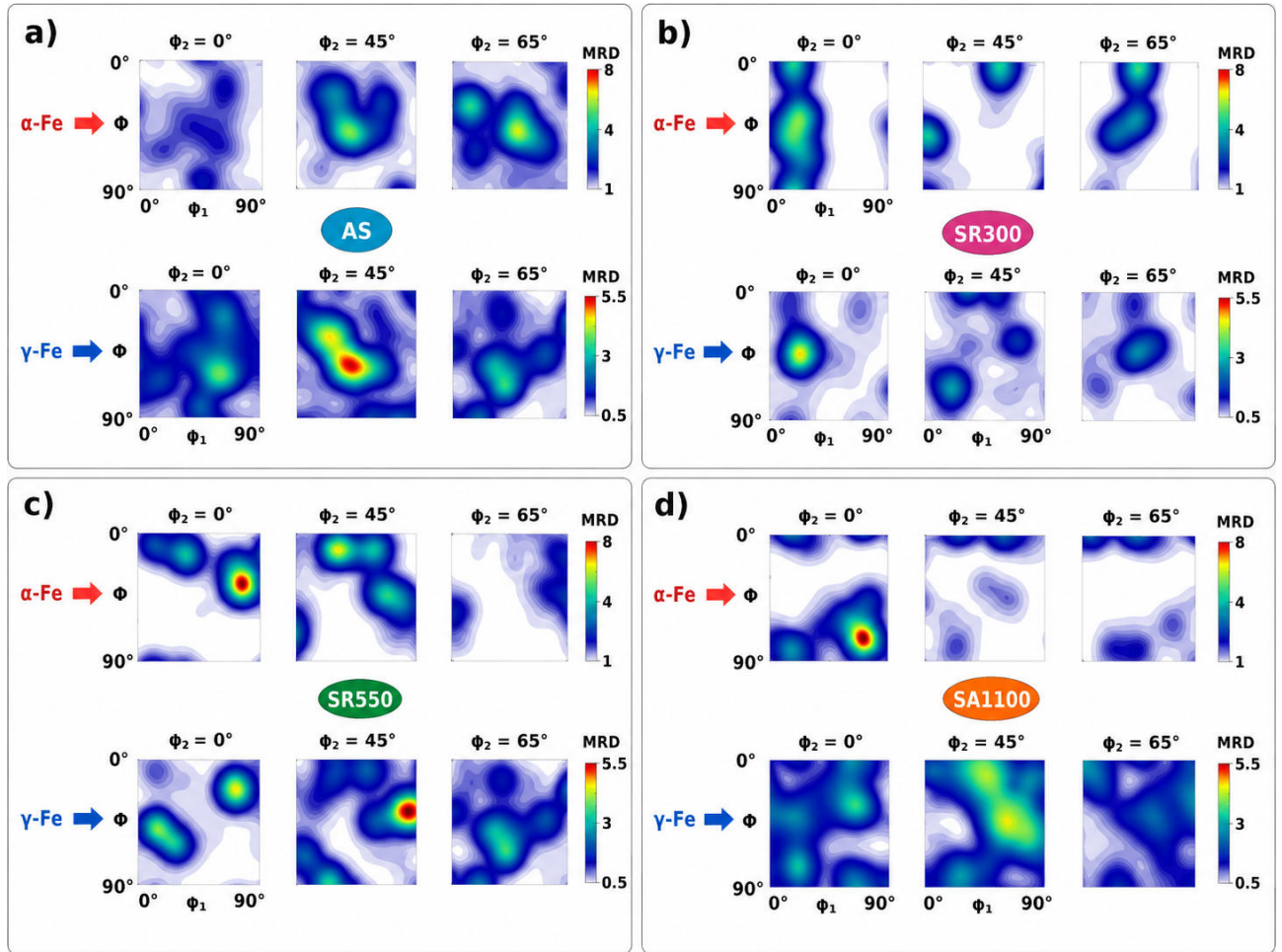


Figure 4.28: Phase-resolved ODF sections of standard LPBF-fabricated SDSS 2507 in (a) AS, (b) SR300, (c) SR550, and (d) SA1100 conditions. Ferrite and austenite ODF sections are shown at $\varphi_2 = 0^\circ, 45^\circ, \text{ and } 65^\circ$.

For the +3 wt.% Ni alloy, the mapped conditions are AS, stress-relieved at 400 °C for 1 h (SR400), stress-relieved at 450 °C for 1 h (SR450), stress-relieved at 500 °C for 1 h (SR500), stress-relieved at 550 °C for 1 h (SR550), and solution-annealed at 1100 °C for 15 min (SA1100) are depicted in Figure 4.29. The maximum ferrite texture intensities are 4.5, 5.5, 8.0, 5.0, 6.0, and 4.5 multiples of random distribution (MRD), respectively. The corresponding maximum austenite texture intensities are 5.5, 3.5, 5.5, 6.0, 5.0, and 2.4 MRD. These values are reported as electron backscatter diffraction

(EBSD)-derived mapped-area texture descriptors for comparing phase-resolved texture intensity across the investigated conditions.

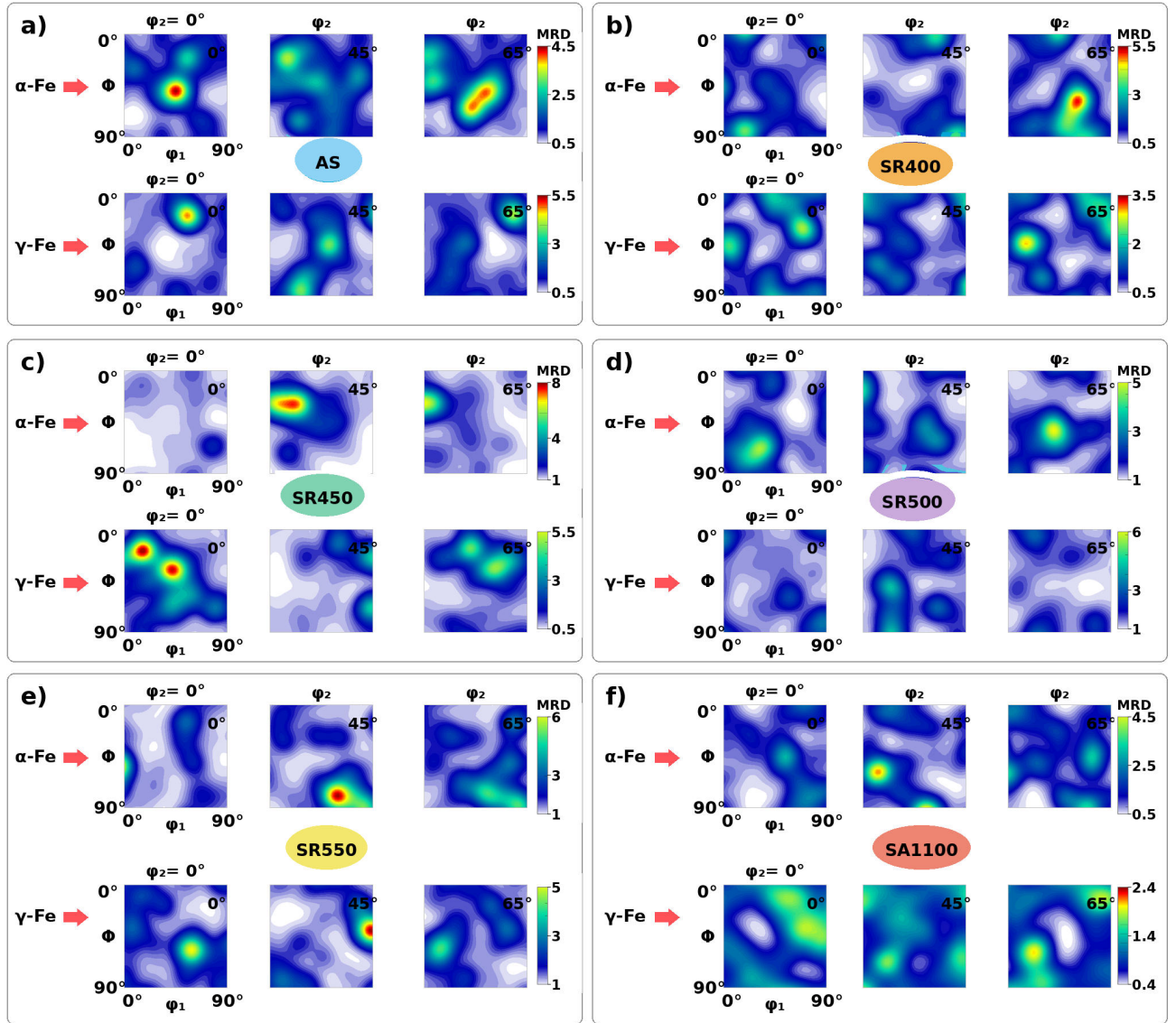


Figure 4.29: Phase-resolved ODF sections of LPBF-fabricated SDSS 2507 +3 wt.% Ni in (a) AS, (b) SR400, (c) SR450, (d) SR500, (e) SR550, and (f) SA1100 conditions. Ferrite and austenite ODF sections are shown at $\phi_2 = 0^\circ$, 45° , and 65° .

4.5 Nanoscale features

4.5.1 Transmission electron microscopy

Transmission electron microscopy (TEM), scanning transmission electron microscopy (STEM), selected-area electron diffraction (SAED), and STEM energy-dispersive X-ray spectroscopy (STEM-EDS) were used to examine the standard laser powder bed fusion (LPBF)-fabricated super duplex stainless steel (SDSS) 2507 alloy in the stress-relieved at 550 °C (SR550) condition. The analyzed results are shown in Figure 4.30.

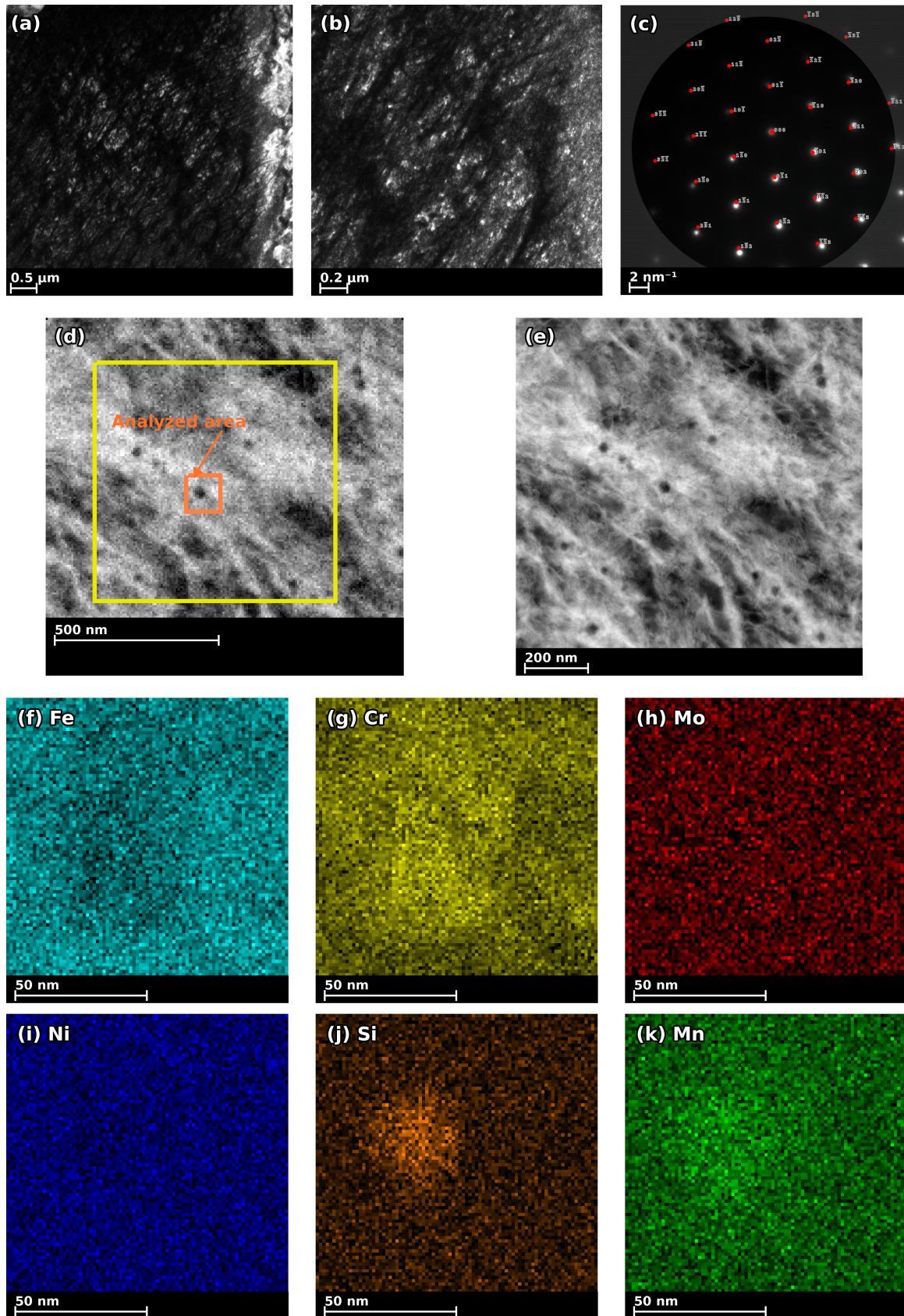


Figure 4.30: TEM, STEM, SAED, and STEM-EDS results for standard LPBF-fabricated SDSS 2507 in the stress-relieved at 550 °C (SR550) condition: (a) bright-field TEM image, (b) dark-field TEM image, (c) SAED pattern indexed to ferrite along the $[111]$ zone axis, (d) STEM image showing the analyzed region, (e) high-angle annular dark-field STEM image, and corresponding STEM-EDS maps of (f) Fe, (g) Cr, (h) Mo, (i) Ni, (j) Si, and (k) Mn.

Dense nanoscale contrast is visible within the ferritic matrix in the bright-field TEM image, Figure 4.30(a). The dark-field TEM image, Figure 4.30(b), shows heterogeneous contrast in the same region, but no discrete particle-like features are resolved. The SAED pattern, Figure 4.30(c), is indexed to ferrite along the $[111]$ zone axis, with no clearly resolved additional diffraction spots or ring features.

The analyzed STEM region is shown in Figure 4.30(d). The HAADF-STEM image, Figure 4.30(e), contains local nanoscale contrast heterogeneity, including small rounded dark-contrast features. The corresponding Fe, Cr, Mo, Ni, Si, and Mn maps are shown in Figure 4.30(f–k). Spatial intensity variation is visible in the Fe and Cr maps, while the Mo, Ni, and Mn maps do not show localized enrichment associated with the HAADF-STEM contrast features. The Si map shows a local higher-intensity region, but this region is not accompanied by correlated Cr, Mo, Ni, or Mn enrichment.

The examined TEM/STEM region shows nanoscale contrast within the ferritic matrix of the standard LPBF-fabricated SDSS 2507 alloy in the SR550 condition. The SAED pattern does not show additional clearly resolved diffraction spots or rings, and the available STEM-EDS maps do not show correlated Cr, Mo, Ni, or Mn enrichment associated with the observed contrast. Therefore, no unambiguous secondary precipitates are identified in the analysed region. This observation is limited to the analyzed standard SR550 TEM/STEM area and does not exclude nanoscale precipitation, clustering, or decomposition in other regions, heat-treatment conditions, or Ni-modified alloy systems.

4.6 Mechanical behavior

This section presents the hardness, tensile, and unnotched Charpy impact results of the standard SDSS 2507, SDSS 2507 +3 wt.% Ni, and SDSS 2507 +6 wt.% Ni alloy systems. The standard alloy includes the as-built (AS), stress-relieved at 300 °C (SR300), stress-relieved at 400 °C (SR400), stress-relieved at 550 °C (SR550), and solution-annealed at 1100 °C (SA1100) conditions. The +3 wt.% Ni alloy includes AS, stress-relieved at 400 °C for 1 h (SR400), stress-relieved at 450 °C for 1 h (SR450), stress-relieved at 500 °C for 1 h (SR500), stress-relieved at 550 °C for 1 h (SR550), and solution-annealed at 1100 °C for 15 min (SA1100). The +6 wt.% Ni alloy includes AS, stress-relieved at 350 °C for 2 h (SR350), stress-relieved at 425 °C for 1 h (SR425), stress-relieved at 525 °C for 1 h (SR525), and solution-annealed at 1150 °C for 10 min (SA1150).

4.6.1 Hardness response

Figure 4.31 compares the Rockwell hardness and Vickers microhardness values of the standard, +3 wt.% Ni, and +6 wt.% Ni LPBF-fabricated SDSS 2507 alloys. In the Rockwell hardness measurements, the core values are higher than the surface values in all alloy systems. The standard alloy shows the highest HRA values in AS and SR300, and the lowest values in SA1100. In the +3 wt.% Ni alloy, AS, and the stress-relieved states show comparable HRA values, while SA1100 gives the lowest values. The +6 wt.% Ni alloy shows the highest HRA values in AS and SR350, lower values in SR425 and SR525, and the lowest surface hardness in SA1150. The Vickers

microhardness results follow the same overall trend, with the lowest values recorded for the solution-annealed condition in each alloy system.

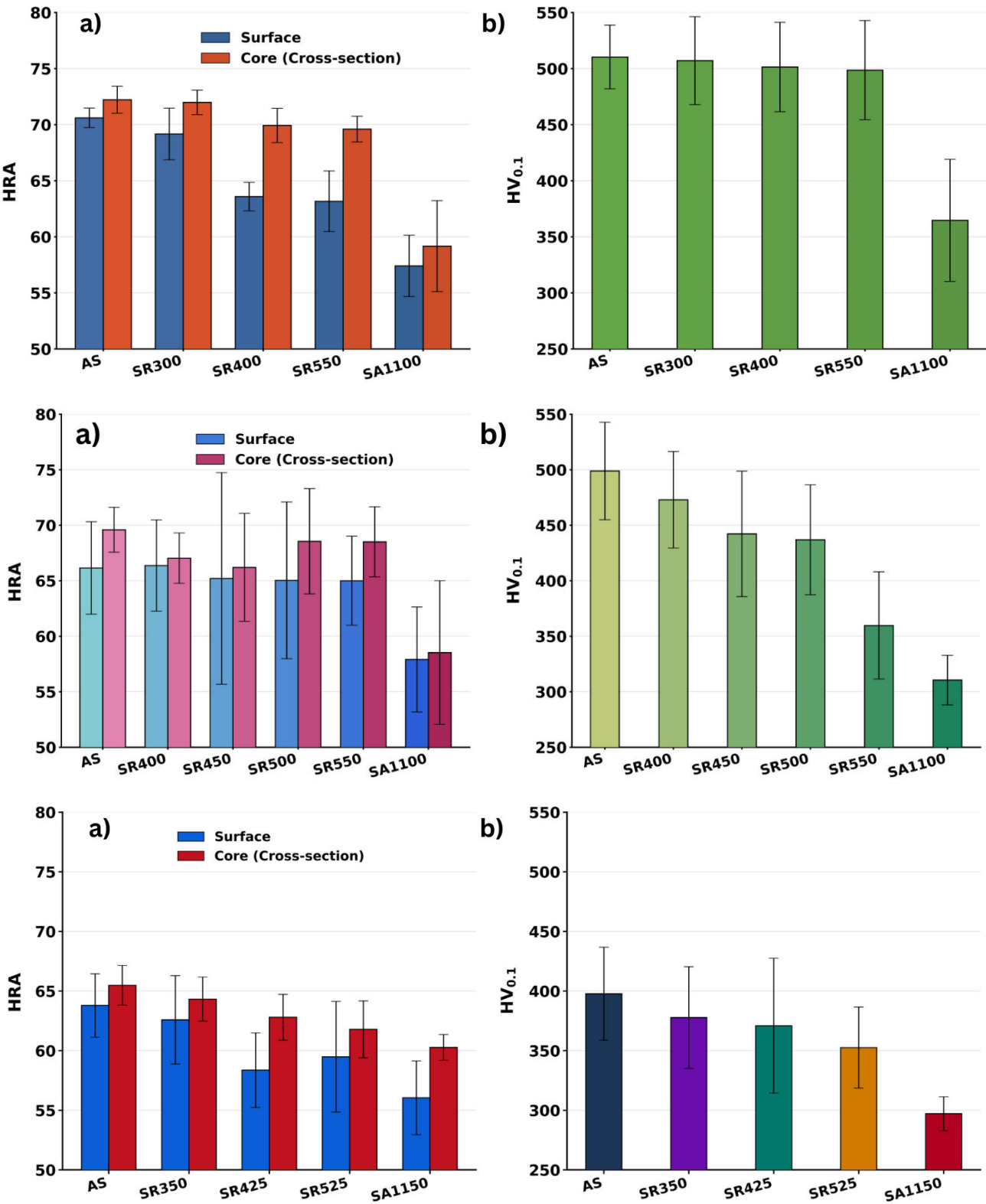


Figure 4.31: Hardness results of the standard, +3 wt.% Ni, and +6 wt.% Ni LPBF-fabricated SDSS 2507 alloy systems, shown from top to bottom.

4.6.2 Tensile and impact behavior

Figures 4.32–4.34 show the engineering stress–strain curves of the standard, +3 wt.% Ni, and +6 wt.% Ni LPBF-fabricated SDSS 2507 alloys, while Tables 4.4–4.6 summarize the corresponding tensile properties and unnotched Charpy impact energies.

In the standard alloy, the AS and stress-relieved conditions reach stress levels above approximately 1200 MPa. SR400 shows the shortest strain range, followed by AS, while SR300 and SR550 show higher strain values.

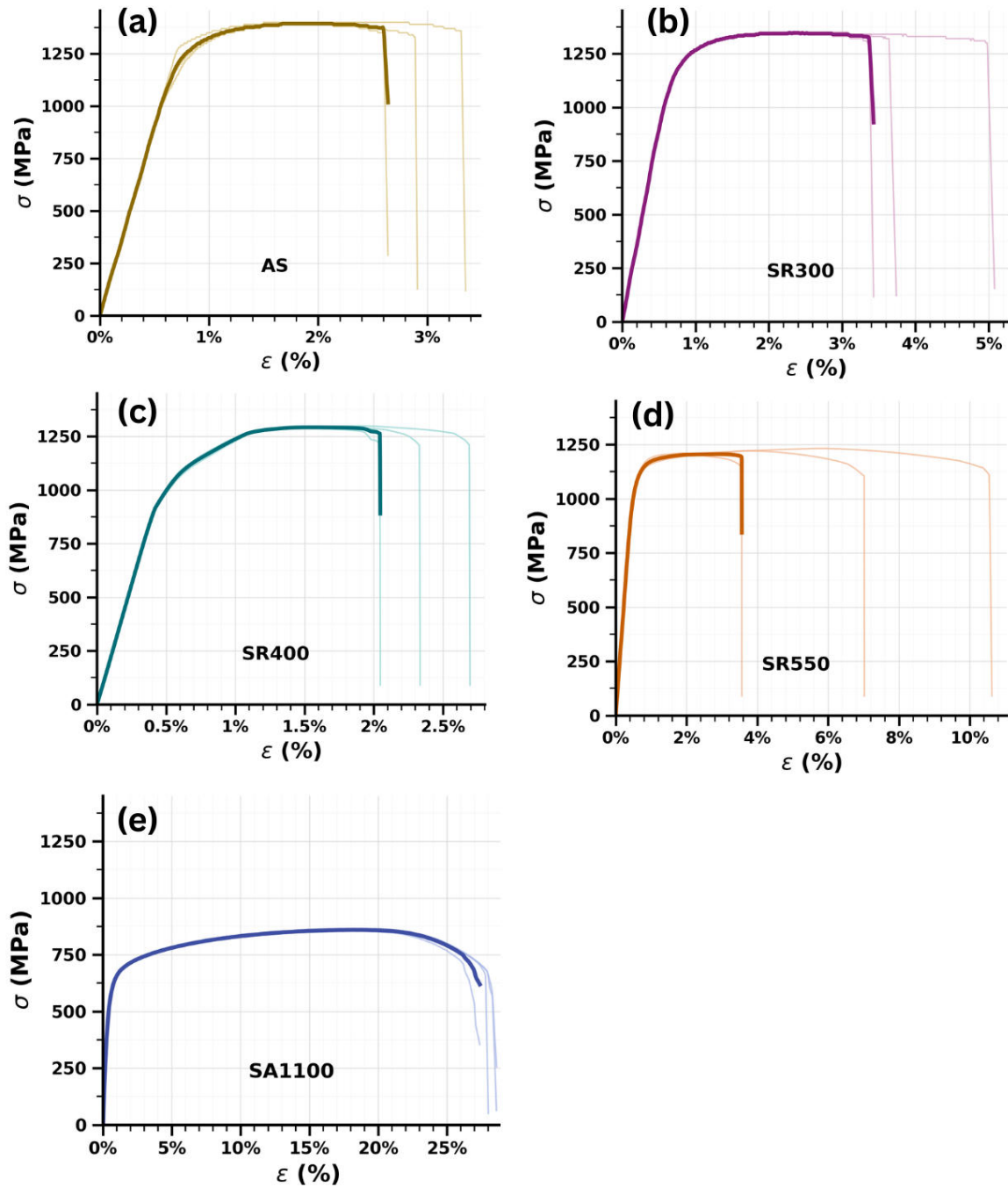


Figure 4.32: Engineering stress–strain curves of LPBF-fabricated standard SDSS 2507 in the (a) AS, (b) SR300, (c) SR400, (d) SR550, and (e) SA1100 conditions.

Table 4.4 summarizes the tensile properties and unnotched Charpy impact energy of the standard alloy. AS gives the highest yield strength and ultimate tensile strength, while SA1100 gives the lowest strength and the highest elongation and impact energy.

Table 4.4: Tensile properties and unnotched Charpy impact energy of LPBF-fabricated standard SDSS 2507.

Condition	YS [MPa]	UTS [MPa]	A [%]	Impact [J]
AS	1211 ± 108	1393 ± 5	2.35 ± 0.52	103 ± 26.67
SR300	1082 ± 162	1350 ± 8	3.18 ± 0.73	122.5 ± 38.89
SR400	1069 ± 25	1292 ± 14	2.0 ± 0.5	60.33 ± 5.77
SR550	1043.4 ± 61	1221 ± 16	6.0 ± 1.5	77 ± 4.24
SA1100	506 ± 75	861 ± 3	27.25 ± 0.96	450

In the +3 wt.% Ni alloy, AS and the stress-relieved conditions also reach stress levels above approximately 1200 MPa. SR400, SR450, and SR500 show comparable strain ranges, while SR550 shows the highest strain among the stress-relieved conditions.

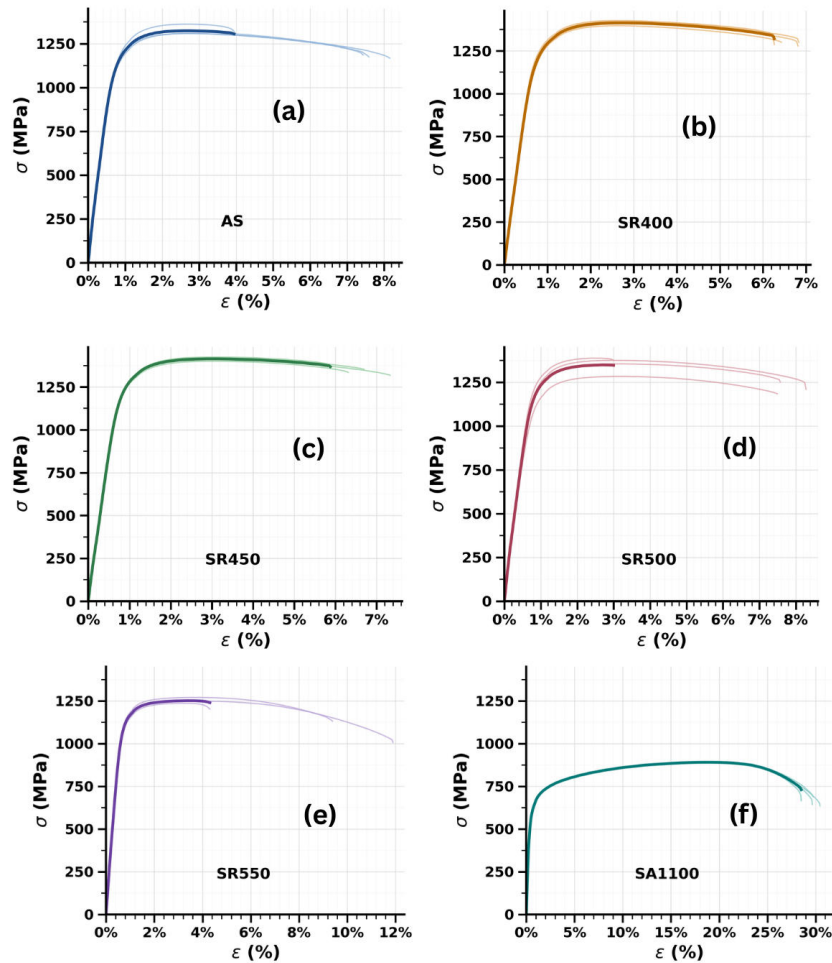


Figure 4.33: Engineering stress–strain curves of LPBF-fabricated SDSS 2507 +3 wt.% Ni in the (a) AS, (b) SR400, (c) SR450, (d) SR500, (e) SR550, and (f) SA1100 conditions.

Table 4.5 summarizes the tensile properties and unnotched Charpy impact energy of the +3 wt.% Ni alloy. SR400 and SR450 give the highest strength values, while SA1100 gives the lowest strength and the highest elongation and impact energy.

Table 4.5: Tensile properties and unnotched Charpy impact energy of LPBF-built SDSS 2507 +3 wt.% Ni.

Condition	Rp0.2 [MPa]	Rm [MPa]	A [%]	Impact [J]
AS	1094.61 ± 17.05	1324.31 ± 25.44	6.78 ± 1.92	168.71 ± 25.30
SR400	1210.25 ± 33.68	1413.74 ± 14.06	6.58 ± 0.28	142.00 ± 76.37
SR450	1210.25 ± 33.68	1414.09 ± 11.14	6.55 ± 0.61	98.67 ± 17.62
SR500	1119.32 ± 100.60	1350.03 ± 46.65	6.58 ± 2.42	178.33 ± 25.17
SR550	1090.22 ± 8.05	1223.70 ± 69.04	6.58 ± 5.02	202.67 ± 21.13
SA1100	559.54 ± 7.99	890.04 ± 1.97	29.59 ± 0.81	> 450

The +6 wt.% Ni alloy shows stress levels close to 1000 MPa and strain ranges of approximately 15–18% in AS and the stress-relieved states. SA1150 shows lower stress and higher strain to fracture than the corresponding AS and stress-relieved conditions.

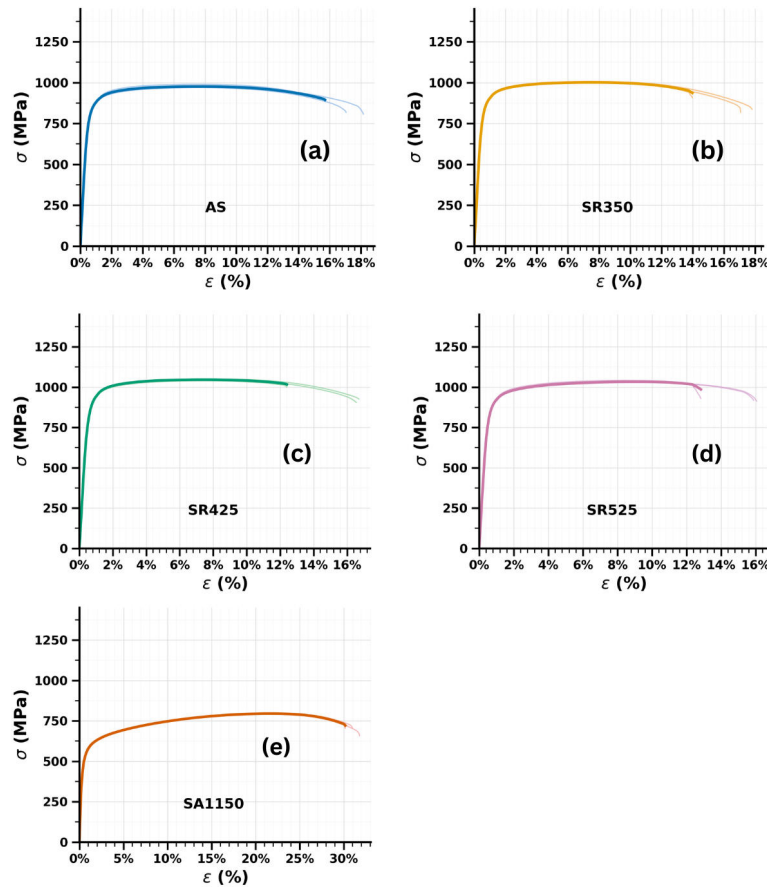


Figure 4.34: Engineering stress–strain curves of LPBF-fabricated SDSS 2507 +6 wt.% Ni in the (a) AS, (b) SR350, (c) SR425, (d) SR525, and (e) SA1150 conditions.

Table 4.6 summarizes the tensile properties and unnotched Charpy impact energy of the +6 wt.% Ni alloy. SR425 gives the highest yield strength and ultimate tensile strength. SR525 gives the lowest elongation, while SR425 gives the lowest unnotched Charpy impact energy. SA1150 gives the lowest strength and the highest elongation, with unnotched Charpy impact energy exceeding 450 J.

Table 4.6: Tensile properties and unnotched Charpy impact energy of LPBF-built SDSS 2507 +6 wt.% Ni.

Condition	YS [MPa]	UTS [MPa]	A [%]	Impact [J]
AS	809.86 ± 13.60	976.57 ± 11.48	16.97 ± 1.23	388.67 ± 83.79
SR350	843.11 ± 2.54	1001.51 ± 5.40	16.28 ± 2.03	403.33 ± 79.10
SR425	870.49 ± 3.76	1046.5 ± 15.2	15.20 ± 2.4	274.33 ± 23.09
SR525	853.61 ± 19.36	1033.51 ± 7.75	14.91 ± 1.81	341.67 ± 54.72
SA1150	480.11 ± 6.26	794.93 ± 3.59	30.99 ± 0.81	> 450

Table 4.7 summarizes the strength, ductility, and impact-energy ranking across the alloy series. The AS and stress-relieved conditions show higher strength values, while the solution-annealed conditions show the highest elongation and impact-energy values. The +6 wt.% Ni alloy shows lower strength than the standard and +3 wt.% Ni alloys, but higher elongation and impact energy in AS and the lower-temperature stress-relieved states.

Table 4.7: Mechanical trade-off ranking of LPBF-fabricated standard and Ni-modified SDSS 2507 alloys.

Alloy system	Condition	R_m / UTS (MPa)	A (%)	Impact energy (J)	Rank profile
<i>Standard SDSS 2507</i>					
	AS	1393 ± 5	2.35 ± 0.52	103 ± 26.67	S1–D4–I3
	SR300	1350 ± 8	3.18 ± 0.73	122.5 ± 38.89	S2–D3–I2
	SR400	1292 ± 14	2.0 ± 0.5	60.33 ± 5.77	S3–D5–I5
	SR550	1221 ± 16	6.0 ± 1.5	77 ± 4.24	S4–D2–I4
	SA1100	861 ± 3	27.25 ± 0.96	450	S5–D1–I1
<i>SDSS 2507 +3 wt.% Ni</i>					
	AS	1324.31 ± 25.44	6.78 ± 1.92	168.71 ± 25.30	S4–D2–I4
	SR400	1413.74 ± 14.06	6.58 ± 0.28	142.00 ± 76.37	S2–D3–I5
	SR450	1414.09 ± 11.14	6.55 ± 0.61	98.67 ± 17.62	S1–D6–I6
	SR500	1350.03 ± 46.65	6.58 ± 2.42	178.33 ± 25.17	S3–D3–I3
	SR550	1223.70 ± 69.04	6.58 ± 5.02	202.67 ± 21.13	S5–D3–I2
	SA1100	890.04 ± 1.97	29.59 ± 0.81	> 450	S6–D1–I1
<i>SDSS 2507 +6 wt.% Ni</i>					
	AS	976.57 ± 11.48	16.97 ± 1.23	388.67 ± 83.79	S4–D2–I3
	SR350	1001.51 ± 5.40	16.28 ± 2.03	403.33 ± 79.10	S3–D3–I2
	SR425	1046.5 ± 15.2	15.20 ± 2.4	274.33 ± 23.09	S1–D4–I5
	SR525	1033.51 ± 7.75	14.91 ± 1.81	341.67 ± 54.72	S2–D5–I4
	SA1150	794.93 ± 3.59	30.99 ± 0.81	> 450	S5–D1–I1

Note: S, D, and I denote within-alloy ranks for strength, ductility, and impact energy, respectively; rank 1 is the highest mean value. Ties are assigned where rounded means are identical. Rankings are descriptive only and do not imply statistical separation. Values reported as > 450 J are machine-limit impact responses.

4.6.3 Fracture-surface morphology

Representative low and high magnification fracture surfaces are compared in Fig. 4.35 for LPBF-built SDSS 2507 in as-built (AS), stress-relieved at 300–550 °C (SR300–SR550) conditions.

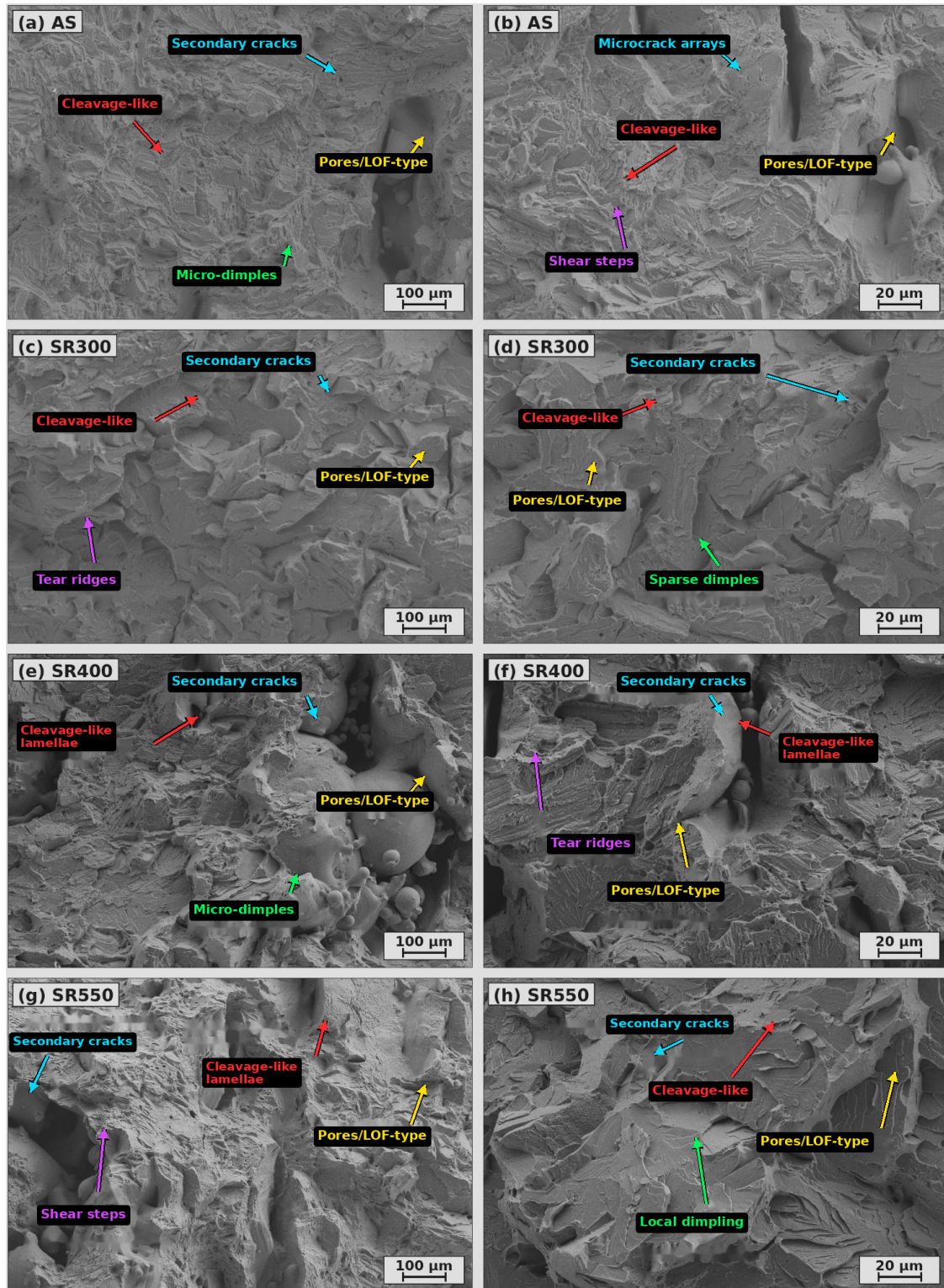


Figure 4.35: Representative SEM fracture-surface morphologies of LPBF-fabricated SDSS 2507. Low-magnification images are shown in the left column and high-magnification images in the right column: (a,b) AS, (c,d) SR300, (e,f) SR400, and (g,h) SR550.

The labeled features include pores or irregular cavities, cleavage-like facets or lamellae, secondary microcracks, tear ridges or shear steps, and local dimples. Mixed fracture-surface morphologies are observed in all four conditions. The AS condition contains local dimples, cleavage-like facets, secondary microcracks, and pores or irregular cavities. The SR300, SR400, and SR550 conditions contain cleavage-like regions, secondary microcracks, tear ridges or shear steps, sparse dimples, and local defect-like features.

4.7 Corrosion behavior

This section presents the laboratory electrochemical response of laser powder bed fusion (LPBF)-fabricated super duplex stainless steel (SDSS) 2507 alloy systems in the as-built and post-processed states. The standard alloy covers the as-built (AS), stress-relieved at 300 °C (SR300), stress-relieved at 400 °C (SR400), stress-relieved at 550 °C (SR550), and solution-annealed at 1100 °C (SA1100) conditions. The +3 wt.% Ni alloy covers AS, stress-relieved at 400, 450, 500, and 550 °C for 1 h (SR400, SR450, SR500, and SR550), and solution-annealed at 1100 °C for 15 min (SA1100). The +6 wt.% Ni alloy covers AS, stress-relieved at 350 °C for 2 h (SR350), stress-relieved at 425 °C for 1 h (SR425), stress-relieved at 525 °C for 1 h (SR525), and solution-annealed at 1150 °C for 10 min (SA1150).

The results are organised by open-circuit potential (OCP), cyclic potentiodynamic polarization (CPP), and electrochemical impedance spectroscopy (EIS). The OCP results report potential evolution during immersion. The CPP results report corrosion current density, polarization resistance, breakdown potential, repassivation potential, and reverse-scan response. The EIS results report Nyquist response, Bode magnitude, Bode phase behavior, and fitted interfacial parameters.

The EIS spectra were additionally fitted using an equivalent circuit model (ECM) to extract solution resistance, constant phase element parameters, and charge-transfer resistance. Fit quality was evaluated using the fitting objective, χ^2 , together with the maximum absolute residuals in the real and imaginary impedance components, $\max |\Delta Z'|$ and $\max |\Delta Z''|$. These metrics are reported to document the agreement between the measured spectra and the fitted ECM response. The section therefore compares electrochemical descriptors across alloy chemistry and heat-treatment condition without assigning a single universal corrosion ranking.

4.7.1 Open-circuit potential response

Figure 4.36 shows the open-circuit potential (OCP) response of the standard alloy during 60 min of immersion. The OCP profiles show clear condition-dependent differences in potential level and time-dependent evolution. The as-built condition (AS) shifts gradually toward less negative values during immersion, indicating a continuous change in the measured surface potential over the test period. SR300 remains relatively stable near -100 mV, while SR400 shows the least negative final response after an initial potential drop. SR550 remains the most negative condition throughout the test, and SA1100 stays within an intermediate potential range. Overall, the standard alloy follows the OCP ordering $SR400 > SR300/SA1100 > AS > SR550$, based on the relative potential levels observed during immersion.

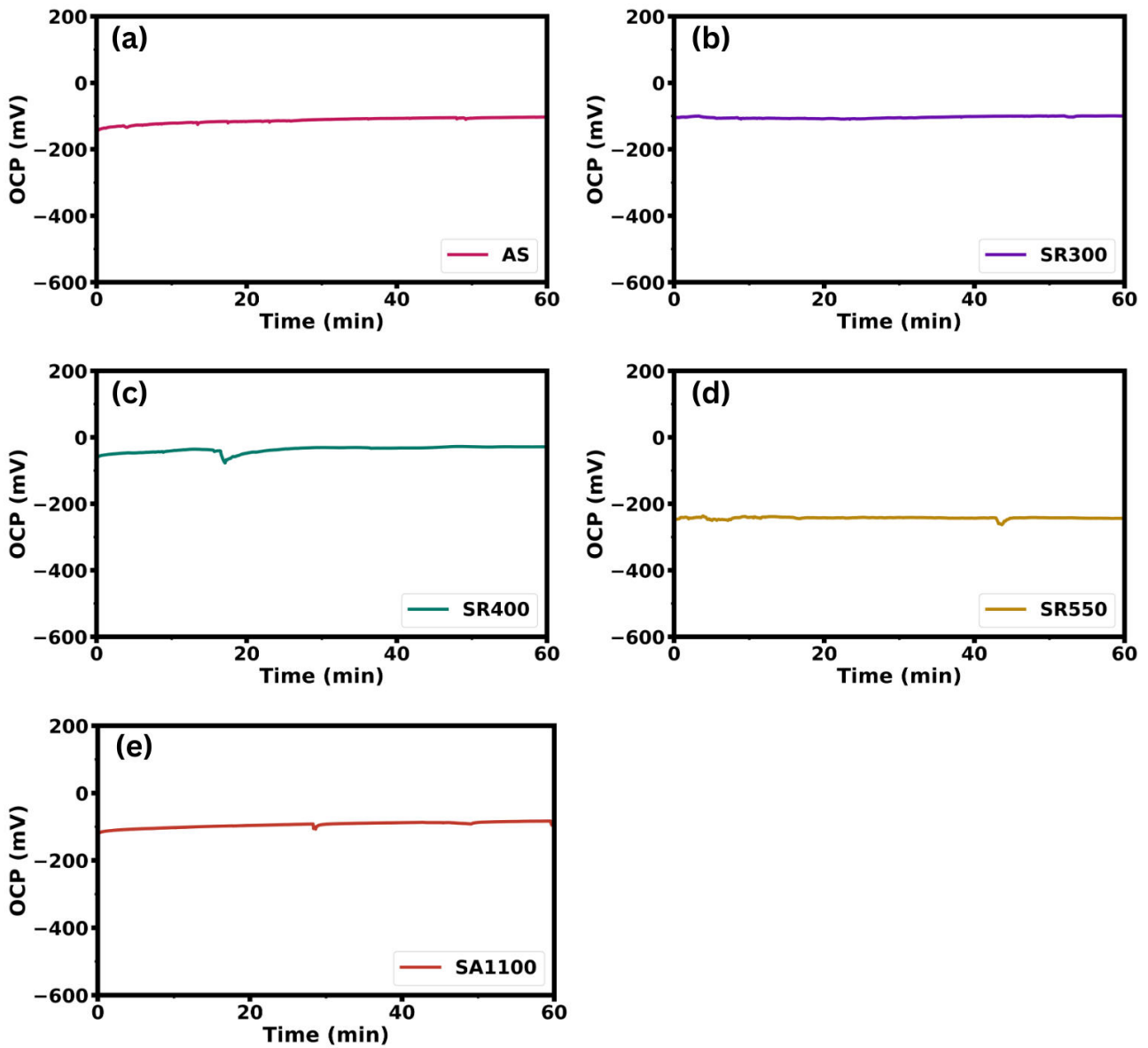


Figure 4.36: Open circuit potential(OCP) profiles of LPBF-fabricated standard SDSS 2507: (a) AS, (b) SR300, (c) SR400, (d) SR550, and (e) SA1100.

Figure 4.37 shows the open-circuit potential (OCP) response of the +3 wt.% Ni alloy during 60 min of immersion. The OCP profiles show clear differences among the as-built, stress-relieved, and solution-annealed conditions. The as-built condition (AS) remains the most negative throughout the immersion period. SR400 and SR550 show intermediate potential levels, whereas SR450 and SR500 show the least negative responses among the +3 wt.% Ni conditions. SA1100 remains close to -200 mV and lies between the most negative AS response and the less negative SR450/SR500 responses. The trace separation is maintained over most of the immersion period, indicating a condition-dependent mixed-potential response. These OCP trends are used as comparative indicators only and are not interpreted alone as proof of improved corrosion resistance. Overall, the +3 wt.% Ni alloy follows the OCP ordering $SR450/SR500 > SR400/SR550/SA1100 > AS$, based on the relative potential levels observed during immersion.

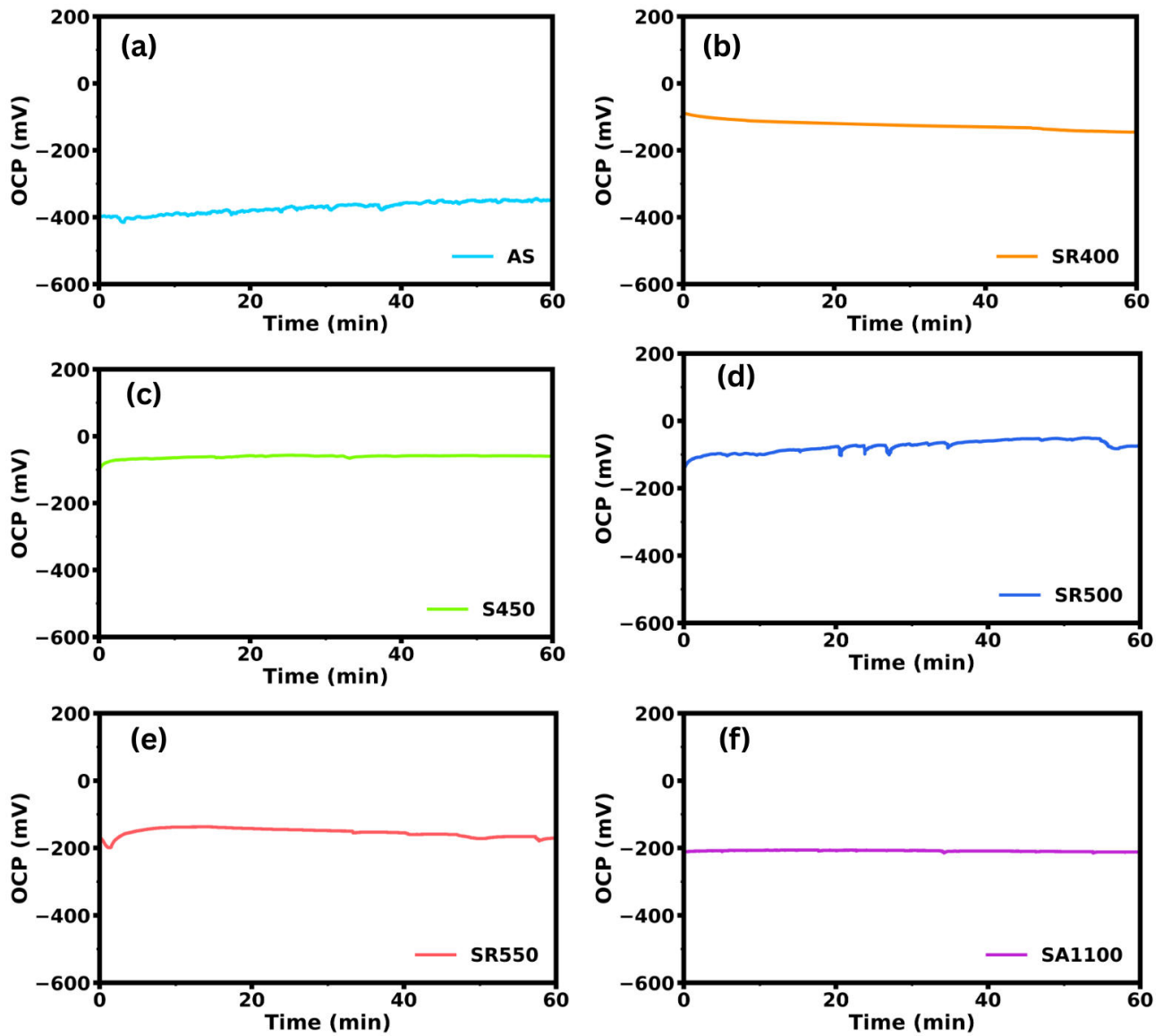


Figure 4.37: Open circuit potential(OCP) profiles of LPBF-fabricated SDSS 2507 +3 wt.% Ni: (a) AS, (b) SR400, (c) SR450, (d) SR500, (e) SR550, and (f) SA1100.

Figure 4.38 shows the open-circuit potential (OCP) response of the +6 wt.% Ni alloy during 60 min of immersion. The OCP profiles show clear condition-dependent differences across the as-built, stress-relieved, and solution-annealed states. AS, SR425, and SR525 remain near or above 0 mV for most of the immersion period, with SR525 showing the most positive response. SR350 remains slightly negative, whereas SA1150 remains negative throughout immersion and gives the lowest potential range in this alloy series. The trace separation is maintained over most of the exposure period, indicating a condition-dependent mixed-potential response. These OCP trends are used as comparative indicators only and are not interpreted alone as proof of improved corrosion resistance. Overall, the +6 wt.% Ni alloy follows the OCP ordering $SR525 > AS/SR425 > SR350 > SA1150$, based on the relative potential levels observed during immersion.

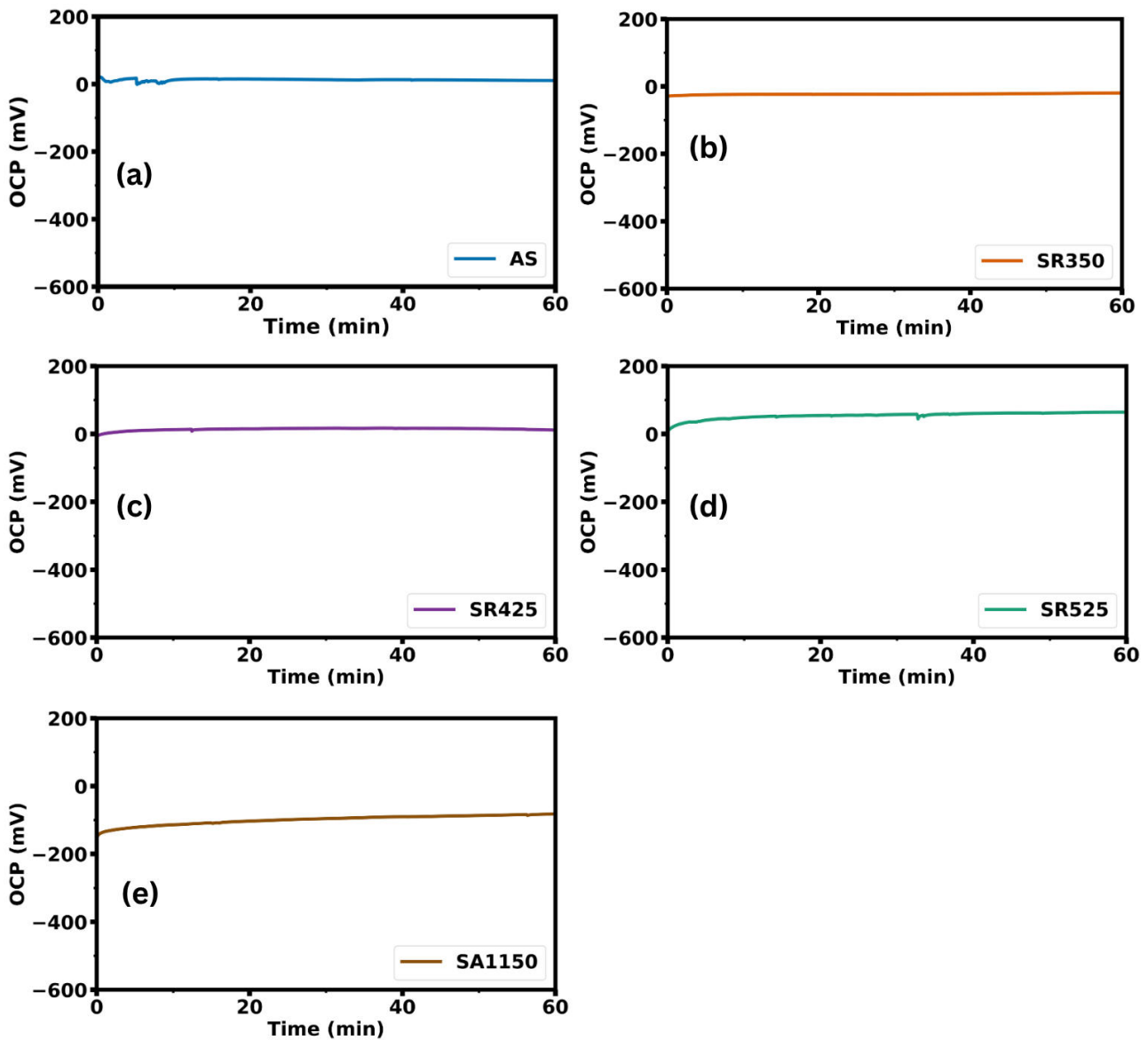


Figure 4.38: Open circuit potential(OCP) profiles of LPBF-fabricated SDSS 2507 +6 wt.% Ni: (a) AS, (b) SR350, (c) SR425, (d) SR525, and (e) SA1150.

4.7.2 Potentiodynamic polarization behavior

Figure 4.39 shows the cyclic potentiodynamic polarization (CPP) response of the standard alloy. The CPP curves show condition-dependent differences in forward-scan current density, breakdown region, reverse-scan trajectory, and hysteresis behaviour. AS, SR300, SR400, and SA1100 cover a broad potential range and show clear separation between the forward and backward scans. SR550 shows a narrower potential range and a more limited polarization response compared with the other standard-alloy conditions. The forward branches also differ in current-density level, with SR400 remaining lower over a wider potential interval than the other standard-alloy states. The reverse branches show different return paths after the upper potential limit, producing condition-dependent hysteresis widths. The plotted curves therefore distinguish the standard-alloy conditions by potential range, current-density response, and forward/reverse scan separation.

Among the standard-alloy states, SR400 shows the most favourable combined CPP profile, with a low current-density response and an extended potential range, whereas SR550 gives the weakest CPP response in this set.

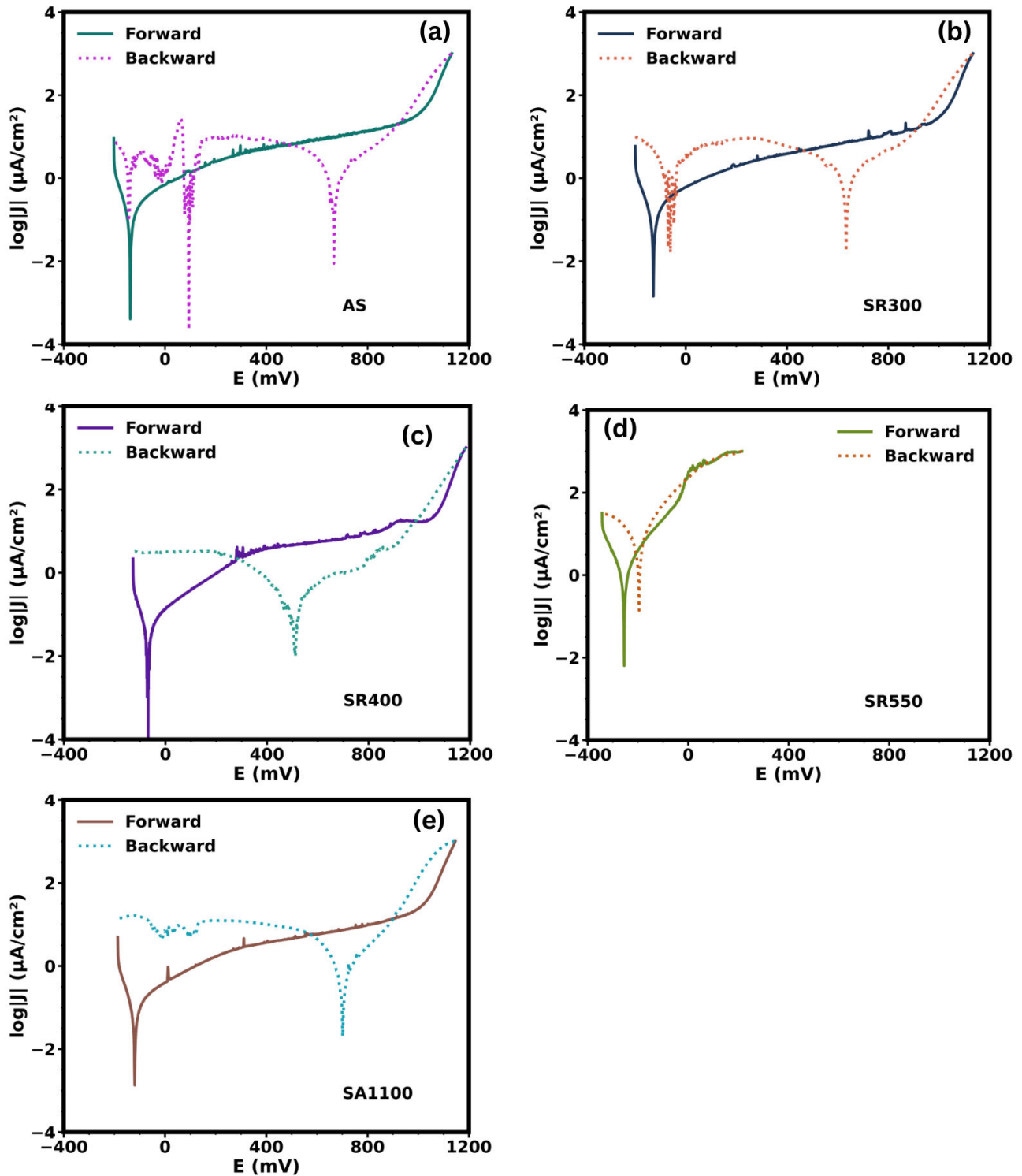


Figure 4.39: Cyclic potentiodynamic polarization (CPP) curves of LPBF-fabricated standard SDSS 2507 in the (a) AS, (b) SR300, (c) SR400, (d) SR550, and (e) SA1100 conditions. Solid lines indicate the forward anodic scan, and dotted lines indicate the reverse scan.

Table 4.8 summarizes the electrochemical parameters obtained from representative cyclic potentiodynamic polarization (CPP) traces of the standard alloy. The extracted descriptors show clear condition-dependent differences in corrosion current density, J_{corr} , polarization resistance, R_p , corrosion potential, E_{corr} , and the breakdown/repassivation potential response. SR400 gives the strongest parameter set among the standard-alloy conditions, with the lowest J_{corr} , the highest R_p , and the most positive E_{corr} . This condition also shows a favourable combination of low current-density response and extended polarization range in the corresponding CPP trace. In contrast, SR550 gives the weakest parameter set, with the highest J_{corr} , the lowest R_p , and the most negative E_{corr} . AS, SR300, and SA1100 show intermediate values between these two limiting responses. Among these intermediate states, SA1100 gives a higher resistance-related response than AS and SR300, while AS and SR300 remain closer to each other in the extracted CPP descriptors. Overall, the tabulated parameters are consistent with the visual CPP trends in Figure 4.39, where SR400 shows the strongest standard-alloy polarization response and SR550 shows the weakest response within this alloy set.

Table 4.8: Summary of electrochemical parameters extracted from representative cyclic potentiodynamic polarization (CPP) traces of LPBF-fabricated standard SDSS 2507 in the AS, SR300, SR400, SR550, and SA1100 conditions.

Condition	E_{ocp} (mV)	E_{rp} (mV)	E_{br} (mV)	J_{corr} ($\mu\text{A}/\text{cm}^2$)	β_a (mV)	β_c (mV)	E_{corr} (mV)	R_p ($\text{k}\Omega\cdot\text{cm}^2$)
AS	-103	999	1084	0.14	195.73	51.98	-147.17	126.82
SR300	-99.7	996	1083	0.16	255.88	59.11	-142.86	125.14
SR400	-28.6	1055	1127	0.06	225.50	50.05	-83.81	301.10
SR550	-244	-96.3	130	1.16	117.99	78.67	-261.03	17.68
SA1100	-88.2	1011	1102	0.13	295.36	53.86	-136.35	149.35

Figure 4.40 shows the cyclic potentiodynamic polarization (CPP) response of the +3 wt.% Ni alloy. The CPP curves show condition-dependent differences in forward-scan current density, potential range, reverse-scan trajectory, and hysteresis behaviour. All conditions extend to high potentials, indicating a broad polarization range across the +3 wt.% Ni series. AS and SR400 show comparatively smoother forward branches, whereas SR450, SR500, SR550, and SA1100 show visible forward-scan fluctuations at higher potentials. Separation between the forward and backward scans is observed in all conditions, confirming that the reverse-scan response differs from the initial anodic scan. The stress-relieved and solution-annealed states show distinct reverse-scan paths after the upper potential limit, with differences in the magnitude and shape of the hysteresis region. Among the investigated +3 wt.% Ni conditions, SR500 and SR550 show the strongest combined CPP response in terms of low current-density behaviour and extended polarization range, while AS gives the least favourable CPP profile within this alloy series.

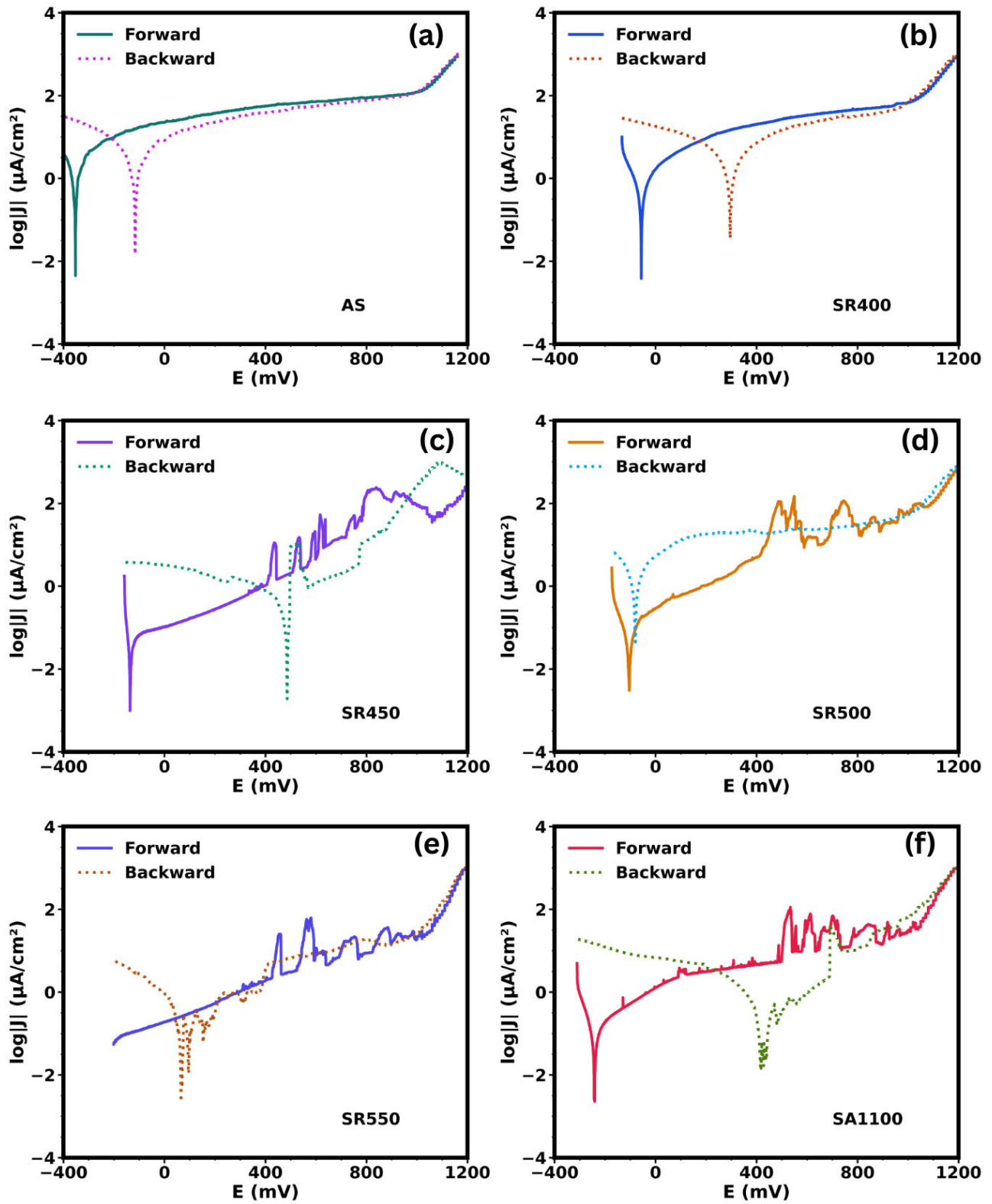


Figure 4.40: Cyclic potentiodynamic polarization(CPP)curves of LPBF-fabricated SDSS 2507 +3 wt.% Ni in the (a) AS, (b) SR400, (c) SR450, (d) SR500, (e) SR550, and (f) SA1100 conditions. Solid lines represent the forward scan and dotted lines represent the backward scan.

Table 4.9 summarizes the electrochemical parameters obtained from representative cyclic potentiodynamic polarization (CPP) traces of the +3 wt.% Ni alloy. The extracted descriptors show condition-dependent differences in corrosion current density, J_{corr} , polarization resistance, R_p , corrosion potential, E_{corr} , breakdown potential, E_{br} , repassivation potential, E_{rp} , and the potential separation $E_{\text{rp}} - E_{\text{corr}}$. AS gives the highest J_{corr} and the lowest R_p within this alloy series. SR500 and SR550 give the lowest J_{corr} values and the highest R_p values among the stress-relieved conditions, showing the strongest resistance-related CPP descriptors in this group. SR400 and SR450 show intermediate parameter sets, with J_{corr} and R_p values positioned between AS and the SR500/SR550 responses. SA1100 also remains within an intermediate range. The breakdown and repassivation descriptors further distinguish the +3 wt.% Ni conditions by the potential range over which the forward and reverse scans remain separated. Overall, the tabulated CPP parameters separate the +3 wt.% Ni conditions by current-density level, resistance-related response, breakdown potential, repassivation potential, and reverse-scan behaviour.

Table 4.9: Electrochemical parameters extracted from representative cyclic potentiodynamic polarization (CPP) traces of LPBF-fabricated SDSS 2507 +3 wt.% Ni in the AS, SR400, SR450, SR500, SR550, and SA1100 conditions.

Condition	E_{ocp} (mV)	E_{br} (mV)	E_{rp} (mV)	J_{corr} ($\mu\text{A}/\text{cm}^2$)	β_a (mV)	β_c (mV)	E_{corr} (mV)	R_p ($\text{k}\Omega \cdot \text{cm}^2$)
AS	-350	1095	1000	2.37	289.22	83.72	-378.62	11.88
SR400	-200	1127	992	0.83	232.75	72.64	-78.36	28.80
SR450	-60	1153	960	0.36	263.70	39.89	-126.10	41.99
SR500	-70	1152	962	0.10	259.17	55.18	-127.00	201.80
SR550	-170	1141	841	0.06	508.07	34.57	-237.79	216.84
SA1100	-210	1123	874	0.11	213.73	54.63	-250.68	165.45

Figure 4.41 shows the cyclic potentiodynamic polarization (CPP) response of the +6 wt.% Ni alloy. The CPP curves show condition-dependent differences in forward-scan current density, potential range, reverse-scan trajectory, and hysteresis behaviour. AS, SR350, SR425, and SR525 cover broad polarization ranges with clear forward/backward scan separation. AS, SR350, and SR425 show pronounced forward-branch fluctuations and distinct backward-scan loops, whereas SR525 shows more localized fluctuations within the plotted potential range. SA1150 covers a narrower potential interval and shows closer agreement between the forward and backward scans than the as-built and stress-relieved conditions. The plotted curves separate the +6 wt.% Ni conditions by current-density level, fluctuation intensity, reverse-scan path, scan separation, and hysteresis-loop shape. These differences show that the +6 wt.% Ni response varies with thermal pathway rather than Ni enrichment alone. The CPP curves are therefore used as comparative laboratory descriptors of breakdown and repassivation behaviour, not as proof of service-level localized-corrosion resistance.

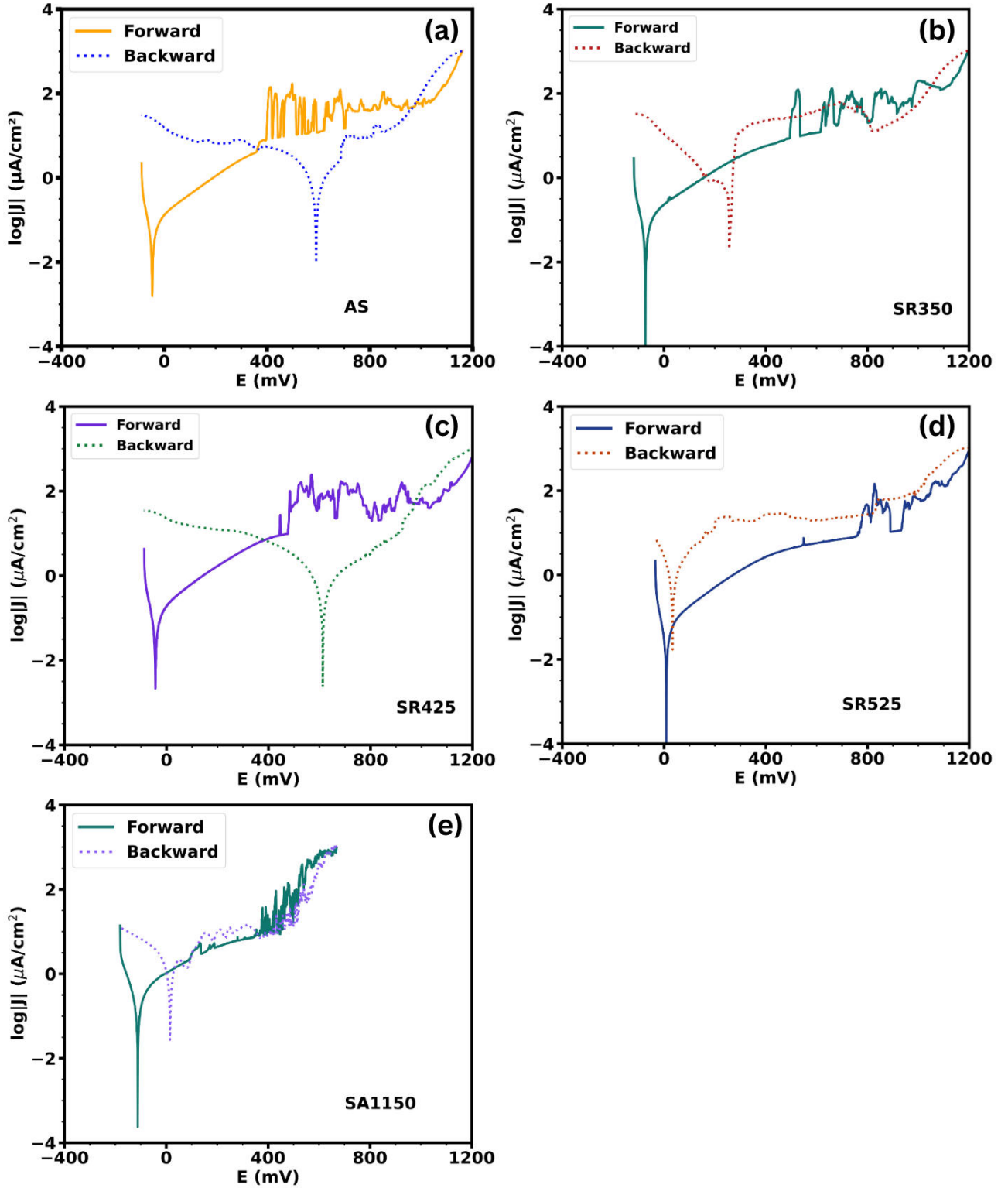


Figure 4.41: Cyclic potentiodynamic polarization(CPP)curves of LPBF-fabricated SDSS 2507 +6 wt.% Ni in the (a) AS, (b) SR350, (c) SR425, (d) SR525, and (e) SA1150 conditions. Solid lines represent the forward scan and dotted lines represent the backward scan.

Table 4.10 summarizes the electrochemical parameters obtained from representative cyclic potentiodynamic polarization (CPP) traces of the +6 wt.% Ni alloy. The extracted descriptors show condition-dependent differences in corrosion current density, J_{corr} , polarization resistance, R_p ,

open-circuit potential, E_{ocp} , corrosion potential, E_{corr} , breakdown potential, E_{br} , repassivation potential, E_{rp} , and the potential separation $E_{rp} - E_{corr}$. AS and the stress-relieved states show low J_{corr} values relative to SA1150, indicating a clear separation between the non-solution-annealed conditions and the solution-annealed condition in the current-density descriptors. Among the stress-relieved states, SR525 gives the highest resistance-related response, whereas SR350 and SR425 remain within the lower-current-density group but differ in potential-related descriptors. Positive E_{ocp} values are recorded for SR425 and SR525, while AS, SR350, and SA1150 remain negative. SA1150 gives the highest J_{corr} , the lowest R_p , and a negative E_{ocp} , representing the weakest tabulated CPP parameter set in the +6 wt.% Ni series. The breakdown and repassivation descriptors further separate the conditions by the extent of the anodic potential range and the reverse-scan response. Overall, the tabulated CPP parameters distinguish the +6 wt.% Ni conditions by current-density level, resistance-related response, OCP position, corrosion-potential position, breakdown response, repassivation response, and reverse-scan descriptor values.

Table 4.10: Electrochemical parameters extracted from representative cyclic potentiodynamic polarization (CPP) traces of LPBF-fabricated SDSS 2507 +6 wt.% Ni in the AS, SR350, SR425, SR525, and SA1150 conditions.

Condition	E_{ocp} (mV)	E_{br} (mV)	E_{rp} (mV)	J_{corr} ($\mu A/cm^2$)	β_a (mV)	β_c (mV)	E_{corr} (mV)	R_p ($k\Omega \cdot cm^2$)
AS	-8	1100	1050	0.07	219.12	32.42	-59.83	166.68
SR350	-18	1150	1080	0.09	235.66	34.27	-88.82	141.00
SR425	12	1040	990	0.13	239.81	33.17	-59.67	95.62
SR525	67	1050	850	0.06	233.22	31.96	-7.89	192.00
SA1150	-82	430	390	0.32	259.07	57.97	-127.53	65.00

4.7.3 Electrochemical impedance behavior

Electrochemical impedance spectroscopy (EIS) results compare the standard super duplex stainless steel (SDSS) 2507, SDSS 2507 +3 weight percent nickel (+3 wt.% Ni), and SDSS 2507 +6 weight percent nickel (+6 wt.% Ni) alloy systems after fabrication and post-build heat treatment. The compared conditions were the as-built (AS), stress-relieved at 300 °C (SR300), stress-relieved at 400 °C (SR400), stress-relieved at 550 °C (SR550), and solution-annealed at 1100 °C (SA1100) conditions for the standard alloy; AS, SR400, stress-relieved at 450 °C for 1 h (SR450), stress-relieved at 500 °C for 1 h (SR500), SR550, and SA1100 for the +3 wt.% Ni alloy; and AS, stress-relieved at 350 °C for 2 h (SR350), stress-relieved at 425 °C for 1 h (SR425), stress-relieved at 525 °C for 1 h (SR525), and solution-annealed at 1150 °C for 10 min (SA1150) for the +6 wt.% Ni alloy. The corresponding Nyquist, Bode magnitude, and Bode phase plots are shown in Figures 4.42–4.46, Figures 4.47–4.52, and Figures 4.53–4.57, respectively.

In the standard alloy, SR400 shows the most extended Nyquist arc and the highest low-frequency impedance modulus, $|Z|$, whereas SR550 gives the smallest impedance response. AS, SR300, and SA1100 show intermediate responses, with SA1100 closer to SR400 than to SR550. The Bode

phase spectra show broad phase-angle maxima in all conditions, with SR400 maintaining the strongest phase response over the widest frequency interval. In the +3 wt.% Ni alloy, AS gives the smallest Nyquist arc and the lowest low-frequency $|Z|$. Stress relief increases the impedance response, most clearly in SR450 and SR550. SR500 also shows a large impedance response, although lower than those of SR450 and SR550. SA1100 remains intermediate, exceeding AS and SR400 but remaining below SR450, SR500, and SR550. The Bode phase spectra show broader and stronger phase-angle responses after stress relief and solution annealing than in AS. In the +6 wt.% Ni alloy, the plotted spectra show a different ordering. AS displays the most extended visual Nyquist response and the highest low-frequency $|Z|$, followed by SR525. SR350 and SR425 show smaller responses, while SA1150 gives the weakest overall impedance response. The Bode phase spectra show broad phase-angle maxima in AS, SR350, SR425, and SR525, while SA1150 gives the lowest impedance response among the +6 wt.% Ni conditions.

The EIS spectra were fitted using the equivalent circuit model shown in Figure 4.58, consisting of the solution resistance, R_s , in series with a parallel constant phase element (CPE) and charge-transfer resistance, R_{ct} . The fitted parameters are summarized in Tables 4.12–4.14. In the standard alloy, the fitted R_{ct} follows the order SR400 > SA1100 > SR300 > AS > SR550, with values of 157.30, 121.50, 87.10, 78.35, and 7.18 $\text{k}\Omega\cdot\text{cm}^2$, respectively. The fitted double-layer capacitance, C_{dl} , ranges from 55 to 184 $\mu\text{F}\cdot\text{cm}^{-2}$, and the CPE exponent, n , ranges from 0.705 to 0.958.

In the +3 wt.% Ni alloy, the fitted R_{ct} values confirm the highest impedance response for SR550 and SR450, with 1883.00 and 1547.00 $\text{k}\Omega\cdot\text{cm}^2$, respectively. SR500 and SA1100 give intermediate values of 214.20 and 98.63 $\text{k}\Omega\cdot\text{cm}^2$, while SR400 and AS give lower values of 24.82 and 8.39 $\text{k}\Omega\cdot\text{cm}^2$, respectively. The fitted C_{dl} values range from 114.00 to 219.00 $\mu\text{F}\cdot\text{cm}^{-2}$, and n remains within 0.78–0.82. The narrow n range indicates a broadly comparable constant-phase-element response, while the dominant condition-dependent separation is expressed mainly through R_{ct} .

In the +6 wt.% Ni alloy, the fitted R_{ct} ranks SR525 highest, with 1251.00 $\text{k}\Omega\cdot\text{cm}^2$, followed by AS and SR350 with 333.40 and 314.30 $\text{k}\Omega\cdot\text{cm}^2$, respectively. SR425 gives 125.20 $\text{k}\Omega\cdot\text{cm}^2$, while SA1150 gives the lowest value, 37.52 $\text{k}\Omega\cdot\text{cm}^2$. The fitted C_{dl} values range from 5.45 to 37.72 $\mu\text{F}\cdot\text{cm}^{-2}$, and n ranges from 0.73 to 0.87. The +6 wt.% Ni alloy therefore shows a difference between the visual Nyquist/ $|Z|$ ranking and the fitted R_{ct} ranking, with AS giving the largest plotted impedance response and SR525 giving the highest fitted charge-transfer resistance. Thus, visual impedance magnitude and fitted circuit parameters are treated as complementary descriptors rather than interchangeable rankings.

Overall, the EIS results identify SR400 as the highest-impedance condition within the standard alloy, SR450/SR550 within the +3 wt.% Ni alloy, and SR525 by fitted R_{ct} within the +6 wt.% Ni alloy. The main condition-dependent differences are expressed through Nyquist arc size, low-frequency $|Z|$, Bode phase response, and fitted R_{ct} . The fitted parameters support comparative impedance interpretation, not unique physical proof of a single interfacial mechanism. Therefore, the EIS results are interpreted together with the OCP and CPP trends when ranking the electrochemical response of each alloy–heat-treatment condition.

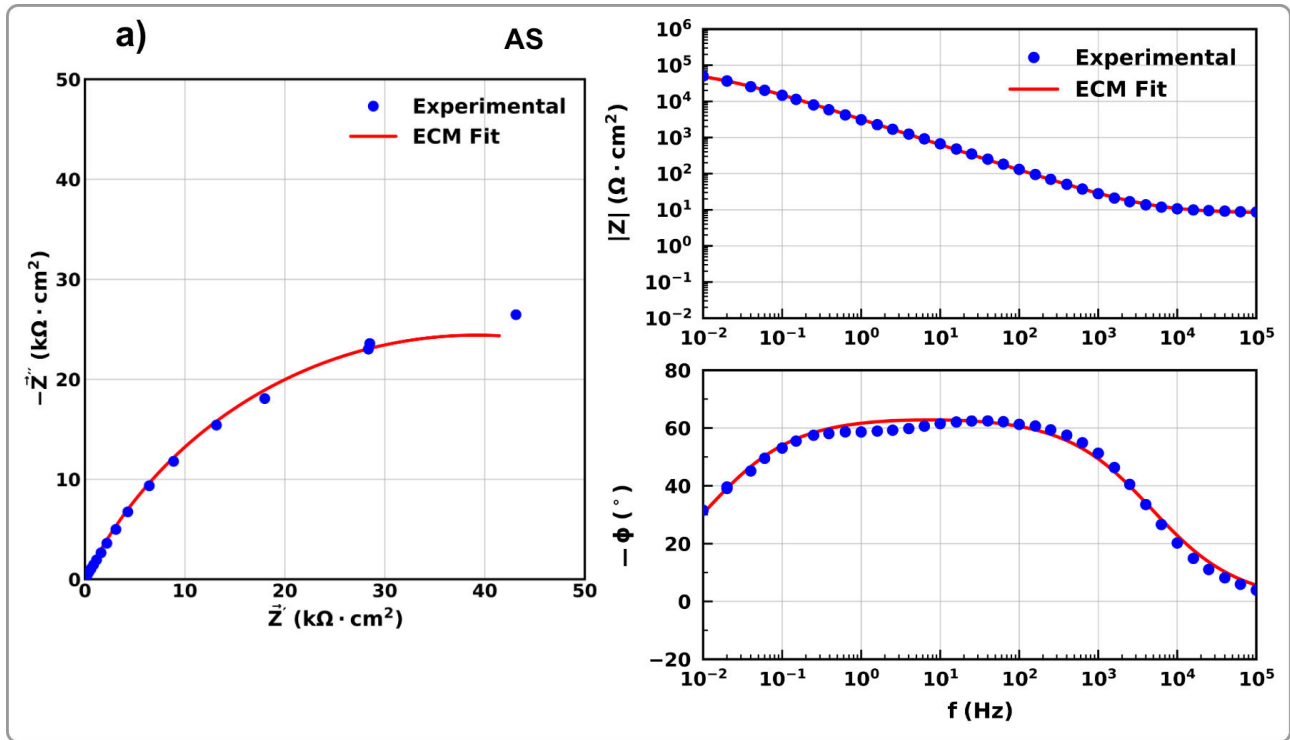


Figure 4.42: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated standard SDSS 2507 in AS, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the equivalent circuit model (ECM) fit.

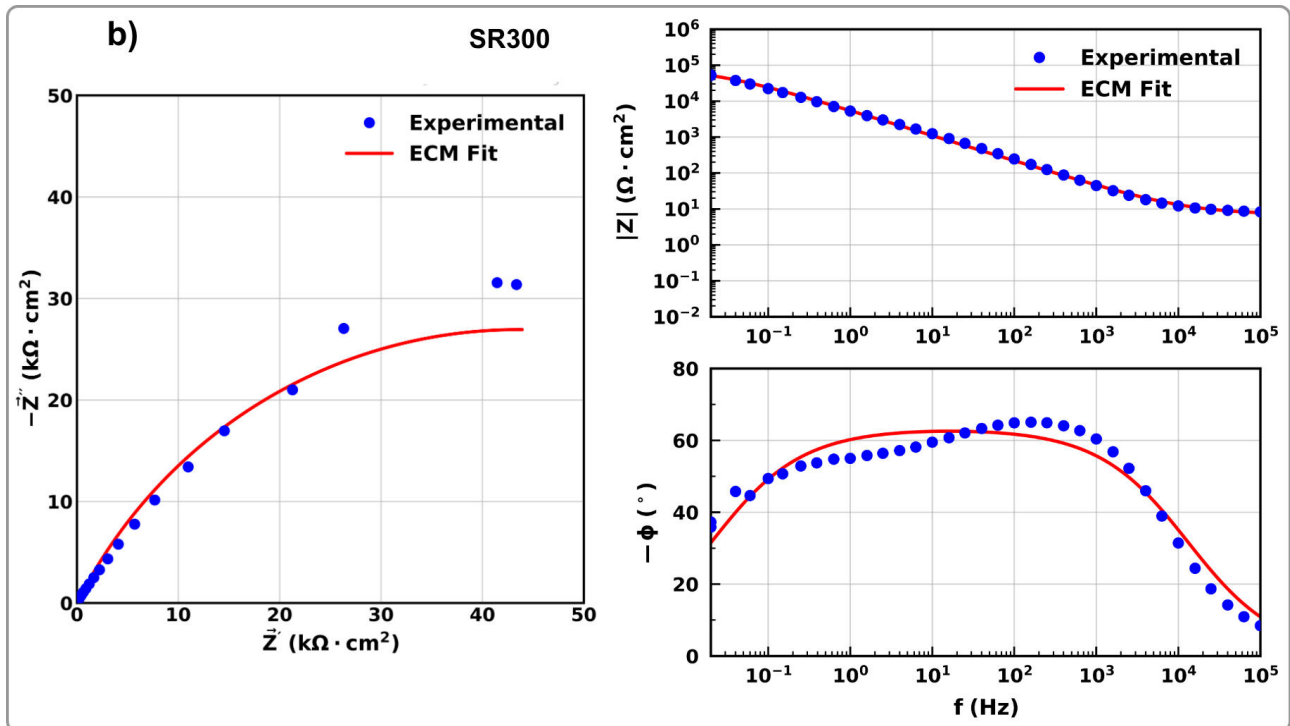


Figure 4.43: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated standard SDSS 2507 in SR300, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

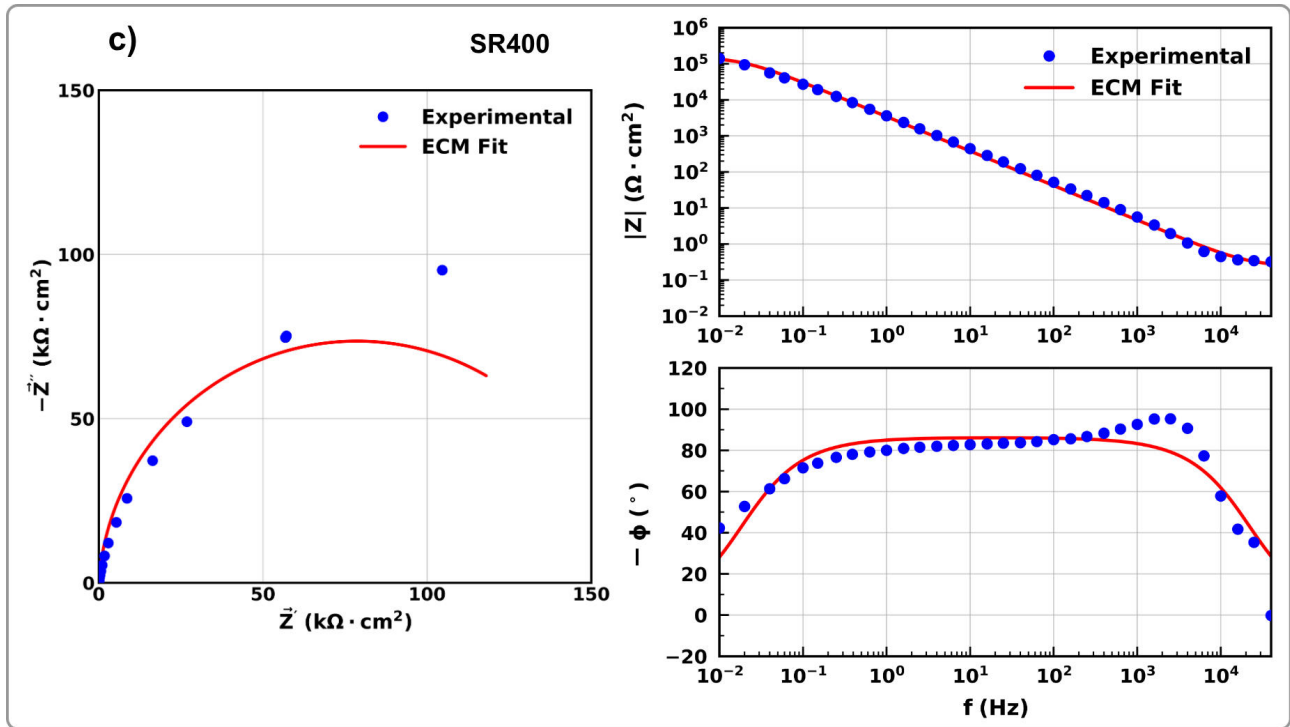


Figure 4.44: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated standard SDSS 2507 in SR400, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

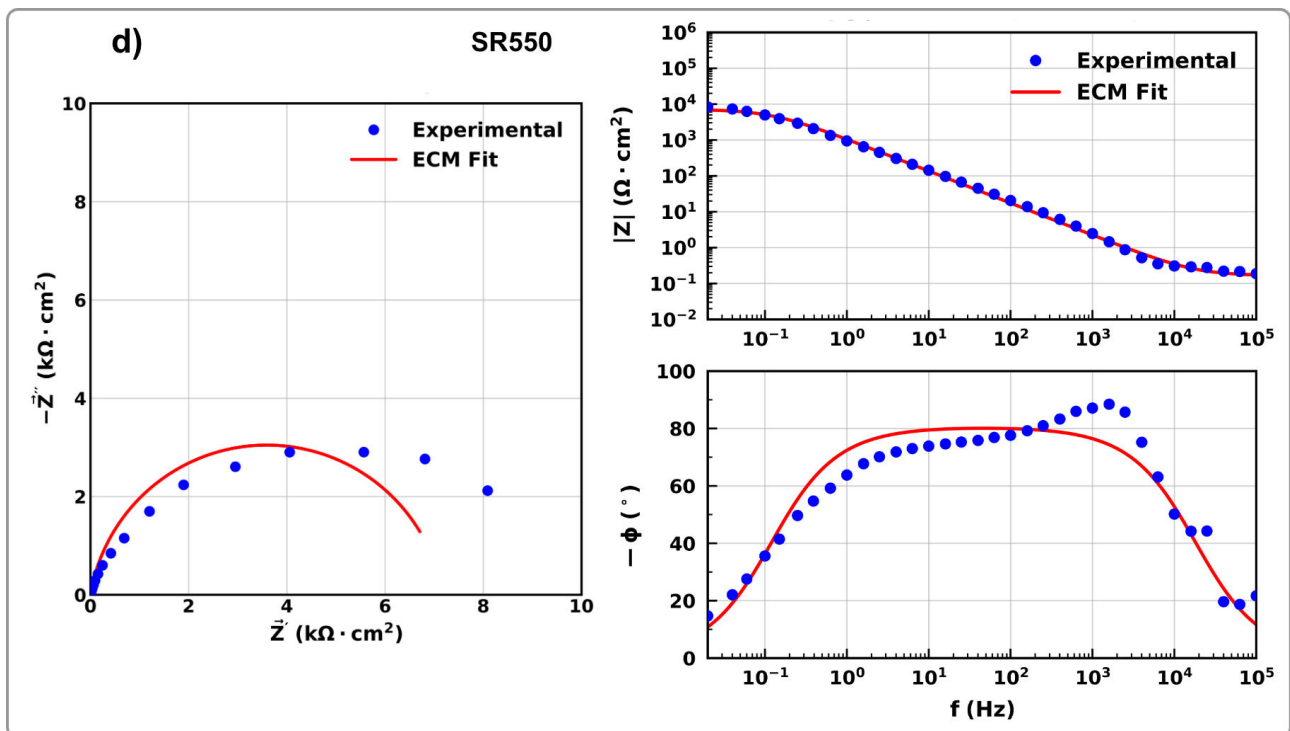


Figure 4.45: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated standard SDSS 2507 in SR550, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

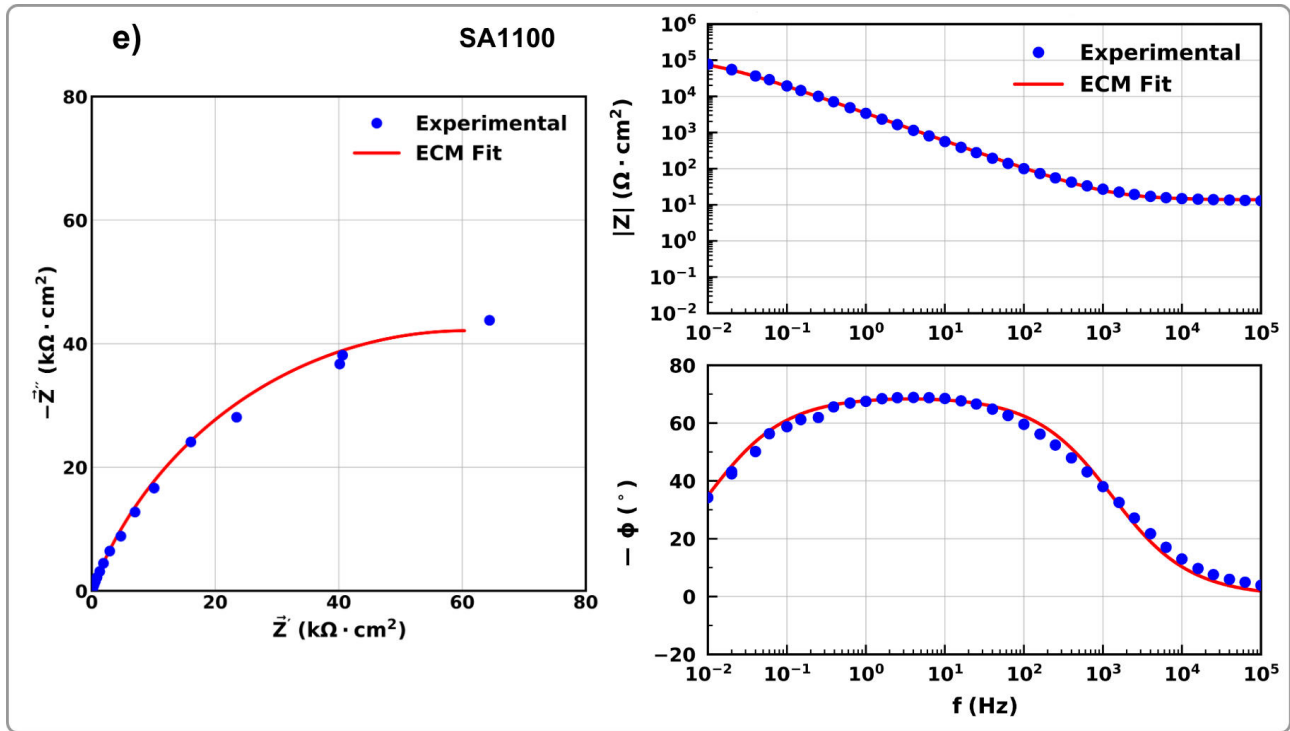


Figure 4.46: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated standard SDSS 2507 in SA1100, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

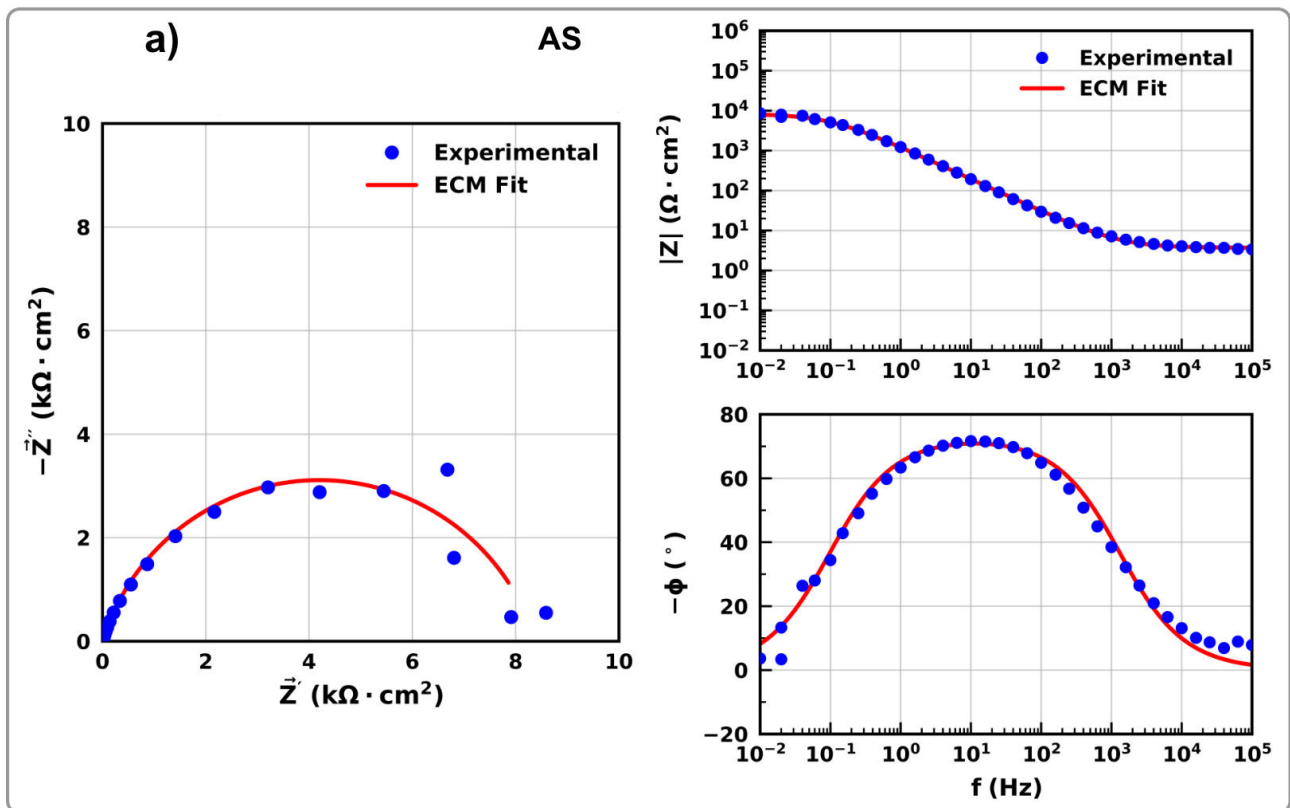


Figure 4.47: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated SDSS 2507 +3 wt.% Ni in AS, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

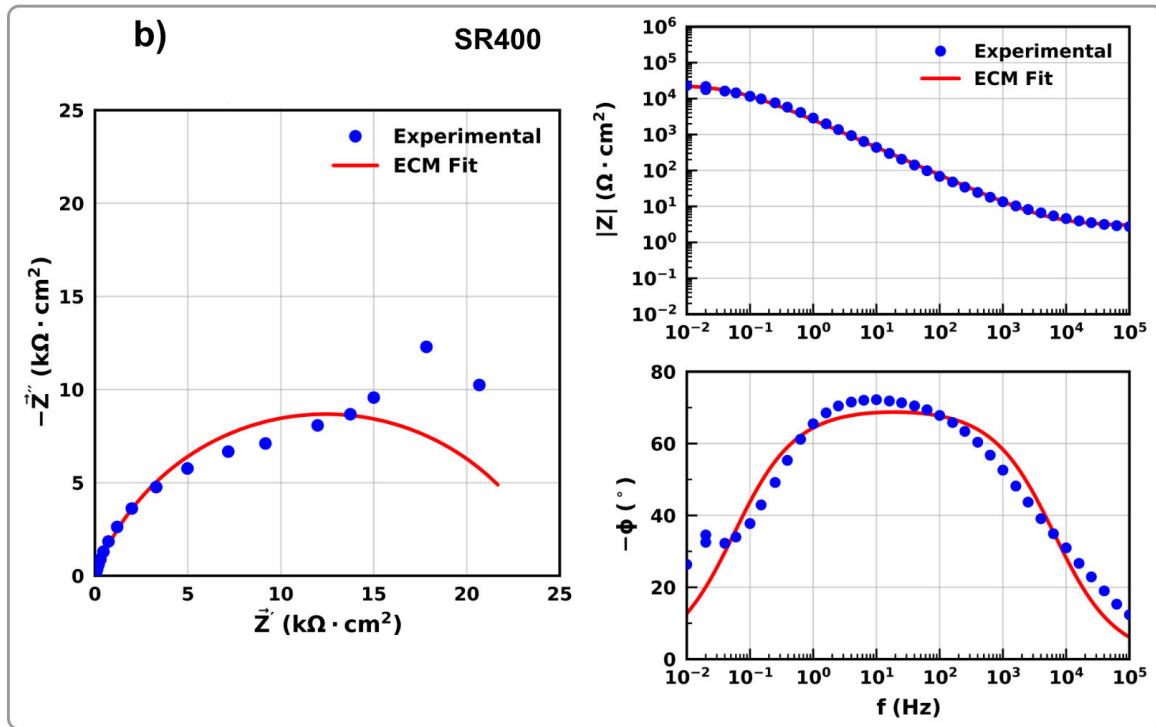


Figure 4.48: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated SDSS 2507 +3 wt.% Ni in SR400, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

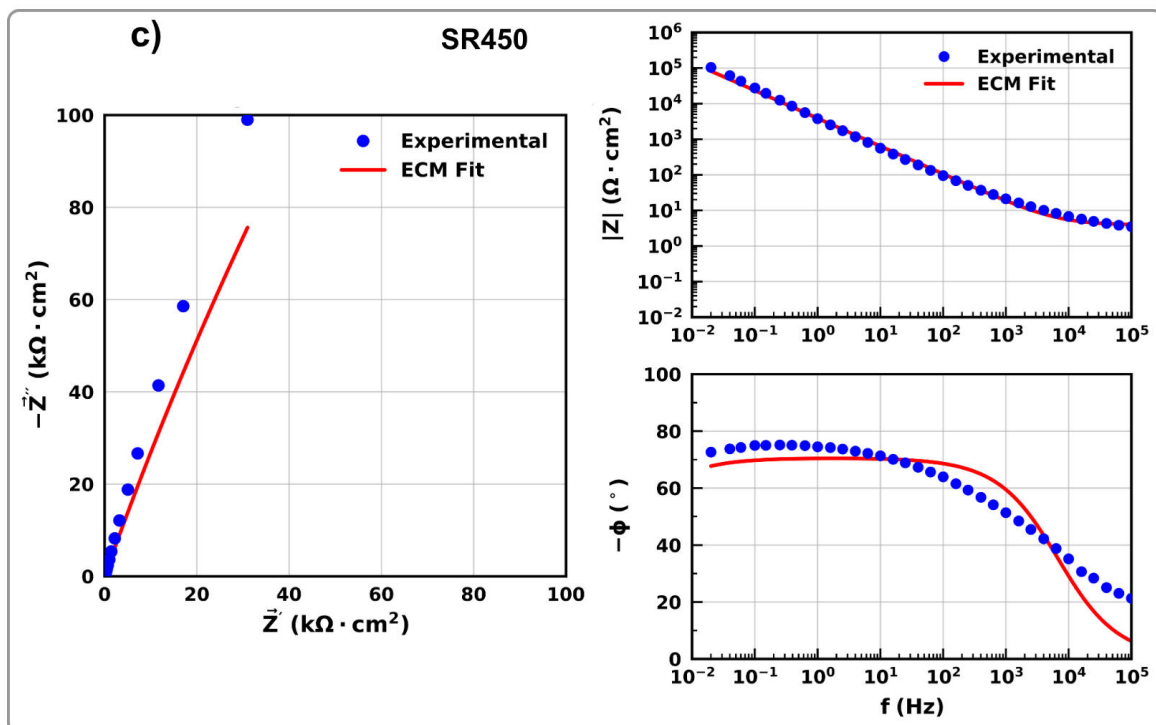


Figure 4.49: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated SDSS 2507 +3 wt.% Ni in SR450, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

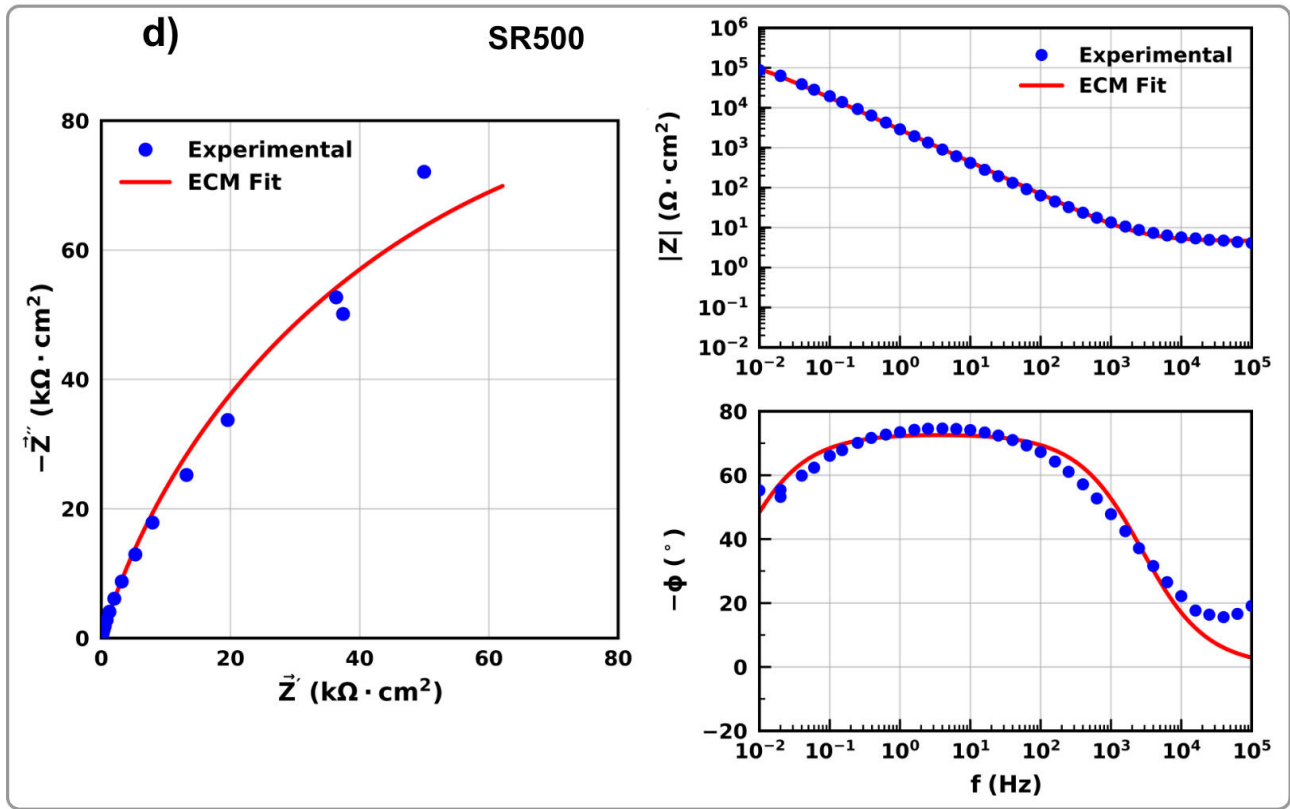


Figure 4.50: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated SDSS 2507 +3 wt.% Ni in SR500, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

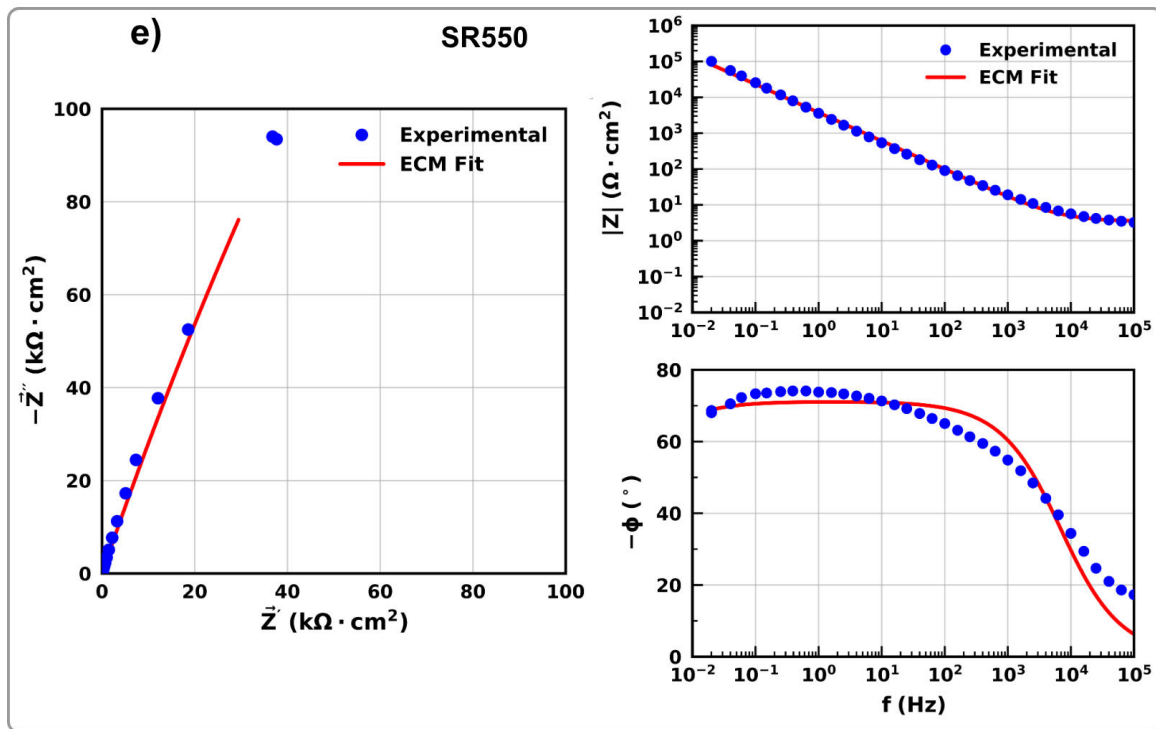


Figure 4.51: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated SDSS 2507 +3 wt.% Ni in SR550, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

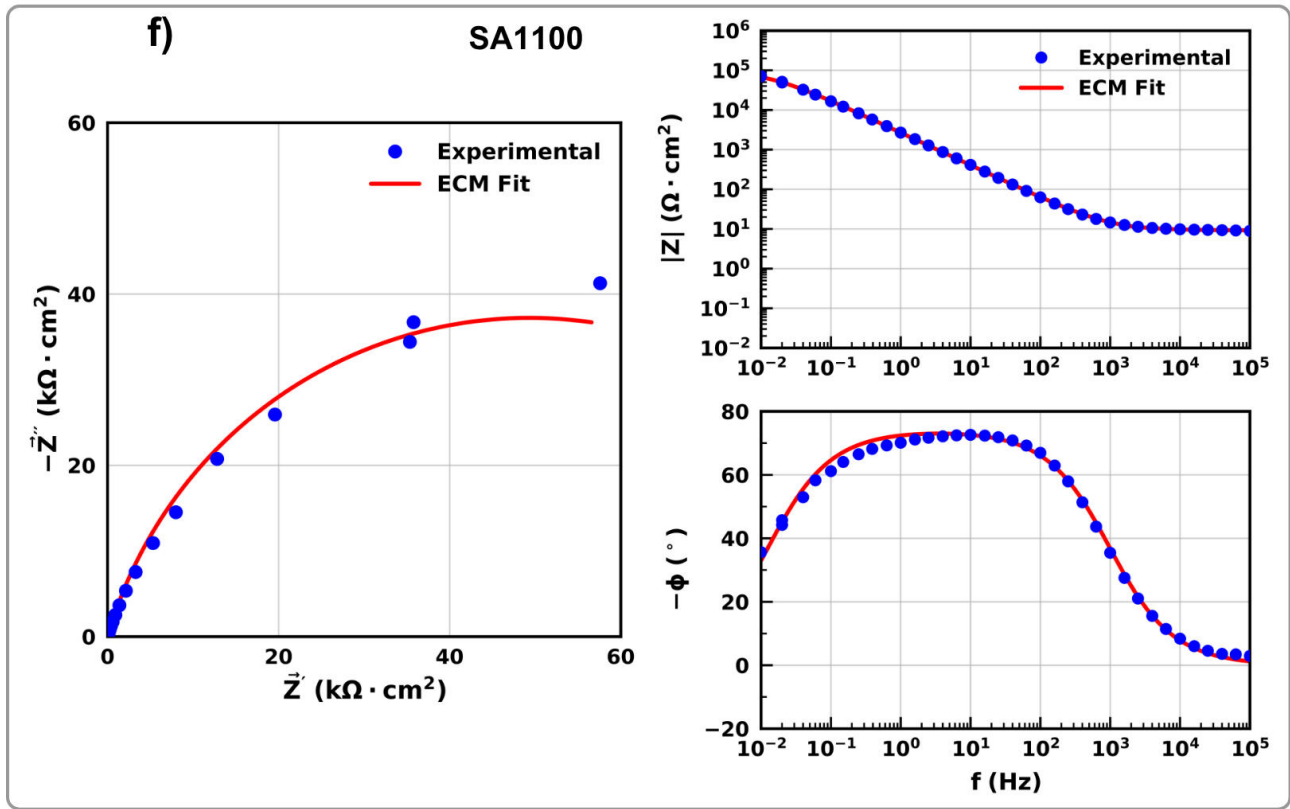


Figure 4.52: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated SDSS 2507 +3 wt.% Ni in SA1100, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

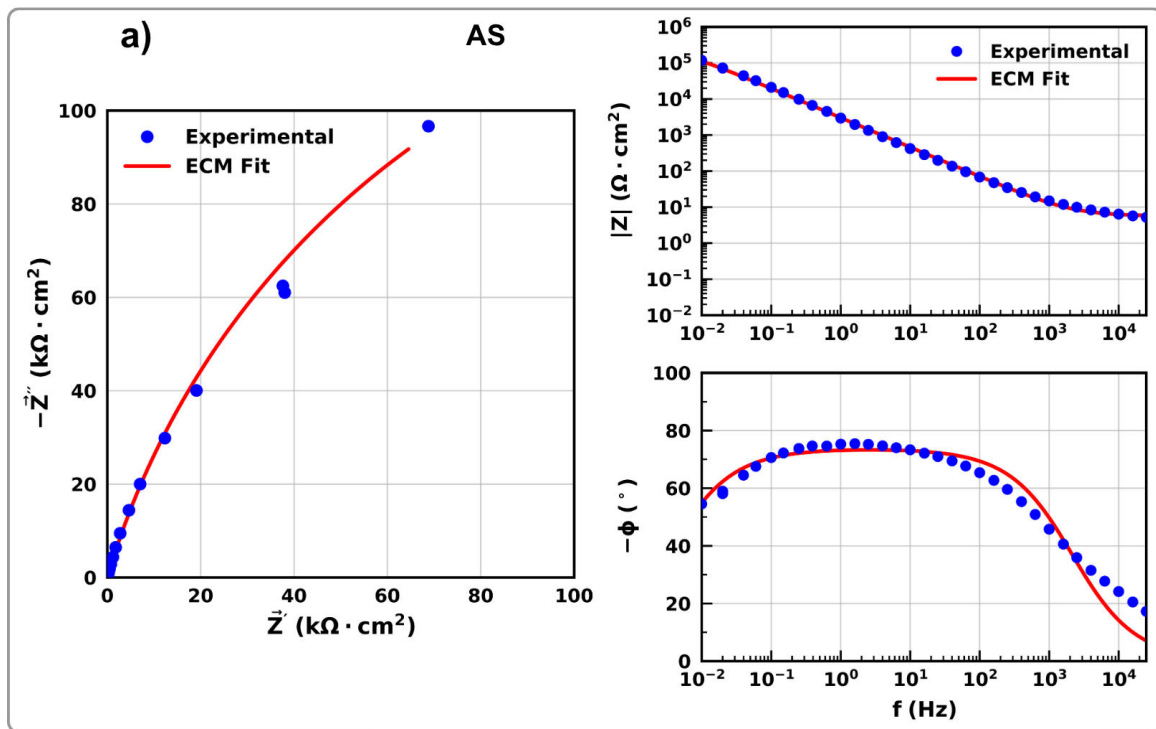


Figure 4.53: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated SDSS 2507 +6 wt.% Ni in AS, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

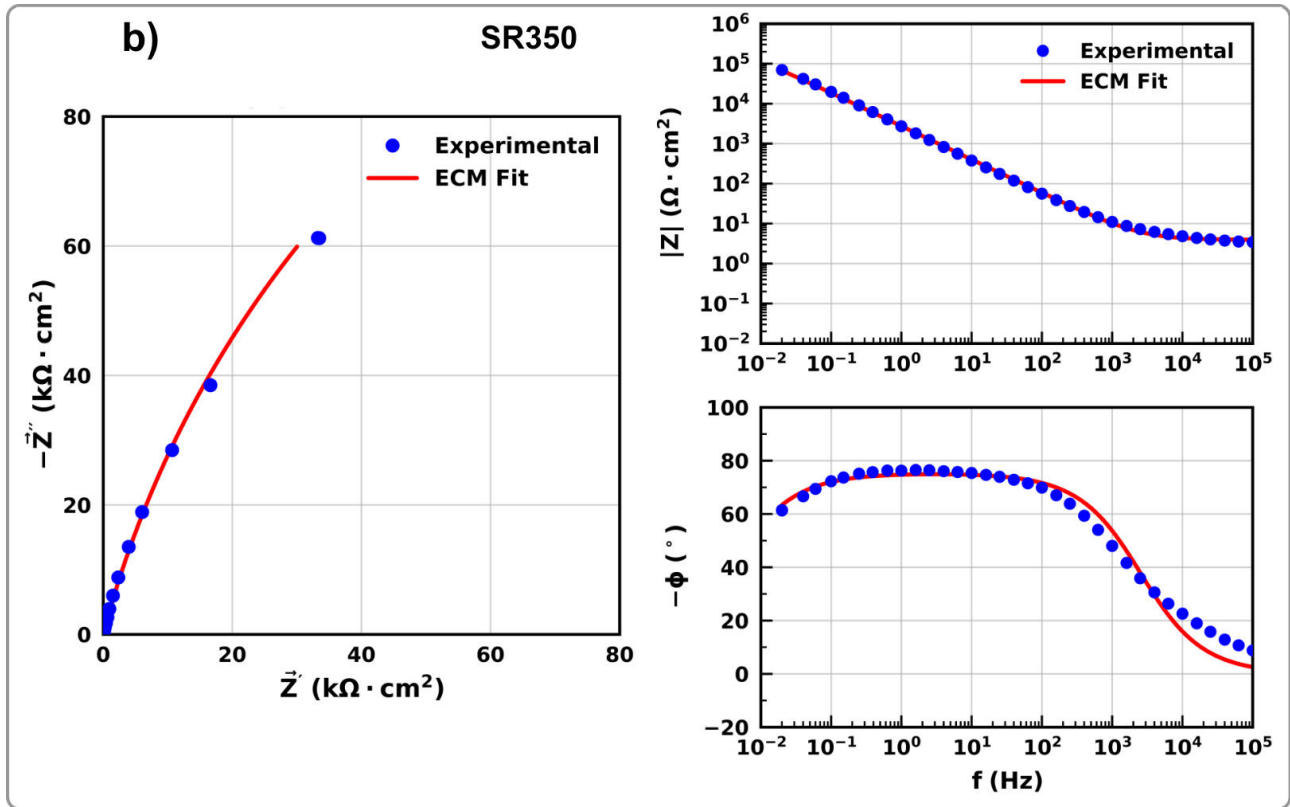


Figure 4.54: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated SDSS 2507 +6 wt.% Ni in SR350, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

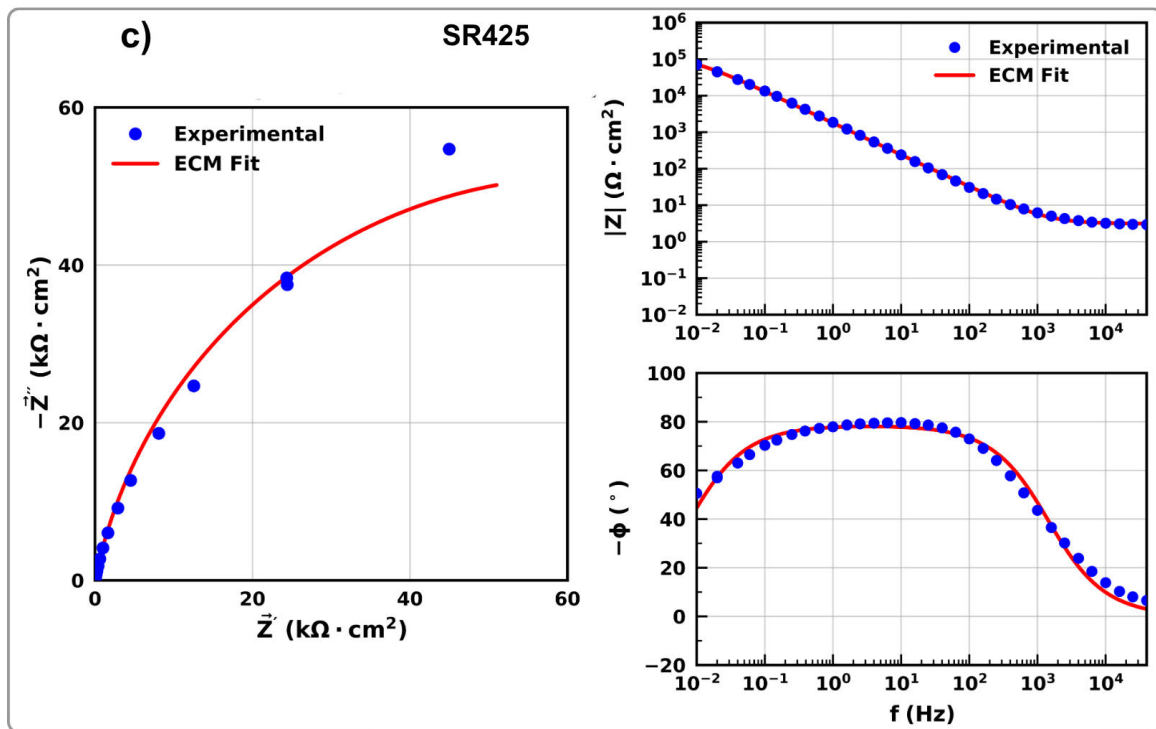


Figure 4.55: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated SDSS 2507 +6 wt.% Ni in SR425, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

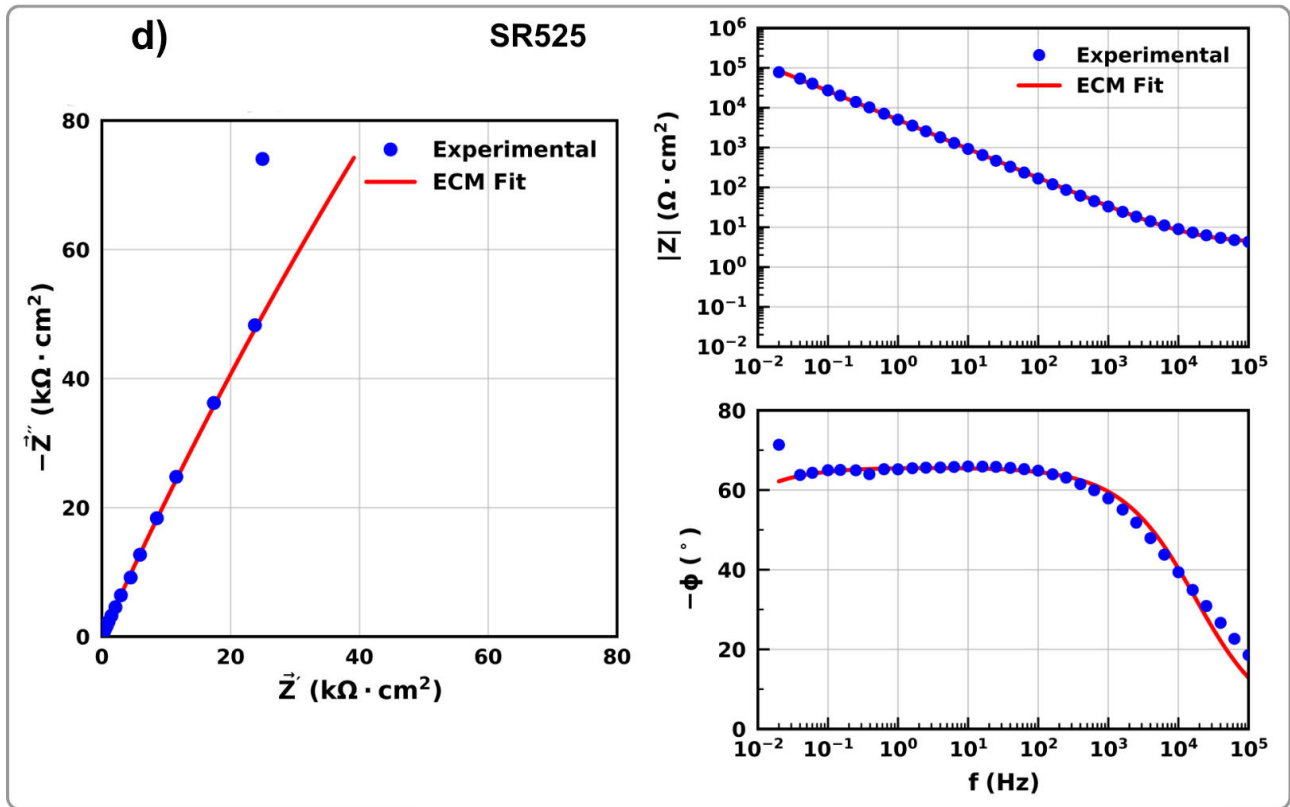


Figure 4.56: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated SDSS 2507 +6 wt.% Ni in SR525, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

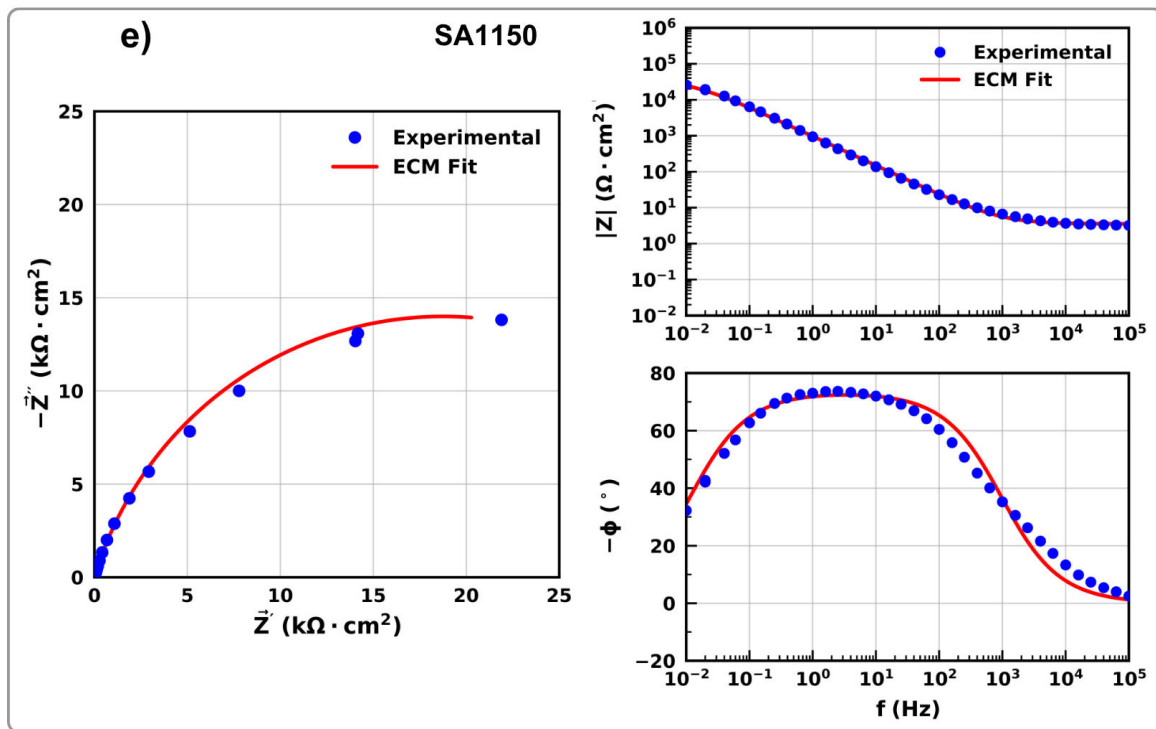


Figure 4.57: Electrochemical impedance spectroscopy (EIS) response of LPBF-fabricated SDSS 2507 +6 wt.% Ni in SA1150, shown as Nyquist, Bode magnitude, and Bode phase plots. Experimental data are compared with the ECM fit.

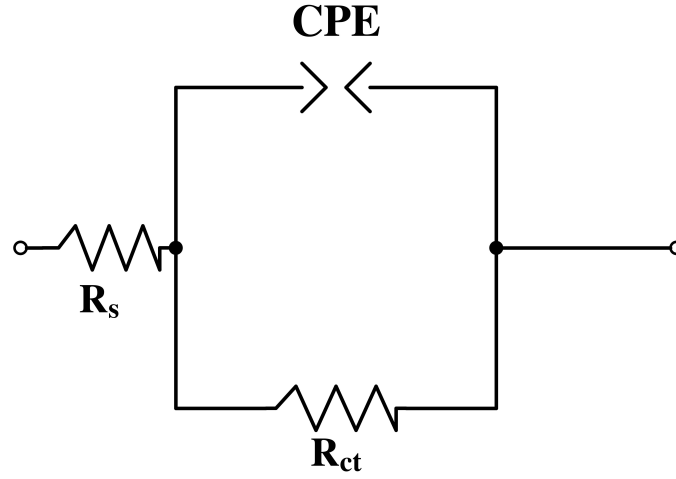


Figure 4.58: Equivalent circuit model employed for fitting the EIS spectra of the standard SDSS 2507 and Ni-modified SDSS 2507 alloy systems. The model comprises the solution resistance, R_s , connected in series with a parallel branch containing a constant phase element, CPE, and the charge-transfer resistance, R_{ct} .

Table 4.11 summarizes the equivalent circuit model (ECM) fit quality for each electrochemical impedance spectroscopy (EIS) spectrum using the fit statistic, χ^2 , and the maximum absolute residuals, $\max|\Delta Z'|$ and $\max|\Delta Z''|$.

Table 4.11: Fit-quality summary for the electrochemical impedance spectroscopy (EIS) equivalent circuit model (ECM) analysis. The table reports the fit statistic or fitting objective, χ^2 , together with the maximum absolute residuals in the real and imaginary impedance components, $\max|\Delta Z'|$ and $\max|\Delta Z''|$, respectively, obtained from the comparison between the measured impedance spectra and the fitted ECM response for each alloy condition.

Alloy system	Condition	χ^2	$\max \Delta Z' $ ($\Omega \cdot \text{cm}^2$)	$\max \Delta Z'' $ ($\Omega \cdot \text{cm}^2$)
Standard SDSS 2507	AS	0.0726	1667.14	2111.20
Standard SDSS 2507	SR300	0.3577	3452.73	4625.21
Standard SDSS 2507	SR400	1.5083	16 209.01	32 136.25
Standard SDSS 2507	SR550	1.1273	1378.96	835.33
Standard SDSS 2507	SA1100	0.1081	4054.74	1686.59
SDSS 2507 +3 wt.% Ni	AS	0.1875	725.29	1323.23
SDSS 2507 +3 wt.% Ni	SR400	0.5671	3977.32	5360.95
SDSS 2507 +3 wt.% Ni	SR450	1.0490	968.05	23 387.85
SDSS 2507 +3 wt.% Ni	SR500	0.4602	12 133.08	2199.08
SDSS 2507 +3 wt.% Ni	SR550	0.6084	8210.04	17 889.83
SDSS 2507 +3 wt.% Ni	SA1100	0.0365	1082.63	4567.03
SDSS 2507 +6 wt.% Ni	AS	0.1982	5986.54	4900.46
SDSS 2507 +6 wt.% Ni	SR350	0.1720	3430.84	2218.26
SDSS 2507 +6 wt.% Ni	SR425	0.0662	6028.00	4568.62
SDSS 2507 +6 wt.% Ni	SR525	0.0534	14 196.17	2415.59
SDSS 2507 +6 wt.% Ni	SA1150	0.1688	1845.51	338.45

Table 4.12: Electrochemical impedance fitting parameters for representative EIS spectra of LPBF-fabricated standard SDSS 2507 in the AS, SR300, SR400, SR550, and SA1100 conditions. Values correspond to best-response replicate-level records selected using the highest fitted charge-transfer resistance, R_{ct} ; therefore, the table reports trace-level upper-bound impedance behaviour. Parameters include solution resistance (R_s), constant-phase-element parameter (Y_{dl}), CPE exponent (n), effective double-layer capacitance (C_{dl}), and fitted charge-transfer resistance (R_{ct}).

Condition	R_s ($\Omega \cdot \text{cm}^2$)	Y_{dl} ($\mu\text{F} \cdot \text{s}^{(n-1)} \text{cm}^{-2}$)	n	C_{dl} ($\mu\text{F} \cdot \text{cm}^{-2}$)	R_{ct} ($\text{k}\Omega \cdot \text{cm}^2$)
AS	7.99	84.0	0.709	181	78.35
SR300	6.99	49.0	0.705	90	87.10
SR400	0.24	50.0	0.958	55	157.30
SR550	0.17	179.0	0.896	184	7.18
SA1100	13.55	70.0	0.772	133	121.50

Table 4.13: Electrochemical impedance fitting parameters for representative EIS spectra of LPBF-fabricated SDSS 2507 +3 wt.% Ni in the AS, SR400, SR450, SR500, SR550, and SA1100 conditions. Values correspond to best-response replicate-level records selected using the highest fitted charge-transfer resistance, R_{ct} ; therefore, the table reports trace-level upper-bound impedance behaviour. Parameters include solution resistance (R_s), constant-phase-element parameter (Y_{dl}), CPE exponent (n), effective double-layer capacitance (C_{dl}), and fitted charge-transfer resistance (R_{ct}).

Condition	R_s ($\Omega \cdot \text{cm}^2$)	Y_{dl} ($\mu\text{F} \cdot \text{s}^{(n-1)} \text{cm}^{-2}$)	n	C_{dl} ($\mu\text{F} \cdot \text{cm}^{-2}$)	R_{ct} ($\text{k}\Omega \cdot \text{cm}^2$)
AS	3.67	177.00	0.81	194.00	8.39
SR400	2.91	91.00	0.78	114.00	24.82
SR450	3.85	61.10	0.78	214.00	1547.00
SR500	4.57	79.00	0.81	152.00	214.20
SR550	3.44	62.30	0.79	219.00	1883.00
SA1100	9.24	81.80	0.82	128.00	98.63

Table 4.14: Electrochemical impedance fitting parameters for representative EIS spectra of LPBF-fabricated SDSS 2507 +6 wt.% Ni in the AS, SR350, SR425, SR525, and SA1150 conditions. Values correspond to best-response replicate-level records selected using the highest fitted charge-transfer resistance, R_{ct} ; therefore, the table reports trace-level upper-bound impedance behaviour. Parameters include solution resistance (R_s), constant-phase-element parameter (Y_{dl}), CPE exponent (n), effective double-layer capacitance (C_{dl}), and fitted charge-transfer resistance (R_{ct}).

Condition	R_s ($\Omega \cdot \text{cm}^2$)	Y_{dl} ($\mu\text{F} \cdot \text{s}^{(n-1)} \text{cm}^{-2}$)	n	C_{dl} ($\mu\text{F} \cdot \text{cm}^{-2}$)	R_{ct} ($\text{k}\Omega \cdot \text{cm}^2$)
AS	5.62	76.36	0.82	11.98	333.40
SR350	3.87	78.24	0.84	14.85	314.30
SR425	3.09	112.20	0.87	37.72	125.20
SR525	3.96	9.75	0.73	5.45	1251.00
SA1150	3.52	181.00	0.82	35.58	37.52

4.8 Corrosion-screening results

This section reports the machine-learning-assisted corrosion-screening outputs for the investigated laser powder bed fusion (LPBF)-fabricated standard and nickel-modified super duplex stainless steel (SDSS) 2507 alloy systems. The screening results are presented after the measured open-circuit potential (OCP), cyclic potentiodynamic polarization (CPP), and electrochemical impedance spectroscopy (EIS) responses so that the direct electrochemical measurements remain the main basis for interpretation. The screening output is used only to compare corrosion-related descriptor trends across alloy chemistry, LPBF processing route, post-build heat-treatment condition, electrolyte environment, and selected electrochemical responses. It is not treated as independent proof of corrosion resistance.

The reported model response is the apparent polarization-resistance descriptor, $R_{p,app}$, derived from a selected point on the CPP curve at the repassivation potential, E_{rp} . Here, $R_{p,app}$ is intentionally distinguished from the conventional polarization resistance, R_p , which is obtained from near-open-circuit polarization behaviour or from the Stern–Geary relation. Therefore, $R_{p,app}$ is interpreted only as an operational CPP-derived screening descriptor. It is not a mechanistic polarization-resistance value and is not used as a direct measure of corrosion resistance. The SDSS 2507 +3 weight percent nickel (+3 wt.% Ni) alloy outputs are reported as training-domain consistency results because six experimental +3 wt.% Ni records were included in model training. In this context, training-domain consistency means agreement within the composition and descriptor space already represented during model fitting. The standard SDSS 2507 and SDSS 2507 +6 weight percent nickel (+6 wt.% Ni) outputs are treated as condition-level comparisons outside the closest experimentally represented composition. This separation is retained throughout the section to avoid presenting training-domain agreement and cross-alloy transfer behaviour as equivalent evidence.

Figure 4.59 summarizes the descriptor space and target-response distribution used for the corrosion-screening analysis. Panel (a) presents the composition-based structure-awareness map, where nickel (Ni), nitrogen (N), molybdenum (Mo), chromium (Cr), combined chromium and molybdenum content (Cr+Mo), and combined nitrogen and nickel content (N+Ni) descriptors are plotted against the scaled response, $\mathcal{R}$, derived from $R_{p,app}$. The outer phase-ring estimates based on the chromium-equivalent to nickel-equivalent ratio, Cr_{eq}/Ni_{eq} , provide composition–phase-balance context. Panel (b) summarizes categorical processing and environmental descriptors, including manufacturing route, post-processing condition, cooling medium, and electrolyte environment. Panel (c) presents numerical processing descriptors, including heat-treatment temperature and cooling duration. Panel (d) shows the distribution of the scaled target response.

The following subsections report internal model consistency, model-domain coverage, feature sensitivity, predicted trends, and comparison between the screening output and the measured electrochemical descriptors for the standard SDSS 2507, SDSS 2507 +3 wt.% Ni, and SDSS 2507 +6 wt.% Ni alloy systems.

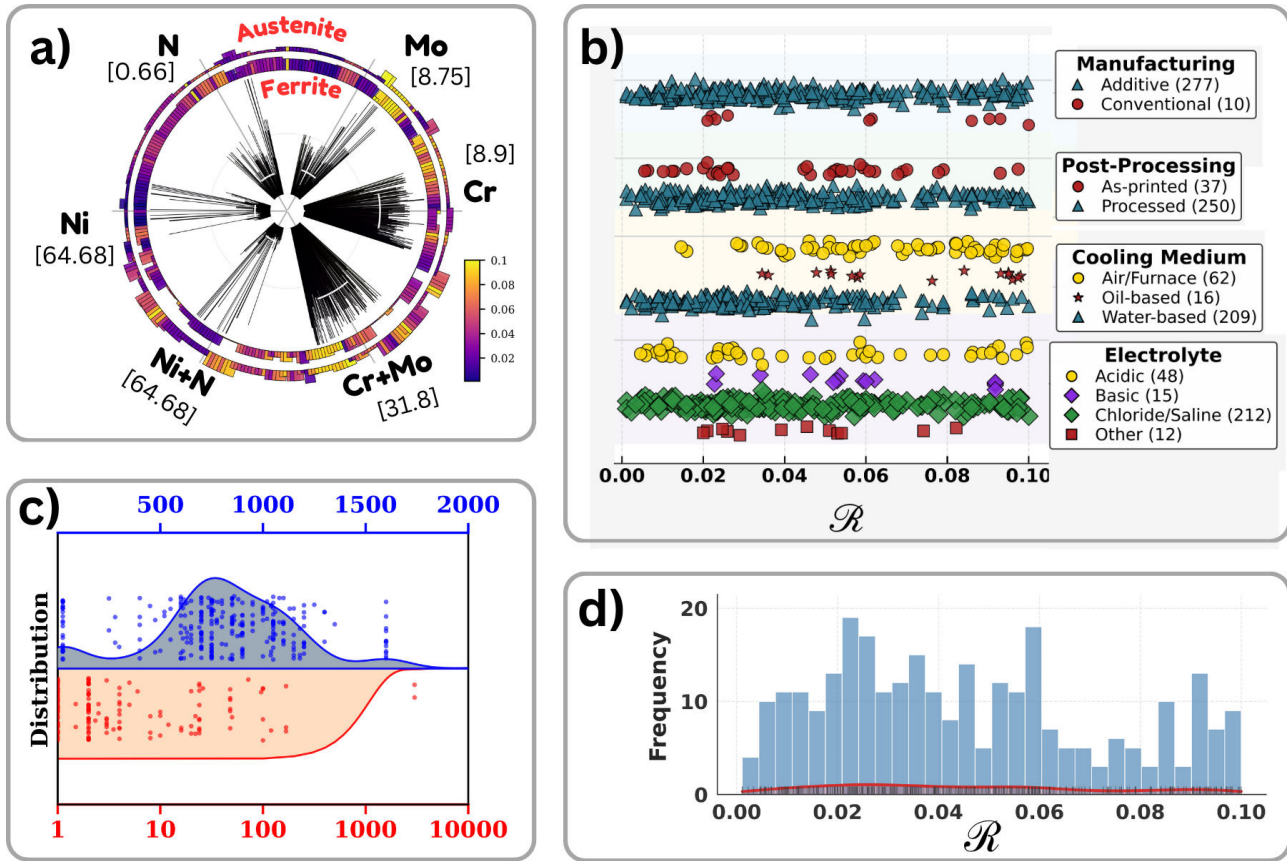


Figure 4.59: Descriptor-space and target-response summary used for corrosion-screening model development: (a) composition-based structure-awareness map, (b) categorical processing and environmental descriptors, (c) numerical processing descriptors, and (d) frequency distribution of the scaled target response, $\mathcal{R}$.

4.8.1 Internal model consistency

The 10-fold ensemble model showed strong internal agreement between reported and predicted apparent polarization-resistance descriptor values, $R_{p,app}$, as shown in Figure 4.60. The parity plot compares measured and predicted $R_{p,app}$ values across the available training-domain records, with the data points positioned close to the one-to-one reference line. The corresponding performance metrics are summarized in Table 4.15, including the coefficient of determination, R^2 , mean absolute error (MAE), root mean square error (RMSE), and fold-wise R^2 values. The model performance values show low prediction error within the available dataset and consistent agreement across the cross-validation folds. The six SDSS 2507 +3 wt.% Ni records were included in the training dataset; therefore, their agreement is reported as training-domain consistency rather than independent external validation. The standard SDSS 2507 and SDSS 2507 +6 wt.% Ni predictions are reported separately in the following subsections as condition-level comparisons outside the closest represented experimental composition.

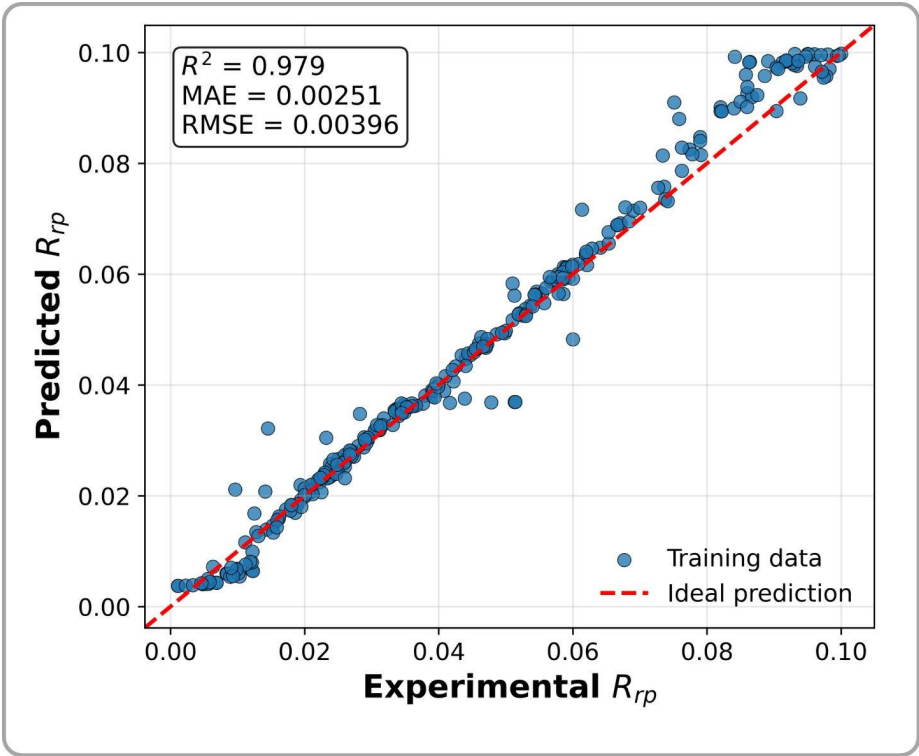


Figure 4.60: Internal parity assessment of the corrosion-screening neural-network model. Reported experimental apparent polarization-resistance values, $R_{p,app}$, are compared with the corresponding predictions obtained from the saved 10-fold neural-network ensemble. The proximity of the data points to the parity line indicates good agreement within the training domain and supports internal consistency of the loaded model. This comparison is used as an internal model-performance check and is not treated as independent external validation.

Table 4.15: Internal performance summary of the corrosion-screening model.

Metric	Value
<i>Internal agreement metrics</i>	
Coefficient of determination, R^2	0.9786
Mean absolute error, MAE	0.0025
Root mean square error, RMSE	0.0040
<i>Internal consistency</i>	
Cross-validation approach	10-fold ensemble
Fold-wise stability	$R^2 > 0.95$
Assessment scope	Training domain

4.8.2 Model-domain coverage

Figure 4.61 summarizes the composition-domain coverage of the corrosion-screening training database. The figure reports the elemental composition ranges represented in the machine-learning training dataset and the number of datasets in which each element is present. These outputs define the compositional space supporting the model response before the predicted screening

trends are compared across the investigated SDSS 2507 alloy systems. The +3 wt.% Ni-modified SDSS 2507 records are located within the training domain because these experimental records were included in the model-development dataset. Their predicted outputs are therefore reported as training-domain consistency results. In contrast, the standard SDSS 2507 and +6 wt.% Ni alloy conditions are reported as transfer-assessment cases because their compositions do not coincide with the closest experimentally represented Ni-modified training condition.

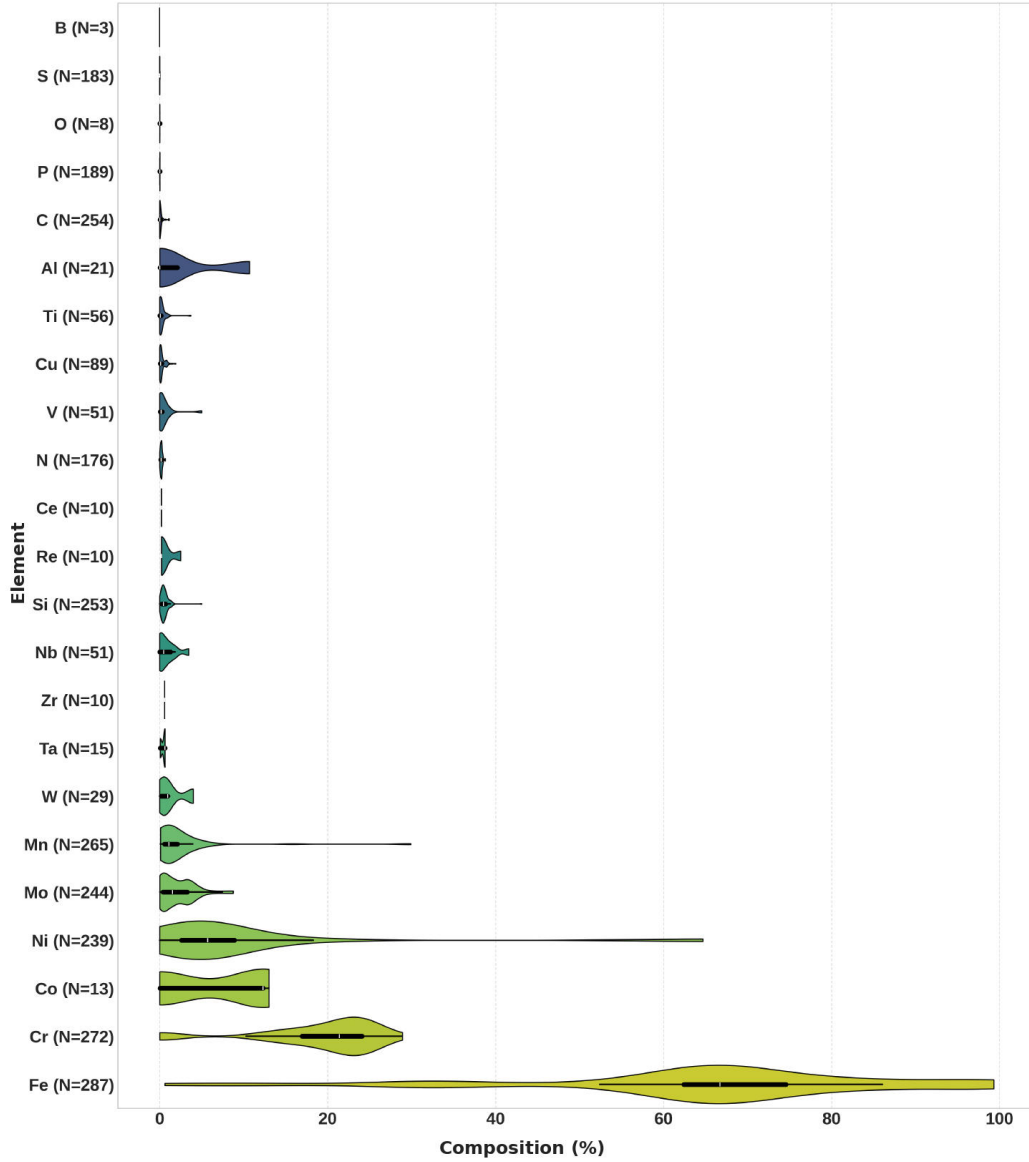


Figure 4.61: Composition-domain coverage of the corrosion-screening training database, showing elemental composition ranges and dataset counts for the represented alloying elements.

4.8.3 Feature sensitivity

Feature-sensitivity analysis ranked Ni, Cr, and Mo among the main contributors to the predicted $R_{p,app}$, while processing-related categorical variables represented manufacturing route, post-processing condition, cooling route, and electrolyte environment. The retained feature groups

and derived interaction terms are summarized in Table 4.16. These feature-sensitivity results indicate model sensitivity only. The retained interaction terms are engineered descriptors for screening-level prediction; they do not independently establish causal corrosion mechanisms, phase-resolved electrochemical behaviour, or passive-film chemistry.

Table 4.16: Feature groups and derived interaction terms retained in the corrosion-screening feature space.

Feature group	Representative descriptors	Screening-level role
Ni-related descriptors	Ni, Ni^2 , $\log(\text{Ni})$, $\text{Ni} \times \text{Cr}$	Represent Ni addition, nonlinear Ni sensitivity, and coupled Ni–Cr composition effects associated with austenite stabilization and composition-dependent screening response.
Cr/Mo and PREN-related descriptors	Cr, Mo, PREN, $\text{PREN} \times T$, $\text{PREN} \times T/t$	Represent passivity-relevant bulk alloy chemistry, ferrite-stabilizing composition effects, nominal pitting-resistance tendency, and its coupling with thermal-processing variables.
Heat-treatment descriptors	de- T , t , $T \times t$, T/t , $T \log(t)$, T^2 , t^2	Represent nominal thermal exposure, coupled temperature–time effects, and nonlinear or threshold-like heat-treatment responses in the descriptor space.
Processing and environment categories	Manufacturing route, post-processing condition, cooling route, and electrolyte descriptors	Retain categorical information describing fabrication route, post-build condition, cooling class, and test environment used for screening-level prediction.

4.8.4 Predicted trends

Figure 4.62 presents the model architecture and screening-level predicted trends. Panel (a) shows the neural network (NN) architecture used to predict the apparent polarization-resistance descriptor, $R_{p,\text{app}}$. The input groups include elemental-composition descriptors, X_{El} , processing variables, X_{Pr} , atomic/bulk descriptors, $X_{\text{Atomic-Bulk}}$, and engineered interaction terms, X_{Intr} . The descriptor representation also includes the Structurally Panelized Element Representation (SPER), which introduces composition-sensitive weighting into the input space. Panel (b) shows the conditional Ni-response projection under *ceteris paribus* conditions, where Ni content is varied and all other descriptors are held fixed. In this projection, Ni content is varied while the remaining descriptors are held fixed within the selected reference condition. The predicted $R_{p,\text{app}}$ response increases nonlinearly with Ni content. The shaded regions mark low-Ni ferrite-favouring and high-Ni austenite-favouring compositional domains; these regions are descriptor-based domains and do not represent directly measured phase fractions. Panel (c) shows the projection of eight alloying elements grouped as pro-ferritic elements, Mo, Cr, Si, and Nb, and pro-austenitic elements, Ni, Mn, C, and Cu. The marker positions are determined by normalized compositional weighting, and the marker colour represents the predicted $R_{p,\text{app}}$. Panel (d) shows the pitting resistance equivalent number (PREN)–SPER triangular projection of candidate stainless-steel compositions. Each marker represents a generated alloy composition positioned by barycentric normalization

with auto-scaled axes. The colour scale in panels (c) and (d) corresponds to the predicted $R_{p,app}$.

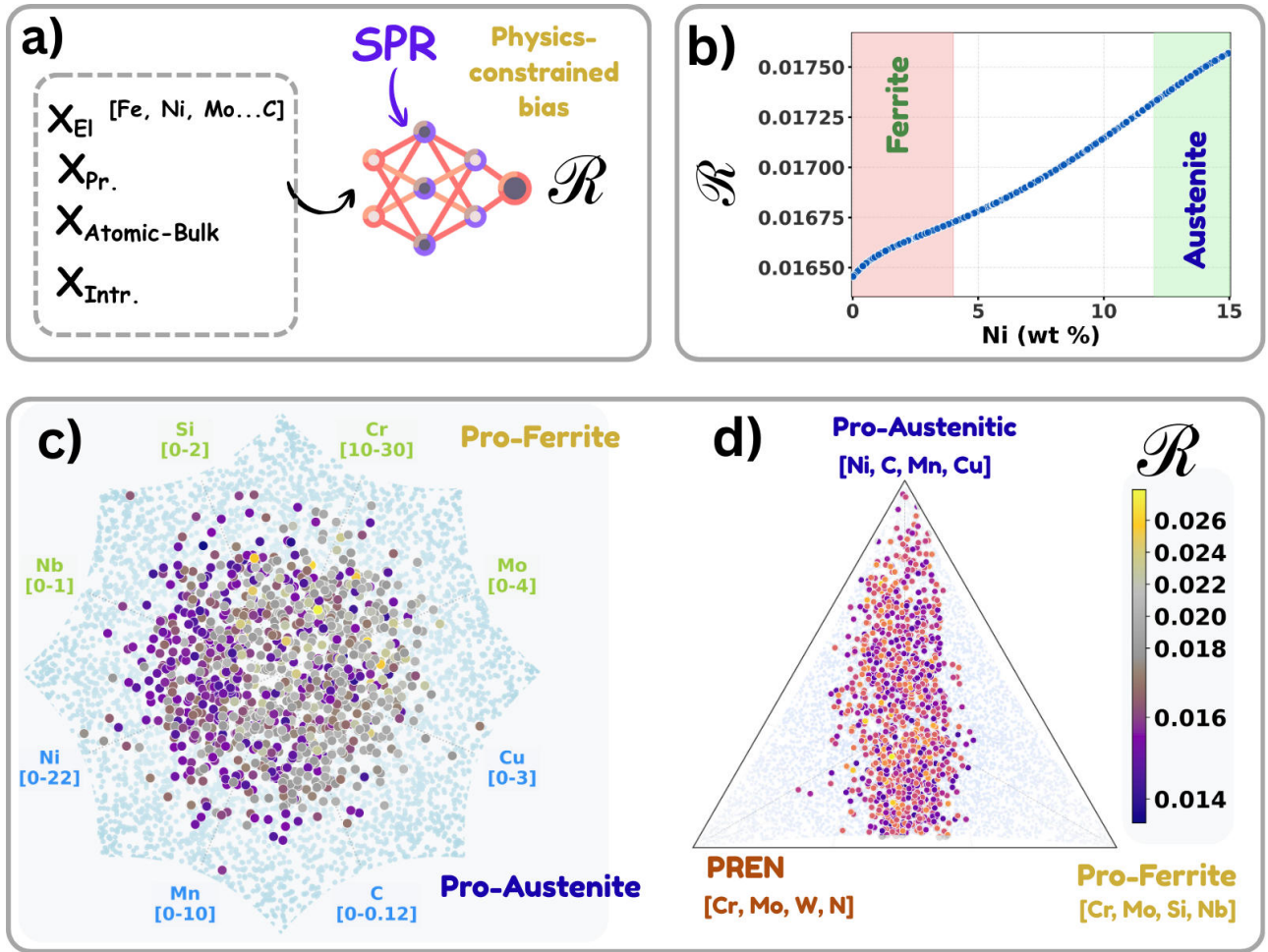


Figure 4.62: Model architecture and screening-level corrosion-response trends: (a) Neural network architecture, (b) Conditional Ni-response projection, (c) Alloying-element projection, and (d) PREN-SPER triangular projection. Marker colour represents the predicted $R_{p,app}$.

For the 3 wt.% Ni-modified SDSS 2507 condition, the model response is reported as training-domain consistency. For the standard SDSS 2507 and +6 wt.% Ni alloys, the outputs are reported as transfer-assessment trends because these compositions lie outside the closest experimentally represented training condition, as summarized in Figure 4.61.

4.8.5 Electrochemical comparison

The model-derived trends were compared with the measured open-circuit potential (OCP), cyclic potentiodynamic polarization (CPP), and electrochemical impedance spectroscopy (EIS). Agreement was assessed using measured R_p , fitted charge-transfer resistance R_{ct} , breakdown behavior, and repassivation behavior. The comparison was used to determine whether the descriptor-level screening response followed the same ranking as the direct electrochemical measurements. Disagreement between model-derived and measured trends was treated as a limitation of descriptor transfer rather than as evidence against the measured corrosion response. The experimental OCP, CPP, and EIS data remained the primary basis for laboratory electrochemical interpretation.

Table 4.17: Experimental comparison used to assess the corrosion-screening trends.

Comparison item	Result
<i>Standard SDSS 2507</i>	
Measured trend	SR400 showed the most favorable CPP-derived response, with the lowest J_{corr} , highest R_p , and most positive E_{corr} . SR550 showed the weakest CPP-derived response. EIS identified SR400 as the highest- R_{ct} standard condition.
Screening interpretation	Treated as a lower-Ni transfer-assessment case; not used as independent validation.
<i>SDSS 2507 +3 wt.% Ni</i>	
Measured trend	The as-built condition showed the highest J_{corr} and lowest R_p . SR500 and SR550 gave the highest CPP-derived R_p values among the stress-relieved states, while EIS showed high R_{ct} values for SR550 and SR450.
Screening interpretation	Treated as a training-domain case because the corresponding datasets were included in model training.
<i>SDSS 2507 +6 wt.% Ni</i>	
Measured trend	The as-built and stress-relieved states showed relatively low J_{corr} values, while SA1150 showed the highest J_{corr} , lowest R_p , and negative E_{OCP} . EIS identified SR525 as the highest- R_{ct} condition and SA1150 as the lowest- R_{ct} condition.
Screening interpretation	Treated as a higher-Ni transfer-assessment case. The SA1150 divergence was retained as a transfer-limit observation.

Overall, the corrosion-screening model captured composition-sensitive trends in $R_{p,\text{app}}$, particularly through Ni-, Cr-, and Mo-related descriptors and the nonlinear composition–processing interaction terms summarized in Table 4.16. This indicates that the model responded to chemically meaningful variables associated with austenite stabilization, passive-film-forming elements, and nominal pitting-resistance descriptors. However, the experimental comparison showed that high predicted apparent resistance did not necessarily imply stable repassivation, higher breakdown resistance, or superior localized-corrosion behaviour. This distinction is important because $R_{p,\text{app}}$ is an apparent CPP-derived screening descriptor extracted from a selected point on the reverse scan, whereas localized-corrosion resistance also depends on breakdown potential, repassivation potential, hysteresis behaviour, phase-specific chemistry, passive-film stability, and defect-assisted pit initiation. Divergent cases are therefore reported here as screening-limit observations rather than as model failures or mechanistic proof. Their metallurgical interpretation is reserved for Chapter 5, where the predicted trends are assessed together with SEM, XRD, EBSD, CPP, and EIS evidence.

Chapter 5

Discussion

The behaviour of laser powder bed fusion (LPBF)-manufactured super duplex stainless steel (SDSS) 2507 is best interpreted as a pathway-dependent metallurgical response. Across the examined conditions, porosity, nominal nickel (Ni) content, post-build heat treatment, phase constitution, crystallographic state, defect population, and electrochemical response evolved interdependently. As a result, ranking the alloys by Ni content, heat-treatment temperature, porosity, or phase fraction alone would underrepresent the metallurgical complexity of the system. Their performance is instead considered through the combined evolution of processing, structure, properties, and electrochemical behaviour. On this basis, the chapter addresses the aim, six objectives, six research questions, and three working hypotheses formulated in Chapter 1, while identifying criterion-specific conditions rather than imposing one universal optimum.

The feedstock and LPBF process-window results provide the baseline for this interpretation. The common LPBF parameter set enabled low-porosity fabrication of the standard SDSS 2507 alloy and the 3 wt.% Ni- and 6 wt.% Ni-modified variants. At the same time, the blended-powder alloy design and alloy-specific heat-treatment schedules place the comparison within a pathway-based framework rather than a fully factorial Ni-content series. The discussion therefore proceeds from the LPBF processing and feedstock-alloy basis, through phase-balance and heat-treatment response, to mechanical behaviour, electrochemical response, and final condition selection.

The first metallurgical implication is that densification and duplex phase restoration are distinct outcomes. The selected LPBF condition produced low-porosity material suitable for comparing the standard, 3 wt.% Ni-modified, and 6 wt.% Ni-modified SDSS 2507 alloy systems (Figure 4.2). However, reduced porosity did not imply restoration of a wrought-like duplex state. The porosity-screened parameter set reduced macroscopic defect content within the examined optical fields, but ferrite-to-austenite transformation remained governed by alloy chemistry and subsequent heat treatment. Low porosity therefore provides a processing basis for comparison, but it does not establish metallurgical suitability in terms of duplex phase balance, local chemistry, or corrosion response. LPBF densification primarily addresses melt-pool consolidation and defect formation, whereas duplex restoration requires sufficient thermodynamic and kinetic conditions for austenite nucleation and growth.

The electron backscatter diffraction (EBSD) phase maps provide the phase-constitution basis for this interpretation. The standard and +3 wt.% Ni alloys remained strongly ferrite-rich in the as-built and stress-relieved states, whereas the +6 wt.% Ni alloy showed a substantially higher mapped austenite fraction before solution annealing (Figures 4.10, 4.11, and 4.12). This indicates that Ni addition altered the available ferrite-to-austenite transformation pathway, but did not act as a uniformly sufficient duplex-restoration mechanism. Within the investigated LPBF thermal cycle, +3 wt.% Ni was insufficient to overcome ferrite retention, while +6 wt.% Ni

shifted the mapped phase constitution toward near-duplex or austenite-rich states depending on heat treatment. This interpretation remains bounded by the investigated powder mixture, LPBF condition, heat-treatment schedules, mapped areas, and section orientations.

The +6 wt.% Ni alloy further demonstrates that phase balance cannot be treated as a monotonic optimisation problem. The as-built (AS), stress-relieved at 350 °C for 2 h (SR350), and stress-relieved at 425 °C for 1 h (SR425) conditions were near-duplex based on the EBSD-mapped local phase-fraction descriptor. In contrast, the stress-relieved at 525 °C for 1 h (SR525) condition was austenite-rich, and the solution-annealed at 1150 °C for 10 min (SA1150) condition was strongly austenite-rich. The SA1150 condition reached 89.1 % γ -Fe, which is not a balanced duplex state. This indicates that stronger austenite stabilisation and high-temperature annealing can shift the mapped phase constitution beyond duplex restoration toward austenite enrichment. Such enrichment can improve elongation and impact response by increasing the austenite contribution to plastic deformation, but it can also reduce strength and modify electrochemical behaviour.

The X-ray diffraction (XRD) results provide a complementary diffraction-based view of the post-processing response. Because the three alloy systems started from different AS baselines and used alloy-specific stress-relief schedules, the normalized trajectories in Figure 4.8 are interpreted as within-alloy trends rather than direct cross-alloy rankings. In the standard SDSS 2507 alloy, the α -Fe-reflection-derived descriptors showed an overall decrease in β_{med} and ε , together with an increase in apparent D , after stress relief, with the lowest stress-relief broadening observed in SR550. Combined with the EBSD phase maps, this supports the interpretation that stress relief mainly modified the ferritic diffraction-broadening state rather than producing substantial ferrite–austenite phase rebalancing.

The +3 wt.% Ni alloy contradicts a simple monotonic recovery-controlled interpretation. SR500 produced the highest stress-relief β_{med} , ρ_D , and ε , together with the lowest apparent D , whereas SR550 shifted toward lower broadening and larger apparent D . This non-monotonic response indicates that the stress-relief sequence was governed by competing microstructural contributions rather than by recovery alone. These may include partial relaxation of the LPBF-inherited substructure, local solute redistribution, thermally activated ferritic-matrix changes, and phase-boundary- or defect-related strain fields. The broader property ranking supports this descriptor-based interpretation: SR400 and SR450 corresponded to the higher-strength conditions, SR500 gave the most favourable combined CPP response, and SR550 showed comparatively high apparent electrochemical resistance but weaker repassivation stability. Thus, the +3 wt.% Ni alloy cannot be ranked by stress-relief temperature alone.

The +6 wt.% Ni alloy provides the strongest phase-resolved diffraction evidence because both α -Fe and γ -Fe reflections were sufficiently resolved across the investigated sequence. Both phases showed increased broadening from AS to SR425, followed by partial reversal in SR525 and lower broadening after SA1150 (Figure 4.9). However, γ -Fe retained lower β_{med} , lower ρ_D , and larger apparent D than α -Fe across the complete sequence, whereas the γ -Fe microstrain descriptor remained slightly higher. This phase-dependent response indicates that ferrite and austenite did

not accommodate the post-build thermal exposure in the same way. The distinction is treated as a descriptor-level diffraction response, not as an absolute measurement of dislocation density, residual stress, or precipitation state.

The XRD-derived broadening descriptors and EBSD kernel average misorientation (KAM) results are therefore complementary rather than interchangeable. XRD peak broadening captured phase-sensitive changes averaged over the irradiated material volume, whereas KAM mapped local orientation gradients within the EBSD measurement field. Together, these descriptors constrain how stress relief and solution annealing modified the investigated alloys through diffraction broadening, local orientation heterogeneity, phase constitution, and boundary-network state. They are interpreted as comparative indicators of microstructural evolution, not as direct measurements of residual stress, absolute dislocation density, or precipitation state.

The nanoscale evidence imposes an important boundary on the interpretation of stress-relief mechanisms. The standard SR550 condition is the only condition in this dataset for which transmission electron microscopy (TEM), scanning transmission electron microscopy (STEM), selected-area electron diffraction (SAED), and STEM energy-dispersive X-ray spectroscopy (STEM-EDS) were used to examine the stress-relieved ferritic matrix directly. Scanning electron microscopy (SEM) indicated a ferrite-dominated structure with intergranular austenite and retained cell/subgrain traces (Figure 4.3). The TEM/STEM analysis identified nanoscale/substructural contrast within the analysed ferritic region (Figure 4.30). However, bright-field TEM, dark-field TEM, high-angle annular dark-field STEM (HAADF-STEM), SAED, and STEM-EDS did not provide direct confirmation of sigma, chi, chromium nitride, or other discrete secondary precipitates. The SAED pattern was indexed to ferrite, with no clearly resolved additional diffraction spots or ring features, and the STEM-EDS maps did not show correlated enrichment of Cr, Mo, Ni, or Mn associated with the observed contrast features.

SR550 should therefore not be interpreted as a confirmed precipitation condition. The evidence-supported interpretation is narrower: the observed contrast is consistent with nanoscale or substructural modification of the ferritic matrix, including recovery-related contrast, local matrix modification, early-stage clustering, ferritic instability, or other thermally activated changes that were not resolved as a discrete precipitate population in the present TEM/STEM dataset. This interpretation is compatible with reported precipitation-sensitive and low-temperature ferritic-instability ranges in duplex stainless steels, but the present evidence supports only nanoscale/substructural contrast, not a verified secondary-phase mechanism [105, 106, 107, 108]. A precipitation-based claim would require direct crystallographic and chemical confirmation.

The same evidential boundary applies to the +3 wt.% Ni and +6 wt.% Ni stress-relieved conditions. Their non-monotonic XRD, EBSD boundary-network, KAM, mechanical, and laboratory electrochemical responses are consistent with condition-specific modification of the LPBF-inherited microstructure. However, these responses do not independently prove sigma, chi, chromium nitride, or any other precipitate formation. Secondary-phase-related interpretations are therefore treated only as possible contributors, while the confirmed interpretation remains condition-specific

microstructural modification, including possible nanoscale/substructural change.

Austenite reformation after solution annealing provides the clearest link between thermal exposure, phase constitution, and crystallography. In the standard alloy, the solution-annealed at 1100 °C for 15 min (SA1100) condition produced an approximately balanced mapped duplex condition, whereas in the +3 wt.% Ni alloy, SA1100 produced an austenite-rich structure (Figures 4.10 and 4.11). High-temperature annealing, therefore, provided conditions consistent with extensive ferrite-to-austenite transformation, while stress relief mainly modified the ferrite-rich inherited state without comparable duplex restoration. This interpretation is consistent with duplex stainless steel metallurgy, where austenite reformation depends on temperature, holding time, local chemistry, ferrite fraction, and available nucleation sites [109, 110].

The boundary-network and orientation-relationship results indicate that austenite reformation was crystallographically constrained, not merely a change in phase fraction. In the +3 wt.% Ni alloy, the same-phase $\Sigma 3$ boundary-length fraction increased from 2.59 % in AS to 11.68 % after SA1100, while the total near-coincident-site-lattice (near-CSL) fraction increased from 10.01 % to 19.65 % (Figure 4.17). This increase is consistent with annealing-assisted austenite growth and the development of twin-related face-centred cubic boundary populations. The ferrite–austenite interphase-boundary misorientation distributions contained a main population near 43–46°, and SA1100 showed a pronounced population near 44–46° together with low deviations from ideal Kurdjumov–Sachs and Nishiyama–Wassermann variants (Figures 4.18 and 4.20). These data support the interpretation that a substantial fraction of ferrite–austenite interfaces followed transformation-related orientation relationships rather than forming as random phase boundaries [110]. The near-CSL data are used only as same-phase boundary-network descriptors, not as complete five-parameter grain-boundary character distributions.

The +6 wt.% Ni boundary-network response provides an additional constraint on the transformation pathway. High-angle grain boundaries (HAGBs) exceeded low-angle grain boundaries (LAGBs) in AS, SR350, SR425, and SA1150, whereas SR525 showed the lowest HAGB fraction, approximately 46 % (Figure 4.15). The dominant HAGB disorientation populations remained near 43–46°, but the SR525 boundary fraction indicates that this condition was not simply a higher-temperature continuation of SR350/SR425. Instead, SR525 represents a distinct mapped boundary-network state, consistent with a different balance among austenite enrichment, local orientation heterogeneity, and thermal modification of the LPBF-inherited structure.

The KAM results reinforce this interpretation. In the +3 wt.% Ni alloy, most indexed pixels had KAM values below 0.5°, indicating a predominantly low local orientation-gradient field (Figure 4.22). In the +6 wt.% Ni alloy, AS, SR350, SR425, and SA1150 also showed predominantly low-KAM fields, but SR525 showed a much broader KAM distribution. The fraction of indexed pixels below 0.5° decreased to approximately 60 %, and the pixel-mean KAM increased to 0.539° (Figure 4.23). This should not be over-interpreted as direct residual-stress evidence. KAM is a local orientation-gradient descriptor affected by step size, indexing quality, phase topology, grain size, boundary density, and local plastic or thermal strain gradients. The evidence-supported conclusion is that

SR525 has the highest mapped local orientation heterogeneity among the investigated +6 wt.% Ni states.

Texture and orientation-spread descriptors complete the crystallographic interpretation. Where grain reference orientation deviation (GROD), grain orientation spread (GOS), and orientation distribution function (ODF) descriptors were available, they are interpreted as comparative EBSD indicators of intragranular orientation heterogeneity and texture redistribution. Taken together, the EBSD and XRD evidence support the interpretation that the investigated heat treatments modified phase constitution, boundary network, local orientation heterogeneity, and diffraction peak broadening in alloy-specific and condition-specific ways.

The mechanical response follows from this processing–structure sequence. The central mechanical issue is whether each condition preserves LPBF-derived strengthening or replaces it with deformation and fracture tolerance through recovery and austenite enrichment. The as-built and stress-relieved conditions generally retained higher hardness and tensile strength, whereas solution annealing reduced strength but improved elongation and unnotched Charpy impact response (Figures 4.31 and 4.32–4.34). This response reflects the balance among phase constitution, LPBF-derived substructure, local orientation heterogeneity, recovery, austenite enrichment, and defect sensitivity [111, 112, 113, 109].

In the standard alloy, AS represented the strength-dominated condition, while SA1100 represented the ductility- and impact-response-dominated condition. The strength reduction after SA1100 is consistent with partial removal of LPBF-derived strengthening sources, including recovery of ferritic substructure, reduction of local orientation heterogeneity, and relaxation of stored-strain effects. At the same time, the approximately balanced ferrite–austenite mapped condition increased the austenite contribution to plastic deformation, improving strain accommodation and impact-energy absorption.

The +3 wt.% Ni alloy followed a related but more strength-favourable response. SR400 and SR450 provided the highest tensile-strength pathway in the investigated matrix, reaching approximately $R_m = 1414$ MPa. These conditions retained the LPBF-inherited strengthening state without producing the full recovery and softening associated with solution annealing. They are therefore suitable when peak tensile strength is the primary selection criterion, although ductility, impact response, and electrochemical descriptors follow different rankings.

The +6 wt.% Ni alloy shifted the mechanical response toward deformation capacity. Its AS and lower-temperature stress-relieved states did not reach the peak strength of the +3 wt.% Ni SR400/SR450 pathway, but they provided a more favourable non-solution-annealed elongation/impact compromise. This is consistent with EBSD evidence showing a much higher mapped austenite fraction in the +6 wt.% Ni alloy than in the standard or +3 wt.% Ni AS/SR states. However, SR525 requires cautious interpretation because it combined increased mapped austenite fraction with the highest KAM heterogeneity and the lowest HAGB fraction among the +6 wt.% Ni conditions. SA1150 provided the strongest elongation/impact-oriented response, but also the lowest strength and a strongly austenite-rich mapped phase balance.

The fracture surface observations are consistent with this condition-dependent mechanical response (Figure 4.35). Pores or lack-of-fusion-type defects acted as local stress concentrators, while cleavage-like facets or lamellae, secondary microcracks, tear ridges, shear steps, and sparse dimpling reflected different balances between brittle fracture, localised plasticity, and crack deflection. Ferrite-rich LPBF structures promoted strength but also increased sensitivity to defect-assisted fracture, whereas austenite reformation/enrichment and recovery improved plastic dissipation and crack-blunting capacity [114, 115, 116]. The Charpy and fractography results are therefore interpreted as comparative impact and fracture-morphology evidence within the tested matrix, not as fracture-toughness measurements.

The electrochemical response follows the same processing–structure logic, but no single corrosion descriptor captures the full behaviour. A condition that favours strength or austenite enrichment does not necessarily maximise passivity, breakdown resistance, repassivation, or fitted interfacial resistance. The electrochemical response is therefore interpreted by separating open-circuit potential (OCP) stability, cyclic potentiodynamic polarisation (CPP) descriptors, electrochemical impedance spectroscopy (EIS) descriptors, and screening-level susceptibility to localised corrosion. OCP is treated as a potential-stability descriptor. CPP provides descriptors for corrosion current density, Stern–Geary polarisation resistance, breakdown potential, and repassivation potential, whereas EIS is interpreted through fitted interfacial-resistance descriptors, primarily the charge-transfer resistance, R_{ct} . These quantities are complementary, but they are not interchangeable measures of corrosion resistance [117, 25, 93]. Accordingly, the EIS ranking is based on representative upper-bound response traces selected from the tested replicate-level records and should not be interpreted as statistical mean corrosion behaviour. Similarly, $R_{p,app}$ is used only as an apparent CPP-derived screening descriptor, not as conventional polarisation resistance or as mechanistic proof of localised-corrosion resistance. Nominal PREN and nominal Cr/Mo contents are also insufficient to define the localised-corrosion response without phase-specific chemistry and defect assessment, because passive-film behaviour and pit initiation are affected by Cr–Mo–N partitioning, local depletion or enrichment, oxide-film chemistry, phase morphology, retained LPBF substructure, and surface-connected or near-surface defects.

In the standard alloy, SR400 gave the most favourable combined CPP/EIS descriptor set, with $J_{corr} = 0.06 \mu A cm^{-2}$, $R_p = 301.10 k\Omega cm^2$, and representative upper-bound $R_{ct} = 157.30 k\Omega cm^2$ (Tables 4.8 and 4.12). This supports SR400 as the preferred standard-alloy condition when laboratory electrochemical descriptors are prioritised. The implication is not that SR400 is universally superior, but that its particular combination of phase constitution, retained LPBF substructure, stress-relief state, and electrochemical response was most favourable within this descriptor set. By contrast, SR550 gave weaker laboratory electrochemical descriptors, with $J_{corr} = 1.16 \mu A cm^{-2}$, $R_p = 17.68 k\Omega cm^2$, $E_{br} = 130 mV$, $E_{rp} = -96.3 mV$, and representative upper-bound $R_{ct} = 7.18 k\Omega cm^2$ (Tables 4.8 and 4.12), showing that higher stress-relief temperature did not produce monotonic electrochemical improvement. Together with the SEM and TEM/STEM evidence discussed above, the SR550 response is most consistently interpreted as

the electrochemical outcome of the LPBF-inherited ferritic matrix, retained substructural features, possible nanoscale/substructural changes, and/or defect-assisted initiation pathways, rather than as proof of a confirmed secondary-precipitate mechanism.

In the +3 wt.% Ni alloy, stress relief improved selected laboratory electrochemical descriptors relative to AS, but the preferred condition depended on the descriptor used for selection. SR550 reduced J_{corr} to $0.06 \mu\text{A cm}^{-2}$ and increased R_p to $216.84 \text{ k}\Omega \text{ cm}^2$, while EIS showed R_{ct} increasing from $8.39 \text{ k}\Omega \text{ cm}^2$ in AS to $1883 \text{ k}\Omega \text{ cm}^2$ after SR550. SR450 also gave a high fitted R_{ct} of $1547 \text{ k}\Omega \text{ cm}^2$ (Tables 4.9 and 4.13). This divergence indicates that the response cannot be reduced to one electrochemical variable. SR400/SR450 remain strength-driven choices, SR450/SR550 are favoured by fitted EIS resistance descriptors, and SR500/SR550 are more favourable when CPP-derived resistance descriptors are prioritised. If a single balanced stress-relief condition must be selected for the +3 wt.% Ni alloy, SR500 is the most evidence-consistent compromise within the present dataset.

In the +6 wt.% Ni alloy, the electrochemical ranking also split by descriptor. SR350 gave the strongest breakdown/repassivation descriptors, with $E_{br} = 1150 \text{ mV}$ and $E_{rp} = 1080 \text{ mV}$, whereas SR525 gave the highest resistance descriptors, with $R_p = 192.00 \text{ k}\Omega \text{ cm}^2$ and $R_{ct} = 1251.00 \text{ k}\Omega \text{ cm}^2$ (Tables 4.10 and 4.14). Thus, SR350 is favoured when breakdown and repassivation behaviour are prioritised, whereas SR525 is favoured when impedance/resistance descriptors are prioritised. This divergence is expected because CPP probes passive-film breakdown and repassivation under forced anodic polarisation, whereas EIS describes the small-signal interfacial response near the stabilised open-circuit condition; therefore, CPP- and EIS-derived descriptors should not be collapsed into one overall corrosion ranking. The EBSD evidence sharpens this interpretation: SR525 was not merely the highest-resistance condition, but also the condition with the highest mapped KAM heterogeneity and the lowest HAGB fraction in the +6 wt.% Ni series. This does not prove a causal mechanism, but it prevents oversimplified attribution of its high R_p and R_{ct} to austenite fraction alone.

Overall, heat treatment affected electrochemical behaviour through condition-specific combinations of phase constitution, local chemistry, local orientation heterogeneity, retained LPBF substructure, possible nanoscale/substructural changes, and defect-assisted initiation sites. In duplex stainless steels, passive-film behaviour is controlled not only by nominal Cr and Mo contents, but also by phase-specific partitioning, local Cr/Mo/N depletion or enrichment, oxide-film chemistry, and defects acting as preferential pit-initiation sites. Selected stress-relief treatments improved laboratory electrochemical descriptors relative to AS, but the effect was not universal or monotonic. The divergence among J_{corr} , R_p , R_{ct} , E_{br} , and E_{rp} indicates that passive-film resistance, pit breakdown, repassivation, and interfacial charge transfer cannot be collapsed into one corrosion ranking. Therefore, post-corrosion morphology, pit-depth statistics, critical pitting temperature, and long-term exposure remain necessary for service-level localised-corrosion validation [117, 25, 118, 119].

The corrosion-screening workflow provides the final integration step, but it remains a descriptor-

level framework rather than independent proof of localised corrosion resistance. Because the target variable, $R_{p,app}$, is derived from CPP, it represents an operational polarisation-resistance descriptor rather than a service-life, pit-growth, or passive-film-stability parameter [120, 121, 122]. Agreement between reported and predicted $R_{p,app}$ values ($R^2 = 0.9786$, $MAE = 0.0025$, $RMSE = 0.0040$, and fold-wise $R^2 > 0.95$; Table 4.15) indicates that the model captured internal descriptor relationships. This agreement should not be overstated, because internal performance does not demonstrate transferability. The +3 wt.% Ni records were included in training; their agreement therefore represents training-domain consistency rather than independent external validation.

Feature-sensitivity outputs indicated sensitivity to Ni, Cr, Mo, pitting resistance equivalent number (PREN), heat treatment, processing, environment, and interaction descriptors. This is plausible because Cr, Mo, and N affect passive-film stability and pitting resistance, while Ni influences austenite stability, phase balance, and phase connectivity in duplex stainless steels [118, 119]. However, the CPP/EIS results show that Ni content alone did not control the electrochemical response. Standard SR400, +3 wt.% Ni SR500/SR550, and +6 wt.% Ni SR350/SR525 were favoured by different criteria, while SA1150 in the +6 wt.% Ni alloy diverged from a simple high-Ni trend. Thus, descriptor-level modelling can organise the corrosion-related response only when interpreted together with electrochemical testing and microstructural evidence.

The central outcome of this thesis is therefore not the identification of one universally optimal alloy condition, but the demonstration that LPBF SDSS 2507 requires criterion-specific condition selection. Composition, LPBF-inherited non-equilibrium structure, and post-build heat treatment jointly determine whether a given condition remains ferrite-rich, recovers, becomes austenite-enriched, or approaches phase rebalancing. Consequently, the preferred condition depends on the target criterion, including phase rebalancing, tensile strength, ductility, impact response, breakdown and repassivation behaviour, impedance response, or combined electrochemical-screening response. Table 5.1 synthesises the pathway-based interpretation developed in this chapter. Rather than providing a single ranking, it shows that each alloy–heat-treatment route occupies a distinct response domain. The standard alloy defines the ferrite-rich LPBF baseline and solution-annealing route for duplex phase rebalancing. The +3 wt.% Ni alloy provides the highest-strength pathway after low-temperature stress relief, although its electrochemical response favours different stress-relief conditions depending on whether CPP- or EIS-derived descriptors are prioritised. The +6 wt.% Ni alloy shows that stronger austenite stabilisation can generate near-duplex mapped phase fractions in selected as-built or stress-relieved states, while SR525 and SA1150 shift toward austenite enrichment. These phase-constitution changes are accompanied by condition-specific differences in strength, ductility, impact response, local orientation heterogeneity, and electrochemical descriptors. Therefore, Ni addition and heat treatment should not be interpreted as independently beneficial variables; their value depends on how they modify the coupled phase–defect–strain state and which criterion is prioritised. The table is used as an evidence-integration summary linking microstructural, mechanical, and electrochemical results, rather than as a universal selection rule.

Table 5.1: Criterion-specific condition-selection map for the investigated LPBF SDSS 2507 alloy systems.

Selection criterion	Preferred pathway	Discussion interpretation
<i>Microstructural selection</i>		
Phase rebalancing in standard SDSS 2507	Standard SA1100	Produces an approximately balanced ferrite–austenite mapped phase constitution; the mechanical consequence is captured separately under ductility and impact response.
Near-duplex and austenite-enriched +6 wt.% Ni states	+6 wt.% Ni AS/SR350/SR425; SR525 and SA1150 for stronger austenite enrichment	AS, SR350, and SR425 are near-duplex by mapped EBSD phase fraction; SR525 is austenite-rich; SA1150 is strongly austenite-rich and overshoots balanced duplex phase constitution.
<i>Mechanical selection</i>		
Highest tensile strength	+3 wt.% Ni SR400/SR450	Represents the peak-strength LPBF/stress-relieved pathway in the tested matrix; not a universal optimum because ductility, impact response, and laboratory electrochemical descriptors follow different rankings.
Ductility	Solution-annealed states	Selected by elongation response; strength decreases because solution annealing removes part of the LPBF-inherited strengthening contribution while promoting recovery and austenite enrichment/reformation.
Impact response	Solution-annealed states; selected +6 wt.% Ni AS/SR states	Selected by unnotched Charpy response within the measured range; values reaching the 450 J machine limit should be interpreted as the highest measurable impact response, not as exact toughness ranking.
<i>Electrochemical-response selection</i>		
Standard-alloy electrochemical descriptors	Standard SR400	Ranks highest by the combined CPP/EIS descriptor set within the standard alloy, but should not be generalised as the best condition for phase balance, ductility, or impact response.
+3 wt.% Ni balanced stress-relief pathway	+3 wt.% Ni SR500	Represents the most appropriate compromise when strength retention and improved CPP/EIS descriptors are considered together; SR400/SR450 remain strength-driven choices and SR550 remains resistance-descriptor driven.
+3 wt.% Ni electrochemical descriptors	CPP: SR500/SR550; EIS: SR450/SR550	Selection depends on whether CPP-derived resistance descriptors or fitted EIS resistance is prioritised.
+6 wt.% Ni repassivation behaviour	+6 wt.% Ni SR350	Selected by breakdown and repassivation potentials; this is a laboratory electrochemical descriptor ranking, not proof of service-level localised-corrosion resistance.
+6 wt.% Ni impedance response	+6 wt.% Ni SR525	Selected by fitted charge-transfer resistance and CPP-derived resistance descriptors; interpretation must account for the distinct SR525 EBSD state with high KAM heterogeneity and reduced HAGB fraction.

This synthesis supports that condition selection, rather than Ni content, porosity, heat-treatment temperature, or phase fraction alone, is the evidence-supported framework for interpreting the investigated LPBF-fabricated SDSS 2507 alloys. It also defines the comparison boundary: because the heat-treatment schedules were alloy-specific, cross-alloy trends should be read as integrated alloy–thermal-pathway responses, not direct factorial temperature–composition effects at identical holding times or cooling histories. Within this experimental domain, the pathway-selection map provides the basis for evaluating the working hypotheses and answering the research questions by linking alloy chemistry, LPBF-inherited state, post-build heat treatment, microstructural evolution, mechanical response, and laboratory electrochemical behaviour.

5.1 Hypothesis verification and research-question answers

This section consolidates the chapter-level findings into direct hypothesis-verification statements and research-question answers. The working hypotheses are addressed first because they test the central framework of this thesis: laser powder bed fusion (LPBF)-fabricated super duplex stainless steel (SDSS) 2507 does not respond as a simple nickel (Ni)-content series or as a monotonic heat-treatment-temperature series, but as a pathway-dependent metallurgical system. Against this background, the research questions are answered by identifying what each alloy–heat-treatment pathway achieved, where it remained limited, and how the preferred condition changed with the selected performance criterion.

The verification is deliberately bounded by the investigated LPBF route, powder-mixture chemistry, alloy-specific heat-treatment schedules, specimen orientations, mapped electron backscatter diffraction (EBSD) areas, X-ray diffraction (XRD) measurement conditions, mechanical-test geometry, surface preparation, and laboratory electrochemical protocol. Accordingly, the hypotheses are evaluated as evidence-based relationships within the defined processing–structure–property–corrosion framework, rather than as universal materials-design rules for all LPBF SDSS 2507 systems.

The first working hypothesis proposed that Ni addition would modify the EBSD-mapped LPBF phase constitution by changing the ferrite-to-austenite transformation tendency. This hypothesis is strongly supported for the SDSS 2507 +6 wt.% Ni alloy and partially supported for the SDSS 2507 +3 wt.% Ni alloy. The standard SDSS 2507 and SDSS 2507 +3 wt.% Ni alloys remained strongly ferrite-rich after LPBF and after stress relief, indicating that modest Ni enrichment alone was insufficient to restore a balanced duplex constitution under the investigated route. Conversely, the SDSS 2507 +6 wt.% Ni alloy showed substantially higher EBSD-mapped austenite fractions, demonstrating that stronger Ni enrichment shifted the mapped phase constitution toward austenite. Nevertheless, this shift should be interpreted as a phase-balance tendency within the mapped regions, not as proof of globally homogeneous duplex restoration throughout the full build volume. The second working hypothesis proposed that post-build heat treatment would modify the inherited LPBF phase–defect state. This hypothesis is supported. Stress-relief treatments modified the LPBF-inherited state mainly through changes in retained substructure, local orientation gradients, diffraction-broadening descriptors, boundary-network characteristics, and electrochemical

response. They did not, however, act as duplex-restoration treatments in the standard and +3 wt.% Ni alloys. Consistent with this interpretation, low- to intermediate-temperature exposure provided limited thermodynamic and kinetic opportunity for extensive ferrite-to-austenite rebalancing. By contrast, solution annealing supplied the thermal conditions required for substantial ferrite-to-austenite transformation and chemical redistribution. This produced the strongest phase rebalancing in the standard and +3 wt.% Ni alloys, whereas solution annealing at 1150 °C (SA1150) produced an austenite-rich mapped phase constitution in the +6 wt.% Ni alloy. More precisely, the +6 wt.% Ni SA1150 condition represents an austenite-stabilisation overshoot rather than a successful duplex-restoration condition.

Importantly, the heat-treatment response must remain descriptor-specific. XRD line broadening, EBSD kernel average misorientation (KAM), grain orientation spread (GOS), local orientation-gradient descriptors, and boundary-network metrics are used here as comparative indicators of microstructural evolution. They should not be presented as direct measurements of absolute residual stress, absolute dislocation density, or precipitation state. Similarly, possible ferritic instability after selected stress-relief treatments remains an interpretation supported indirectly by structure, property, and corrosion trends. It should not be stated as unambiguous phase identification unless confirmed by diffraction or nanoscale chemical evidence..

The third working hypothesis proposed that the mechanical and electrochemical responses would reflect the coupled metallurgical state. This hypothesis is supported within the investigated dataset. Mechanical behaviour was associated with the combined effects of phase constitution, LPBF-inherited substructure, orientation heterogeneity, and heat-treatment condition. Ferrite-rich standard and +3 wt.% Ni stress-relieved states retained higher strength, whereas solution-annealed states favoured ductility and unnotched Charpy impact response. In parallel, electrochemical behaviour was condition-dependent, but it did not collapse into a single universal ranking. Rather, the preferred condition changed depending on whether open-circuit potential (OCP), cyclic potentiodynamic polarization (CPP) breakdown/repassivation descriptors, CPP-derived resistance descriptors, electrochemical impedance spectroscopy (EIS)-derived resistance descriptors, or apparent CPP-derived $R_{p,app}$ -based screening was prioritised. The machine-learning-assisted screening was therefore used only as a descriptor-level consistency check against the measured CPP and EIS trends, not as independent validation of service-level localised-corrosion resistance.

Collectively, the three working hypotheses are supported to different extents within the investigated experimental domain. The evidence confirms that the thesis outcomes are not governed by one isolated variable. Instead, Ni level, LPBF-inherited non-equilibrium phase constitution, post-build heat treatment, local orientation state, and electrochemical descriptor selection act together. From this evidence base, the research questions can be answered directly.

The first research question addressed the as-built phase-defect state produced by LPBF. The selected LPBF route produced low-porosity but non-equilibrium material. The standard SDSS 2507 and SDSS 2507 +3 wt.% Ni alloys were strongly ferrite-rich in the as-built (AS) condition, whereas SDSS 2507 +6 wt.% Ni showed a much stronger austenite-stabilisation response. Consequently,

densification and duplex phase balance were decoupled under the investigated processing route: achieving low porosity did not automatically establish a balanced ferrite–austenite constitution. The second research question examined the effect of stress relief and solution annealing on the inherited LPBF state. Stress relief modified the inherited LPBF state in an alloy- and condition-specific manner, mainly through substructure relaxation, orientation-gradient modification, diffraction-broadening changes, boundary-network changes, and electrochemical response. It should therefore not be interpreted as a phase-restoration route. Conversely, solution annealing produced the strongest ferrite-to-austenite rebalancing in the standard and +3 wt.% Ni alloys, although this was accompanied by the expected strength reduction and improved ductility/impact response. In the +6 wt.% Ni alloy, SA1150 shifted the mapped phase constitution beyond balanced duplex and produced an austenite-rich state.

The third research question considered the response of the Ni-modified alloy conditions. The +3 wt.% Ni alloy retained a ferrite-rich AS and stress-relieved state and required solution annealing at 1100 °C (SA1100) for strong austenite enrichment. The +6 wt.% Ni alloy produced much higher mapped austenite fractions already in the AS and stress-relieved states. Nevertheless, this response was not intrinsically beneficial. The +6 wt.% Ni pathway improved austenite stabilisation, but excessive austenite enrichment after SA1150 and condition-specific KAM/boundary-network behaviour show that higher Ni content must be evaluated against the target property, rather than treated as inherently superior.

The fourth research question addressed mechanical response. Mechanical behaviour followed a strength–ductility–impact trade-off. The SDSS 2507 +3 wt.% Ni stress-relieved at 400 °C (SR400) and stress-relieved at 450 °C (SR450) conditions provided the strongest tensile-strength route, consistent with retention of a ferrite-rich, LPBF-inherited strengthened state. In contrast, solution-annealed states and selected +6 wt.% Ni AS or stress-relieved states favoured elongation and unnotched impact response. Thus, the mechanical optimum is contingent upon whether strength retention or damage tolerance is prioritised.

The fifth research question examined the electrochemical response. Electrochemical behaviour was descriptor-specific rather than governed by one universal ranking. In the standard alloy, SR400 was favoured by the combined CPP/EIS descriptor set. In the +3 wt.% Ni alloy, stress-relieved at 500 °C (SR500) and stress-relieved at 550 °C (SR550) were favoured by CPP-derived resistance descriptors, whereas SR450 and SR550 were favoured by fitted EIS resistance descriptors. In the +6 wt.% Ni alloy, SR350 was favoured by breakdown and repassivation descriptors, while SR525 was favoured by resistance-related descriptors. Accordingly, electrochemical ranking depends on the selected criterion: passivity stability, breakdown resistance, repassivation ability, impedance-derived resistance, or screening-level apparent polarization resistance.

The sixth research question asked whether a best alloy–heat-treatment pathway could be identified. The results show that no single pathway was universally best. For tensile strength, the SDSS 2507 +3 wt.% Ni SR400/SR450 pathway was preferred. For phase rebalancing in the standard and +3 wt.% Ni alloys, SA1100 was preferred. For breakdown and repassivation descriptors in the

+6 wt.% Ni alloy, SR350 was preferred. For resistance-related descriptors in the +6 wt.% Ni alloy, SR525 was preferred. Hence, the thesis identifies criterion-specific processing pathways rather than one universal alloy–heat-treatment optimum.

Taken together, the working hypotheses are supported within the thesis domain, but they should not be read as universal or monotonic rules. The main conclusion is that LPBF-fabricated SDSS 2507 behaves as a pathway-dependent metallurgical system. Densification provides a necessary low-defect processing basis, but it does not independently establish duplex phase restoration or corrosion suitability. Stress relief modifies the LPBF-inherited stress/substructure state, whereas duplex restoration requires solution annealing. Excessive Ni enrichment combined with high-temperature solution annealing can overshoot the duplex target and produce an austenite-rich mapped phase constitution. Ultimately, the preferred alloy–heat-treatment pathway is governed by the selected performance criterion rather than by Ni content, heat-treatment temperature, or phase fraction alone.

Chapter 6

Conclusions and Contributions

Building on the pathway-based interpretation developed in Chapter 5, this thesis established an evidence-bounded processing–microstructure–property–electrochemical response framework for laser powder bed fusion (LPBF)-fabricated standard, +3 wt.% Ni, and +6 wt.% Ni super duplex stainless steel (SDSS) 2507. The measured mechanical and laboratory electrochemical responses are better understood through the coupled effects of alloy chemistry, LPBF process history, phase–defect state, orientation heterogeneity, and corrosion-response descriptors, rather than through porosity, nominal Ni content, or heat-treatment temperature alone. In this thesis, the term phase–defect state denotes the combined phase constitution, defect population, residual-strain indicators, and LPBF-inherited microstructural heterogeneity. The main outcome is a criterion-specific pathway-selection approach for linking phase constitution, mechanical response, and laboratory electrochemical descriptors within the investigated experimental domain.

6.1 Major findings

LPBF processing window. The selected LPBF condition, $P = 180 \text{ W}$, $v = 300 \text{ mm s}^{-1}$, $h = 120 \text{ }\mu\text{m}$, and $t = 30 \text{ }\mu\text{m}$, corresponding to $V_{\text{ED}} = 166.7 \text{ J mm}^{-3}$, provided a common low-porosity basis for comparison, but did not independently ensure duplex phase balance.

Phase-balance pathways. Electron backscatter diffraction (EBSD) revealed three alloy-specific phase-balance pathways. The standard and +3 wt.% Ni alloys remained strongly ferrite-rich before solution annealing, whereas the +6 wt.% Ni alloy showed near-duplex mapped phase fractions in the as-built (AS), stress-relieved at 350 °C (SR350), and stress-relieved at 425 °C (SR425) states. Solution annealing produced different outcomes: the standard alloy reached an approximately balanced mapped structure after solution annealing at 1100 °C (SA1100), while the +3 wt.% Ni and +6 wt.% Ni alloys became austenite-rich after SA1100 and solution annealing at 1150 °C (SA1150), respectively. Complementary transmission electron microscopy (TEM) and scanning transmission electron microscopy (STEM) analysis of the selected standard alloy stress-relieved at 550 °C (SR550) condition showed ferritic nanoscale/substructural contrast, but did not confirm sigma, chi, or chromium nitride precipitates in the analysed region.

Non-monotonic stress-relief response. X-ray diffraction (XRD), EBSD boundary analysis, and kernel average misorientation (KAM) mapping showed that stress relief did not follow a universal recovery trend. The standard alloy showed reduced ferrite-reflection-derived peak broadening after stress relief, while the Ni-modified alloys responded non-monotonically.

Mechanical response. The mechanical results showed a strength–ductility–impact trade-off. The SDSS 2507 +3 wt.% Ni SR400/SR450 conditions gave the highest tensile-strength route,

whereas the solution-annealed states gave the highest elongation and unnotched impact-energy response but reduced strength. Selected SDSS 2507 +6 wt.% Ni AS/SR states provided the strongest non-solution-annealed compromise between ductility and impact response. Therefore, the mechanically preferred condition depends on the target property rather than on a single overall ranking.

Electrochemical response. The laboratory electrochemical response was criterion-dependent. Cyclic potentiodynamic polarization (CPP)-derived J_{corr} , R_p , E_{br} , and E_{rp} did not give the same ranking as electrochemical impedance spectroscopy (EIS)-derived R_{ct} . Thus, different alloy-heat-treatment conditions were favoured depending on whether the comparison emphasized corrosion-current response, polarization resistance, breakdown/repassivation behaviour, impedance resistance, or combined descriptor trends. These are laboratory descriptor rankings, not service-level corrosion-resistance claims.

Machine-learning corrosion-screening response The machine-learning workflow showed close internal agreement for the apparent polarization-resistance descriptor, $R_{\text{p,app}}$, with $R^2 = 0.9786$, $\text{MAE} = 0.0025$, and $\text{RMSE} = 0.0040$. Nickel (Ni), chromium (Cr), molybdenum (Mo), pitting resistance equivalent number (PREN), heat-treatment, processing, and environment descriptors influenced the predicted response, while nickel showed a nonlinear effect. This result is treated only as descriptor-level trend analysis. It is not an independent validation of localized-corrosion resistance; open-circuit potential (OCP), CPP and EIS remain the primary electrochemical evidence.

Criterion-specific selection. No investigated alloy-heat-treatment pathway was universally optimal. The results showed clear trade-offs between phase balance, mechanical performance, and laboratory electrochemical descriptors. Therefore, the preferred condition should be selected according to the dominant performance requirement, such as duplex phase balance, tensile strength, ductility, impact response, repassivation behaviour, or EIS-derived resistance.

The criterion-specific selection map is given in Table 6.1. The synthesis is organized as a pathway-selection summary rather than as a universal ranking of the investigated materials. It separates phase constitution, tensile response, ductility, unnotched Charpy impact energy behaviour, CPP-derived descriptors, EIS-derived resistance indicators, and corrosion-screening trends. This separation is necessary because the condition favoured for one response was not necessarily favoured for another. Solution annealing was most favourable for phase rebalancing and ductility/impact response, whereas selected stress-relieved states retained higher strength or gave more favourable laboratory electrochemical descriptors. The comparison therefore shows that pathway selection depends on the dominant engineering requirement and prioritized evidence category, rather than on nickel content, heat-treatment temperature, or a single overall performance score.

Table 6.1: Preferred alloy–condition pathways for the main performance criteria evaluated in this thesis.

Selection criterion	Favoured alloy–condition pathway	Basis for selection
Phase rebalancing	Standard SDSS 2507 SA1100	Closest recovery toward a balanced EBSD-mapped ferrite–austenite phase constitution among the compared pathways.
Tensile strength	SDSS 2507 +3 wt.% Ni SR400/SR450	Highest strength-oriented pathway, associated with retention of a ferrite-rich LPBF-inherited state after stress relief.
Ductility	Solution-annealed states	Favoured elongation response relative to the highest-strength ferrite-rich stress-relieved pathways.
Unnotched impact response	Solution-annealed states and selected +6 wt.% Ni AS/SR conditions	Favoured impact-energy response compared with the strongest stress-relieved conditions.
Breakdown/repassivation descriptors	SDSS 2507 +6 wt.% Ni SR350	Favoured by CPP-derived breakdown and repassivation behaviour.
Resistance-related electrochemical descriptors	SDSS 2507 +6 wt.% Ni SR525	Favoured by resistance-related CPP/EIS descriptor trends.
Balanced electrochemical compromise in +3 wt.% Ni	SDSS 2507 +3 wt.% Ni SR500	Provided the most balanced CPP-based corrosion-screening response within the +3 wt.% Ni set.

Criterion-based engineering recommendation. Based on the pathway-selection summary in Table 6.1, the practical recommendation is not to select the alloy or heat treatment by Ni content or temperature alone. Instead, the pathway should be selected according to the governing engineering criterion within the investigated LPBF route, alloy compositions, heat-treatment schedules, specimen configurations, and electrochemical protocol. Standard SDSS 2507 SA1100 is preferred for phase-balance recovery, SDSS 2507 +3 wt.% Ni SR400/SR450 for maximum tensile strength, solution-annealed states for ductility, and solution-annealed states or selected +6 wt.% Ni AS/SR conditions for impact response. For laboratory electrochemical descriptors, the preferred pathway depends on the selected metric: standard SR400 within the standard alloy, +3 wt.% Ni SR500/SR550 for CPP-based descriptors, +3 wt.% Ni SR450/SR550 for EIS-derived R_{ct} , +6 wt.% Ni SR350 for breakdown/repassivation descriptors, and +6 wt.% Ni SR525 for resistance-related descriptors.

6.2 Scientific and methodological advances

1. Evidence-bounded pathway framework for LPBF-fabricated SDSS 2507.

A pathway-dependent framework is established for interpreting LPBF-fabricated SDSS 2507. The measured response is linked to the coupled effects of alloy chemistry, LPBF thermal history, post-build heat treatment, phase constitution, defect population, local orientation-heterogeneity descriptors, and laboratory electrochemical response.

2. EBSD-resolved comparison of standard and Ni-modified pathways.

An EBSD-resolved comparison is established for phase-balance evolution in standard SDSS 2507, SDSS 2507 +3 wt.% Ni, and SDSS 2507 +6 wt.% Ni. The +3 wt.% Ni alloy did not restore duplex balance before solution annealing, whereas the +6 wt.% Ni alloy produced near-duplex/austenite-rich mapped phase fractions in the AS and stress-relieved states.

3. Clarification of non-monotonic alloying and heat treatment effects.

The results show that neither Ni addition nor heat treatment produced a simple monotonic improvement. The +3 wt.% Ni alloy provided the highest-strength stress-relieved route but remained ferrite-rich before SA1100, whereas the +6 wt.% Ni alloy increased the mapped austenite fraction substantially but became strongly austenite rich after SA1150.

4. Integrated microstructure, mechanical, electrochemical interpretation.

SEM, XRD, EBSD, TEM/STEM, tensile testing, impact testing, CPP, EIS, and descriptor-level modelling are integrated as complementary evidence streams. This links LPBF inheritance, phase constitution, local orientation heterogeneity, mechanical response, fracture morphology, and laboratory electrochemical descriptors without unsupported service-level corrosion claims.

5. Criterion-specific pathway selection and electrochemical screening.

A criterion-specific pathway selection framework is developed for the investigated alloy systems. Preferred conditions differ according to phase balance, tensile strength, elongation, impact response, breakdown/repassivation descriptors, impedance/resistance descriptors, or electrochemical-screening consistency. The $R_{p,app}$ -based workflow is retained only as a dataset-specific trend organisation tool.

6.3 Limitations and future work

The following boundaries of inference define the main directions for extending this thesis toward a more complete qualification framework for LPBF-fabricated SDSS 2507 and its Ni-modified variants. Additional evidence is required before extending the present conclusions to broader process windows, component geometries, surface states, scale-dependent defect populations, and service-relevant environments.

Heat-treatment pathway. The heat-treatment matrix was alloy-specific rather than fully factorial. Therefore, comparisons among the standard SDSS 2507, SDSS 2507 +3 wt.% Ni,

and SDSS 2507 +6 wt.% Ni systems are pathway-based and should not be interpreted as direct factorial comparisons at identical temperatures, holding times, or cooling histories. Future work should use matched thermal schedules to isolate composition–thermal-cycle interactions.

Mapped/descriptive microstructure. EBSD phase fractions, boundary fractions, KAM, GROD, GOS, and XRD-derived broadening quantities are comparative descriptors measured within defined mapped areas or diffraction conditions. They do not establish full-volume phase balance, three-dimensional phase connectivity, absolute residual stress, absolute dislocation density, or precipitation state. Future work should combine larger-area and multi-location EBSD, three-dimensional phase/connectivity analysis, direct residual-stress measurement, and rigorously corrected phase-resolved XRD.

Nanoscale/electrochemical validation. TEM/STEM analysis was limited to selected regions, and the absence of confirmed precipitates in the analysed standard SR550 region does not exclude nanoscale decomposition, clustering, nitrides, or local precipitates elsewhere. Similarly, OCP, CPP, EIS, and the $R_{p,app}$ -based screening model provide laboratory electrochemical descriptors rather than service-level corrosion validation. Future work should include broader nanoscale analysis, post-corrosion morphology, pit-depth and pit-density statistics, critical pitting temperature testing, crevice-corrosion testing, stress-corrosion cracking assessment, and long-term exposure.

Process optimisation. The LPBF parameter selection was based primarily on achieving a low-porosity processing condition. Future work should optimise LPBF parameters for phase balance, defect morphology, residual-stress control, and corrosion-relevant surface states, not only densification. For the +6 wt.% Ni alloy, particular attention should be given to controlling excessive austenite enrichment after solution annealing while retaining the favourable deformation and impact response of selected non-solution-annealed states.

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Appendix A

Data-driven corrosion-screening descriptors and model metadata

A.1 Dataset construction and preprocessing

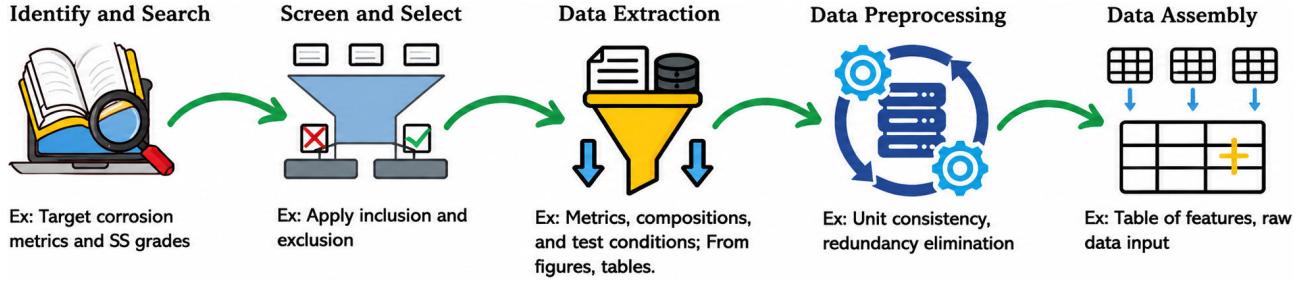


Figure A.1: End-to-end workflow used to build the corrosion-screening dataset for stainless steels under varied thermal and chemical conditions. From left to right, the workflow includes literature identification; screening and eligibility assessment; data extraction from text, tables, and polarization curves; preprocessing, including digitization, unit harmonization, metadata standardization, de-duplication, and quality control; and final dataset assembly into tidy feature tables. Here, E denotes electrode potential in V versus the reference electrode, J denotes current density in A cm^{-2} , and the derived electrochemical response variable is the apparent polarization-resistance descriptor used for model training.

Stainless-steel grades and electrochemical response variables relevant to corrosion-screening model development were identified from the literature. Studies reporting corrosion behavior under different thermal conditions, including heat-treatment temperature, cooling duration, alloy composition, and corrosive environment, were considered. Boolean keyword combinations were used to search databases such as Web of Science, Scopus, and ScienceDirect. Search terms included combinations such as stainless steel OR 2205 DSS and cyclic potentiodynamic polarization OR corrosion rate, combined with electrolyte-related terms such as pH and chloride. Only studies that provided complete Tafel or potentiodynamic polarization data were considered for dataset construction.

Publications were evaluated using two main criteria. First, the study had to include quantitative E and J values, either reported directly or derivable from polarization curves. Second, the thermal history had to be sufficiently documented, including heating temperature and cooling method, as well as compositional information. Studies involving coated samples, corrosion inhibitors, or inadequate post-processing documentation were excluded. Studies lacking defined cooling conditions, such as water quenching, furnace cooling, or air cooling, and studies lacking manufacturing-process information, such as additive manufacturing or conventional processing, were also excluded to preserve feature completeness.

For each selected publication, alloy composition, processing information, and electrochemical variables were extracted. Composition data included chromium, nickel, molybdenum, and other relevant alloying elements. Processing metadata included heating temperature, cooling duration,

manufacturing route, cooling technique, and post-processing condition, such as solution annealing, stress relieving, or tempering. Environmental variables included electrolyte composition, for example, 0.5 M NaCl, acidic soil, 2 M H₂SO₄, and water. The target variable, $\mathcal{R}$, expressed in V cm² A⁻¹, was obtained directly from reported values or derived from polarization data where only graphical information was available. In this appendix, $\mathcal{R}$ denotes the apparent polarization-resistance descriptor used in the model metadata and corresponds to $R_{p,app}$ in Chapter 4. This notation is retained here to preserve consistency with the original dataset and model-development workflow, whereas $R_{p,app}$ is used in the main thesis text for electrochemical clarity.

The extracted data were organized into a structured template, with one row assigned to each corrosion test. All entries were standardized before model development. Temperatures were converted to °C, time was converted to hours, and alloy compositions were expressed in wt.%. Categorical variables, including cooling route and environment type, were normalized using a consistent vocabulary. Records lacking critical information, such as heat-treatment temperature, alloy composition, or the target electrochemical variable, were excluded. Outlier screening was performed on $\mathcal{R}$ values, and extreme cases were checked against the original source before inclusion.

A.2 Dataset variables and target definition

The data-driven corrosion-screening workflow used literature-derived potentiodynamic polarization data for stainless steels and related corrosion-resistant alloys. Each entry contained corrosion-relevant variables describing alloy composition, processing route, heat-treatment condition, cooling condition, electrolyte environment, and apparent polarization resistance.

Apparent polarization resistance was used as the prediction target. It was interpreted as an indicator of electrochemical response rather than a complete measure of localized corrosion resistance. Apparent polarization resistance does not, by itself, define breakdown resistance, repassivation stability, passive-film robustness, or service performance. Model-derived trends were therefore interpreted together with cyclic potentiodynamic polarization, electrochemical impedance spectroscopy, and microstructural characterization.

A.3 Feature blocks

The feature space contained four feature blocks: elemental composition X_{El} , processing variables X_{Pr} , atomic and bulk descriptors X_{AB} , and interaction descriptors X_{Intr} . These feature blocks were used to connect alloy chemistry and processing history with apparent polarization resistance. Table A.1 summarizes their role in the workflow.

A.4 Composition-based corrosion descriptors

The pitting resistance equivalent number (PREN) was included as a bulk-composition descriptor of pitting-resistance tendency:

$$PREN = Cr + 3.3(Mo + 0.5W) + 16N, \quad (A.1)$$

Table A.1: Feature blocks used in the data-driven corrosion-screening workflow.

Block	Role in the feature space
X_{El}	Alloying-element contents relevant to stainless-steel passivity and duplex phase stability.
X_{Pr}	Manufacturing route, post-processing condition, cooling condition, heat-treatment temperature, and holding time.
X_{AB}	Composition-derived atomic and bulk descriptors representing stoichiometry, elemental-property statistics, bonding tendency, and thermodynamic stability.
X_{Intr}	Engineered composition–processing terms, phase-sensitive descriptors, and corrosion-relevant interaction terms.

where Cr, Mo, W, and N are expressed in wt.%. For alloys without tungsten addition, the W term was taken as zero. PREN was used as a descriptor rather than as a direct measure of corrosion performance because it does not account for phase-specific partitioning, residual strain, defects, surface condition, or heat-treatment-induced precipitation.

Chromium-equivalent and nickel-equivalent terms were used to introduce phase-sensitive composition information into the feature space:

$$C_{\text{eq}} = \text{Cr} + \text{Mo} + 1.5\text{Si} + 0.5\text{Nb}, \quad (\text{A.2})$$

$$N_{\text{eq}} = \text{Ni} + 30\text{C} + 0.5\text{Mn} + 0.3\text{Cu}. \quad (\text{A.3})$$

These equivalent terms were used as feature-level estimates of ferrite- and austenite-stabilizing tendencies. They were not treated as substitutes for experimentally measured ferrite/austenite phase fractions.

Estimated phase-tendency terms were calculated as:

$$f_{\gamma} = \frac{N_{\text{eq}}}{N_{\text{eq}} + C_{\text{eq}}}, \quad (\text{A.4})$$

$$f_{\alpha} = 1 - f_{\gamma}. \quad (\text{A.5})$$

The terms (f_{γ}) and (f_{α}) were used only as composition-derived descriptors within the model feature space. They were not interpreted as measured austenite and ferrite fractions.

A.5 Engineered interaction terms

Interaction terms were included to represent nonlinear and coupled composition–processing effects. These terms were used to improve the model’s ability to compare corrosion-relevant feature combinations without treating nominal composition as the only controlling variable.

The main engineered descriptors included bulk passivity descriptors, phase-stability descriptors,

nonlinear nickel terms, composition–processing interactions, and encoded processing/environment variables. Table A.2 summarizes the descriptor groups.

Table A.2: Engineered descriptors used in the data-driven corrosion-screening workflow.

Descriptor	Role in the feature space
PREN	Bulk pitting-resistance tendency.
C_{eq}	Ferrite-stabilizing composition tendency.
N_{eq}	Austenite-stabilizing composition tendency.
f_{γ}, f_{α}	Estimated phase-stability tendencies.
$Ni^2, \log(Ni)$	Nonlinear nickel-sensitive terms.
Ni–Cr	Coupled austenite-stabilization and chromium-passivity term.
PREN–T	Coupled passivity–thermal-exposure term.
T–t	Heat-treatment temperature–time exposure term.
Process	Encoded manufacturing or post-processing route.
Cooling	Encoded cooling condition.
Electrolyte	Encoded corrosion-environment condition.

A.6 Preprocessing and model-training metadata

Categorical variables, including manufacturing route, post-processing condition, cooling route, and electrolyte environment, were encoded before model training. Continuous variables were checked for unit consistency, missing values, and physically unreasonable entries. Potentials, current densities, temperatures, times, and alloy compositions were converted to a common unit basis before descriptor construction. Records lacking the information required to define the target variable or the main corrosion-relevant descriptors were excluded.

The model target was the CPP-derived $R_{p,app}$ descriptor. Records with missing, zero, negative, or physically inconsistent J_{rp} or $R_{p,app}$ were checked against the original source and removed unless a clear transcription, digitization, or unit-conversion error could be corrected. EIS-derived quantities, including R_{ct} , were not used as model inputs or targets.

Continuous input features were scaled within the training workflow. Scaling parameters, categorical encodings, feature construction, and model fitting were generated from the training fold only and then applied to the corresponding validation fold. This fold-specific workflow was used to reduce preprocessing leakage. Controlled perturbation and augmentation were restricted to the measured compositional range.

The predictive model was trained to estimate $R_{p,app}$ from composition, processing, heat-treatment, environment, and engineered descriptors. Cross-validation was used to assess prediction stability within the available dataset. Model performance was evaluated using the coefficient of determination, mean absolute error, and root mean square error. Feature sensitivity was assessed using tree-based feature-importance analysis and permutation scoring, and the resulting rankings were interpreted as model-sensitivity indicators.

A.7 Applicability-domain considerations

The corrosion-screening workflow was interpreted within the represented composition, processing, heat-treatment, environment, and CPP-protocol domain. Literature-derived electrochemical records may differ in surface preparation, exposed area, electrolyte chemistry, reference electrode, scan rate, immersion history, and parameter-extraction method. These protocol differences were treated as part of the applicability-domain boundary.

Model-derived trends were used as descriptor-level screening support and compared with direct experimental evidence from the tested alloy states. The +3 wt.% Ni records provided the closest training-domain reference for the present alloy family. The standard SDSS 2507 and +6 wt.% Ni datasets were used to assess qualitative cross-alloy consistency outside that closest represented composition.