

Controlling microsegregation-induced phase formation and pitting corrosion in wire-arc additively manufactured precipitation-hardened martensitic stainless steel

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ABSTRACT

Wire-arc additive manufacturing (WAAM) often leads to elemental microsegregation and non-equilibrium phase formation during solidification, which can compromise the corrosion performance of alloys. This study presents the first detailed investigation of WAAM processed Custom 465® (C465), a precipitation-hardened martensitic stainless steel, focusing on the effect of post-printing heat treatment strategies on the microstructure and its pitting corrosion behaviour, benchmarked against the conventionally manufactured wrought C465 material. As-printed C465 exhibited large columnar prior austenite grains (PAGs) and pronounced microsegregation, which stabilised the retained austenite and Laves phases. The standard solution treatment (SST) refined the PAGs but was insufficient to eliminate microsegregation. Guided by diffusion-controlled transformation (DICTRA) simulations, a suitable homogenisation treatment was developed that effectively reduced WAAM-induced microsegregation and associated secondary phases. Ageing treatment after the SST only partially promoted martensite-to-austenite reversion, likely driven by pre-existing retained austenite and Ni-segregation. In contrast, ageing treatment after homogenisation and SST largely suppressed this transformation, resulting in a microstructure comparable to the wrought counterpart. The corrosion performance of C465 produced by both WAAM and conventional routes was further evaluated using cyclic potentiodynamic polarisation and electrochemical noise techniques. With the standard heat treatment, the WAAM material exhibited inferior pitting resistance compared to its wrought counterpart. Surface characterisation following anodic polarisation revealed pit initiation at Cr-depleted interdendritic retained austenite regions and Laves/matrix interfaces in the WAAM material. However, homogenisation prior to SST and ageing treatment significantly enhanced the pitting resistance by eliminating preferential pit nucleation sites, thereby achieving a performance comparable to the wrought counterpart. Overall, this work demonstrates the critical role of post-printing heat treatment strategies in addressing WAAM-induced microstructural heterogeneities and mitigating pitting susceptibilities.

1. Introduction

Additive manufacturing (AM) has recently emerged as a promising route for fabricating ultra-high strength martensitic steel components for critical aerospace applications [1]. By enabling near-net-shape geometries with minimal machining and material wastage, AM offers a more sustainable and efficient alternative to conventional manufacturing (CM) routes [2]. Among the various AM techniques, wire-arc additive manufacturing (WAAM) has particularly garnered

interest due to its ability to produce large-scale metallic components with high deposition rates (typically ~ 1–10 kg/h) [3]. WAAM is a direct-energy deposition technique that employs an electric or plasma arc as the heat source and a wire feedstock to build components layer-by-layer on a substrate, following principles similar to the traditional arc-welding process. Compared to powder-based AM techniques, WAAM also offers higher feedstock utilisation and eliminates the need for powder handling and recycling, thereby reducing both operational costs and health and safety concerns [3].

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Over the past decades, the critical components of aircraft landing gear have relied on ultra-high strength martensitic steels such as AISI 4340 and 300 M due to their excellent combination of tensile strength, fatigue resistance and fracture toughness [4,5]. Despite their exceptional mechanical properties, these low-alloyed martensitic steels suffer from inherently poor corrosion resistance, necessitating the use of toxic cadmium surface coatings to ensure long-term durability in demanding service environments [6,7]. To address the need for materials that can provide similar mechanical performance, with better corrosion resistance, Carpenter Technology Corporation developed Custom 465® (C465), a precipitation-hardened martensitic stainless steel (PHMSS) which contains high Cr (~12 wt%) and Ni (~11 wt%) [8]. C465 is reported to have an excellent combination of tensile strength, fracture toughness and corrosion resistance as compared to 300 M martensitic steel [7].

The conventional manufacturing route and subsequent heat treatments for C465 are now well established for aerospace applications [7]. In the CM condition, C465 typically has a fully martensitic structure with homogeneous distribution of key alloying elements, owing to the optimised initial thermo-mechanical processing. According to Aerospace Material Specifications (AMS) 5936 C [9] and Carpenter Technology data sheet [10], the recommended standard solution treatment (SST) involves solution annealing (at $982 \pm 8^\circ\text{C}$) to form a homogeneous austenite structure, and subsequently deep cryogenic cooling (at -73°C) to promote complete austenite to martensitic transformation. In the final step, an ageing treatment is performed to precipitate strengthening η -Ni₃Ti particles and control the formation of reverted austenite, both of which are widely reported to significantly influence the alloy's mechanical and corrosion performance [11–15]. However, the complex thermal cycles, relatively higher cooling rates (compared to CM), and non-equilibrium solidification conditions inherent to the WAAM process often result in coarse columnar grains, with pronounced elemental segregation during solidification and may also form phases that are not seen in a fully processed wrought material [16,17]. For C465, such microstructural features may significantly alter the alloy's response to SST and adversely affect the performance of the WAAM fabricated alloy, unless they are successfully removed by post-processing treatments.

To date, no studies have reported in the open literature on the fabrication of C465 by WAAM. In addition, very limited studies have explored its processing via powder bed fusion by laser (PBF-L), which involves much higher solidification rates [18,19]. For example, Sun et al. [19] have reported hot cracking in C465 during PBF-L processing and attributed it to retained austenite formation and selective partitioning of Ti during solidification. In another study, Liu et al. [18] have demonstrated that PBF-L processed C465 after post-processing treatments, involving hot isostatic pressing (HIP) exhibited a combination of strength (UTS = 1761 MPa, YS = 1691 MPa) and ductility (6%) comparable to wrought C465 (UTS = 1780 MPa, YS = 1684 MPa and ductility = 9%). Unlike the PBF-L process, which is characterised by extremely high cooling rates and rapid thermal cycles, due to its fast-moving, highly focused heat source, WAAM involves larger melt pools, and slower cooling rates owing to its higher heat input and lower translation rates [20]. These distinct solidification dynamics are expected to present unique challenges for WAAM-fabricated C465 that remain unexplored, highlighting a critical research gap.

The more heterogeneous and complex microstructural features induced by WAAM process can readily influence the corrosion performance of alloys. Some microstructural features inherent to AM processing, such as elemental segregation, and non-equilibrium phases formed during solidification, have been shown to reduce pitting corrosion resistance [21–23]. For instance, in PBF-L 316 L stainless steel, depletion of passivating elements (Cr and Mo) at melt pool boundaries reduced repassivation ability and promoted pit stability [24]. MnS inclusions have long been recognised as preferential sites for pitting initiation in stainless steels [21,23]. However, some studies on AM

fabricated stainless steels indicate that when these inclusions are refined to nanoscale dimensions (100–200 nm) by rapid solidification, their influence on pitting nucleation becomes negligible [25,26]. In addition, previous studies have shown that post-printing heat treatments can significantly improve the corrosion resistance of martensitic stainless steels produced by AM [27,28]. For instance, Nemani et al. [27] investigated the corrosion performance of WAAM-fabricated 13–8Mo PHMSS and reported that solution treatment at 1050°C could both eliminate the columnar structure and remove residual δ -ferrite, significantly enhanced pitting corrosion resistance compared to in the as-printed condition. Similarly, Stoudt et al. [28] demonstrated enhanced pitting resistance in AM 17–4 PHMSS after solution annealing, which was attributed to microstructural homogenisation and the elimination of elemental segregation. However, the effect of WAAM processing and post-printing heat treatment strategies on the corrosion performance of C465 is expected to differ due to its distinct chemical composition and solidification characteristics, underscoring the need for further research. In particular, the localised pitting corrosion behaviour is of critical interest because pitting often acts as the precursor to stress corrosion cracking (SCC) in chloride-containing environments [29]. As SCC failures can be catastrophic and difficult to predict, understanding the factors governing pit initiation and stability provides an essential early indication of the material's susceptibility to more severe degradation modes.

While general heat treatment optimisation has been widely reported across various additively manufactured alloys, these studies have overwhelmingly emphasised mechanical performance. In contrast, very limited work has examined corrosion behaviour, particularly its mechanistic connection to WAAM specific solidification features. Moreover, for C465 specifically, no systematic comparison between WAAM and wrought material exists in the literature, especially regarding microstructural evolution, heat treatment response, and the resulting pitting corrosion behaviour. This work therefore provides the first comprehensive process–structure–corrosion relationship for WAAM C465. A comprehensive evaluation has been carried out to assess the effect of post-printing heat treatments on the solidified WAAM microstructure and its pitting corrosion behaviour, benchmarked against the CM wrought C465. Guided by CALPHAD diffusion-controlled transformation (DICTRA) simulations [30], a suitable homogenisation treatment was developed to mitigate WAAM-induced microsegregation and associated non-equilibrium phases and achieve a microstructure comparable to that of the wrought counterpart. To compare the localised corrosion behaviour of WAAM produced and wrought C465 material, a cyclic potentiodynamic polarisation (CPP) technique was employed. Metastable pitting events, which act as precursor to stable pit growth, are characterised by rapid initiation and repassivation events. Although, potentiostatic methods are commonly used to monitor these events, they can introduce artefacts due to externally applied potentials, leading to biased interpretations [31,32]. To address this limitation, the electrochemical noise (ECN) technique was employed to capture spontaneous metastable pitting events under free corrosion conditions more closely replicating real-service environments [31,33]. High-resolution surface characterisation after anodic polarisation was performed to explore the pit nucleation sites in both WAAM produced and wrought C465 materials.

Collectively, this work advances the additive manufacturing field by (i) establishing a comprehensive process–structure–corrosion relationship for WAAM material, (ii) demonstrating whether conventional heat-treatment protocols designed for wrought material are directly transferable to WAAM produced material, and (iii) provides a modelling-based post processing strategy to mitigate WAAM-induced pitting susceptibility and improve corrosion performance.

2. Methodology

2.1. WAAM deposition

A single-pass wide WAAM C465 straight wall sample with overall dimensions of approximately 60 mm (height) × 15 mm (thickness) was deposited on a mild steel substrate using plasma transferred arc (PTA) heat source. An external wire feeder was used to feed the PHMSS wire into the melt pool. The motion of the torch was controlled by a 6-axis robot. The interlayer temperature during WAAM deposition was monitored using a handheld industrial probe thermometer applied directly to the surface of the newly deposited layer. Deposition of each subsequent layer was resumed once the surface temperature of the preceding layer had decreased to approximately 150°C. This approach maintained the interlayer temperature within a narrow range, ensuring consistent thermal conditions throughout the build. The processing parameters used during the WAAM deposition are provided in Table 1. While the full exploration of the WAAM process window for C465 is beyond the scope of the current study, the single set of parameters provided in Table 1 lie within a stable operating range that ensured reliable material deposition. The wall reference axes are defined as Z (build direction, BD), Y (normal direction, ND) and X (transverse direction, TD) as shown later in Fig. 3(a). The chemical composition of the WAAM C465 material is listed in Table 2. The elemental concentrations were determined using a combination of inert gas fusion (for nitrogen and oxygen), inductively coupled plasma (ICP) spectroscopy (for metallic elements) and combustion analysis (for carbon and sulphur).

2.2. Thermodynamic simulations

Thermodynamic simulations were performed using Thermo-Calc software® (version 2022a), with thermodynamic database TCFe10 and mobility database MOBF5. The Scheil-Gulliver model was employed to predict the non-equilibrium solidification path during the WAAM process [34]. To predict homogenisation behaviour across various thermal conditions, kinetic simulations were performed using the DICTRA model in the Thermo-Calc software package [30], using a segregation profile measured after deposition by energy dispersive X-ray spectroscopy (EDS) technique (see Section 3.5) as a starting condition. DICTRA simulations were automated using the TC-Python application programming interface (API), which enabled batch execution of multiple thermal conditions and efficient extraction of composition profiles for homogenisation analysis.

2.3. Heat treatments

For heat treatments, sample coupons were sectioned longitudinally from the as-printed wall, parallel to the BD. A 50 mm diameter wrought C465 bar, received from Carpenter Technology [8], was also investigated to compare the heat treatment response with WAAM C465. All the heat treatments were performed using a Carbolite tube furnace under a controlled argon atmosphere. An N-type thermocouple wire attached to

a thermometer was placed near the sample to ensure that the sample was exposed to the target temperatures. The heat treatments applied are detailed below and their shorthand designations are summarised in Fig. 1 and Table 3, respectively.

1. Standard Solution Treatment (SST): The standard solution treatment process for wrought products, as recommended in AMS 5936 C [9] and Carpenter Technology C465 datasheet [10], was applied to WAAM deposited sample. This involved solution annealing at 982°C for 1 h and water quenching, followed by cryogenic treatment at −73°C for 8 h (within a 24 h duration after solution annealing).
2. Homogenisation (H): As-printed WAAM samples were also subjected to homogenisation trials prior to SST, based on DICTRA simulation results. These were conducted at 1100°C, 1150°C and 1200°C for 4 h, and air cooled. Following microstructure characterisation, the 1200°C for 4 h homogenisation condition was selected and applied to WAAM samples.
3. Ageing H950 (A): Following SST and H + SST treatments of the WAAM samples, ageing treatment was performed at 510°C for 4 h followed by air cooling [7]. Additionally, a wrought C465 bar baseline (received in the SST condition) was subjected to the same ageing treatment.

Three material conditions were selected for detailed electrochemical testing, to reflect the main possible routes to obtaining full in-service corrosion performance of C465. These samples were also subjected to detailed microstructure characterisation.

- i. WAAM_SST + A: The WAAM material subjected to the standard heat treatment normally applied to wrought products (SST + A).
- ii. WAAM_H + SST + A: The WAAM material subjected to a suitable first homogenisation step before the (SST + A) treatment.
- iii. WR_SST + A: A wrought bar baseline (received in the SST condition) subjected to the same ageing treatment.

2.4. Microstructure characterisation

For metallographic preparation, the samples were ground using silicon carbide papers of reducing grit size down to P4000, followed by sequential polishing with diamond suspension (3 µm, 1 µm and 0.25 µm). A final polishing step was performed using colloidal silica with a 0.05 µm particle size. Secondary electron (SE) imaging, back-scattered electron imaging (BSE), EDS and electron backscattered diffraction (EBSD) analyses were acquired using a Thermo Fisher Apreo C, field emission gun scanning electron microscopy (SEM), equipped with an Oxford Instruments' Symmetry 2 EBSD detector and Ultim Max 170 detector. EBSD acquisition was performed using an accelerating voltage of 20 kV and a beam current of 51 nA with a step size of 0.02 — 0.5 µm, depending on the region of interest. Large area maps were montaged and stitched using Aztec 6.1 SP2 software and PAGs were reconstructed using the Aztec Crystal 3 parent grain reconstruction tool with the Kurdjumov-Sachs orientation relationship. The EBSD maps are shown in inverse pole figure (IPF) colouring with the orientation reference parallel to Z. The grain size was measured using the area weighted average equivalent circle diameter method. For the EDS analysis, beam conditions of 10 kV and 13 nA were used and the TruMap tool in Aztec software was used to remove background noise from the maps.

2.5. Electrochemical tests

Electrochemical tests were performed using a potentiostat. To ensure comparable surface conditions, all specimens were ground using silicon carbide papers up to P4000. All tests were conducted using samples with an exposed area of 1 cm². CPP measurements were performed in a conventional three-electrode cell, in an aerated 0.6 M NaCl solution

Table 1
WAAM processing parameters.

Process parameters	Value
Current	210 A
Wire feed speed	2.35 m/min
Travel speed	3.3 mm/s
Wire diameter	1.2 mm
Deposition strategy	Back-and-forth
Total layers	52 layers
Interlayer temperature	~150°C
Contact to work distance	8 mm
Torch gas flow rate	0.8 L/min
Local shielding gas flow rate	80 L/min

Table 2
Chemical composition of the WAAM C465 material.

Elements	Cr	Ni	Ti	Mo	Si	Al	Co	Cu
Content (in wt%)	11.4	10.96	1.55	0.95	0.08	0.04	0.02	0.02
Elements	Mn	Ta, Nb, V, Tn	P	C	S	O	N	H
Content (in wt%)	0.01	< 0.01	< 0.005	0.009	< 0.003	0.0039	< 0.003	0.0002

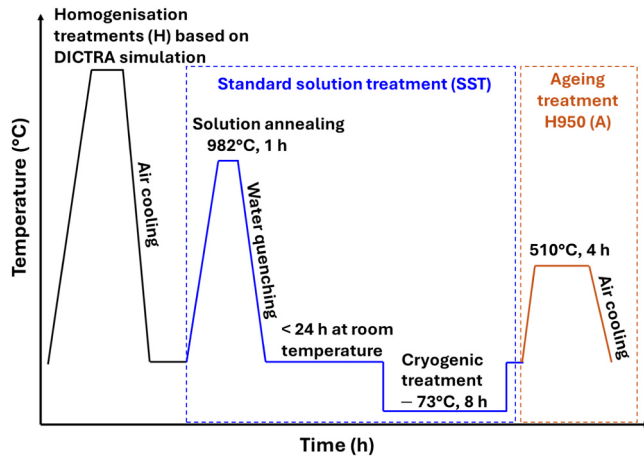


Fig. 1. Heat treatment process diagram used for WAAM produced and wrought C465 materials.

Table 3
Sample nomenclature used in this work.

Manufacturing method	Heat treatment process	Sample nomenclature
Wrought	SST	WR_SST
WAAM	SST	WAAM_SST
WAAM	H	WAAM_H
Wrought	SST + A	WR_SST + A
WAAM	SST + A	WAAM_SST + A
WAAM	H + SST + A	WAAM_H + SST + A

prepared using de-ionised water. Under these conditions, dissolved oxygen (approximately 8 ppm) typically increases the open circuit potential (OCP) measured with respect to an oxygen-free environment. However, the pitting potential (E_{pit}), would not be altered since it depends on the microstructure and chloride concentration [35–37]. Although the dissolved oxygen content was not continuously monitored, the temperature and solution composition were kept constant across all experiments, ensuring consistent and comparable electrochemical conditions. The C465 specimens were used as a working electrode (WE), while Ag/AgCl and a platinum electrode were used as reference (RE) and counter (CE) electrodes, respectively. The OCP of all specimens was monitored for 1 h prior to polarisation measurements. This stabilisation period ensured that any residual potential drift was negligible relative to the applied potential during the subsequent electrochemical tests. Subsequently, CPP scans were conducted at a scanning rate of 0.5 mVs^{-1} , started 20 mV below the OCP and the scan direction was reversed when the measured current reached $\sim 0.5 \text{ mA}$. The CPP tests were performed to assess the pitting corrosion behaviour of WAAM produced and wrought C465 materials under applied potential. From this test, two key electrochemical parameters were recorded: the pitting potential (E_{pit}), above which stable pit formation occurs due to the breakdown of the passive film, and the repassivation potential (E_{rep}), below which pit growth ceases and the passive film begins to reform [32]. Since pitting initiation and growth are stochastic processes, the measured electrochemical parameters can exhibit significant scatter [32]. Therefore,

each CPP test was repeated three times on freshly repolished specimens to ensure the reproducibility and the average values and standard deviations of E_{pit} and E_{rep} were reported.

Further, metastable pitting was assessed using the ECN technique under open-circuit conditions to avoid artefacts commonly associated with potentiostatic methods, where externally applied potential can lead to biased interpretations [31,38]. ECN tests were performed by coupling two specimens at their OCP connected by a zero-resistance ammeter (ZRA), with potential fluctuations recorded against an Ag/AgCl RE. Each condition was evaluated in a pairwise manner, with one specimen designated as the WE and the other as the ground (G), enabling a direct comparison of their metastable pitting behaviour and relative susceptibility to localised corrosion. ECN records were acquired with an acquisition rate of 10 Hz in a $4.6 \text{ mM FeCl}_3 + \text{NaCl}$ solution to reach a total chloride ion concentration of 0.6 M. The addition of FeCl_3 , acting as an oxidising agent, was used to raise the OCP and allow the metastable pitting region to be reached within a shorter experimental time than in NaCl alone. Importantly, the presence of Fe^{3+} does not influence the pitting potential, which is governed solely by the chloride concentration [35–37]. Because the chloride concentration in the solution was identical to that used in the CPP tests, the pitting potential remains comparable. ECN was primarily used as a relative ranking technique to compare the frequency of metastable pitting events between samples with different microstructures under identical conditions. A Python program was subsequently used to quantify the number of metastable events. By setting positive and negative current thresholds, the program determined the number of positive and negative peaks within a given time window increment. The program scanned the whole transient within the time window, with an overlap of 50% with the previous increment. The reported data were obtained by selecting a 1000 s window length and $\pm 1 \mu\text{A}$ as a threshold for the positive and negative current events. It should be noted that the current events below these thresholds may have been missed; however, their occurrence was negligible.

2.6. Localised corrosion characterisation

To identify the preferential sites for pit initiation in WAAM and wrought C465 materials, the WAAM_SST + A and WR_SST + A samples were polished using colloidal silica suspension. Subsequently, the samples were anodically polarised at +1 V (vs Ag/AgCl) for 2 s in a 0.6 M NaCl solution, to trigger pit initiation while suppressing further propagation [39,40]. For the post corrosion analyses, SEM-EDS analyses were performed using a Zeiss Merlin Field Emission Gun (FEG) attached with an Oxford Instruments X-Max Extreme 100 windowless SDD at low accelerating voltage of 5 kV and a beam current of 5 nA to maximise resolution.

3. Results

3.1. Thermodynamic modelling of solidification and segregation in WAAM C465

To support interpretations of the WAAM solidification microstructures, the Scheil-Gulliver model [34] was employed to predict the non-equilibrium solidification path and extent of microsegregation

during the WAAM process. The model assumes infinitely fast diffusion in the liquid phase and negligible diffusion in the solid phase for the metallic elements present, which is a reasonable assumption for arc welding comparable solidification rates. The predicted non-equilibrium solidification path, with C and N entered as fast diffusers, is shown in Fig. 2(a). Solidification primarily begins with the formation of austenite phase, along with a small amount of corundum and Ti-rich carbides, oxides and nitrides. As the solid fraction increases to about 0.45, the model predicted the formation of δ -ferrite phase. However, δ -ferrite phase was not observed in the as-printed WAAM C465 microstructure, as shown later. The equilibrium phase fraction diagram predicted for C465 (Fig. 2(b)) showed a small fraction (~ 0.12) of δ -ferrite phase forming at higher temperature. However, the high solidification and cooling rates during WAAM process are likely to suppress the formation of δ -ferrite, as more rapid advancement of the solid-liquid interface promotes competitive selection of austenite due to its higher dendrite tip temperature [41]. Towards the final stage of solidification, the Laves phase was predicted to form at $\sim 1230^\circ\text{C}$ under non-equilibrium conditions, whereas the equilibrium phase diagram predicted its formation at $\sim 1050^\circ\text{C}$. This discrepancy arises from the fundamental difference in diffusion assumptions between the Scheil-Gulliver model and the equilibrium model. The Scheil model [34], which neglects solid-state diffusion, simulates rapid solidification conditions where solute elements such as Ti, Ni, and Mo enrich the remaining liquid, leading to localised solute enrichment in the inter-dendritic regions. This solute build-up can result in the eutectic solidification of Laves phase at higher temperature. Conversely, the equilibrium model assumes complete diffusion in both solid and liquid phases, resulting in delayed Laves formation at lower temperature.

3.2. As-printed WAAM C465 microstructure

Fig. 3 presents the microstructure of the as-printed WAAM C465 material. A cross-sectional view along the Z-X plane is shown in Fig. 3(a) and a large-area montaged EBSD map of the PAGs, reconstructed from the highlighted region in Fig. 3(a), is provided in Fig. 3(b). The WAAM microstructure was characterised by elongated columnar PAGs, extending up to millimetres in length and aligned close to the Z direction, with slightly tilted towards the Y direction. This grain morphology is typical of the WAAM process and results from epitaxial grain growth during solidification [42,43]. Low-magnification SEM-EDS elemental maps of the as-printed structure are shown in Fig. 3(c). The EDS analysis

revealed pronounced microsegregation of Ti, Ni, Mo, Fe, and Cr between the dendritic and inter-dendritic regions, demonstrating significant chemical heterogeneity. Fig. 4 presents a detailed microstructural and compositional characterisation of the WAAM solidified microstructure. Higher magnification BSE micrograph (Fig. 4(a)) revealed strong atomic contrast, indicating significant solute segregation within the inter-dendritic regions during solidification. These regions appeared darker than the surrounding matrix and were populated with discrete micron-sized secondary particles of distinct morphologies and contrast, some of which were also distributed in the dendritic regions.

A high-resolution EBSD scan shown in Fig. 4(b) was performed on the marked region (white box) in Fig. 4(a). The EBSD phase map indicated that the inter-dendritic regions predominantly consisted of retained austenite. Within these regions brittle intermetallic hexagonal Laves particles were also observed, reaching sizes up to $\sim 4\ \mu\text{m}$. The EBSD analyses also indicated the presence of TiN particles distributed in the dendritic and inter-dendritic regions. To confirm the elemental composition of these phases, EDS analysis shown in Fig. 4(c) was performed on the exact location corresponding to the BSE image in Fig. 4(a). The EDS results revealed that the retained austenite in the inter-dendritic regions was mainly enriched in Ni and Ti. The Laves phase within the retained austenite region exhibited selective enrichment in Ti, Mo and Si and depletion of Fe, Ni, Cr and Mn. Some Mo/Ti-rich needle-like particles were also observed within the inter-dendritic regions but could not be indexed in the EBSD map. The EDS analysis revealed that the TiN particles also contained traces of oxygen, indicating the possible presence of Ti-rich oxides and nitrides in the solidified WAAM microstructure, consistent with the Scheil-Gulliver solidification simulation in Fig. 2(a) and reported in PBF-L fabricated C465 martensitic stainless steel [18].

3.3. Effect of standard solution treatment on WAAM C465

The microstructure of the as-printed WAAM C465 after standard AMS 5936 C solution treatment (WAAM_SST) is shown in Fig. 5. The EBSD IPF map of the reconstructed PAGs revealed a near equiaxed grain structure with an average PAG size of $62 \pm 22\ \mu\text{m}$ (Fig. 5(a)), indicating that the SST significantly refined the PAG size compared to the coarse columnar PAGs observed in the as-printed condition (Fig. 3(b)). The EBSD phase map demonstrated that retained austenite still persisted in the inter-dendritic regions, but its volume fraction reduced from $\sim 8.4\%$ in the as-printed condition (refer Fig. S1 in supplementary file) to $\sim 3\%$

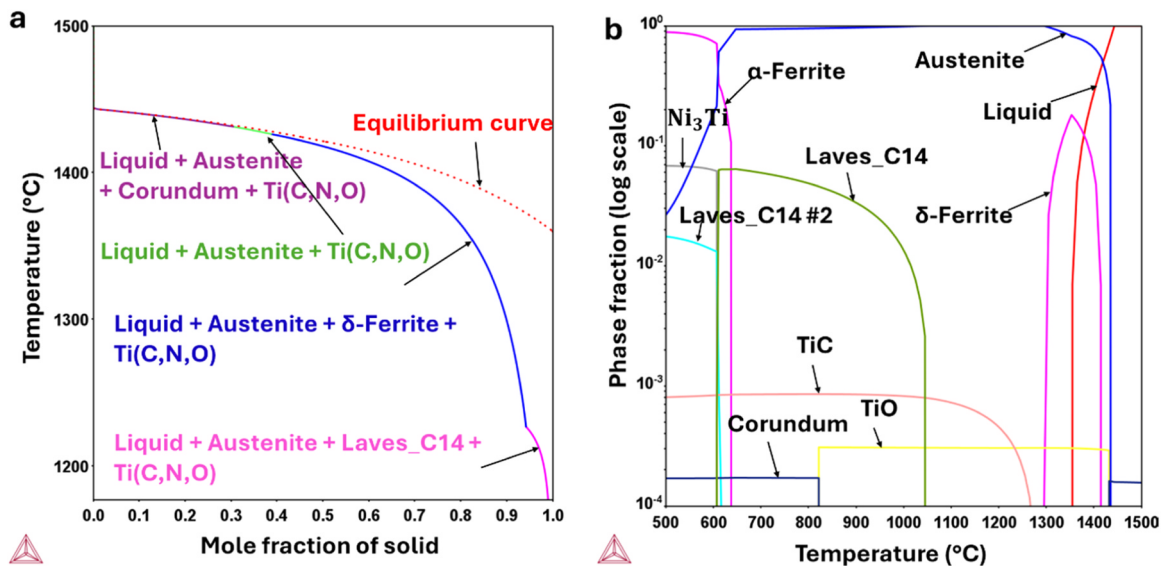


Fig. 2. (a) Non-equilibrium solidification path predicted by Scheil-Gulliver model and (b) equilibrium phase fraction diagram of WAAM C465 material. The chemical composition for the thermodynamic simulations was simplified to only contain Fe, Cr, Ti, Mo, Si, Al, C, O and N, as displayed in Table 2.

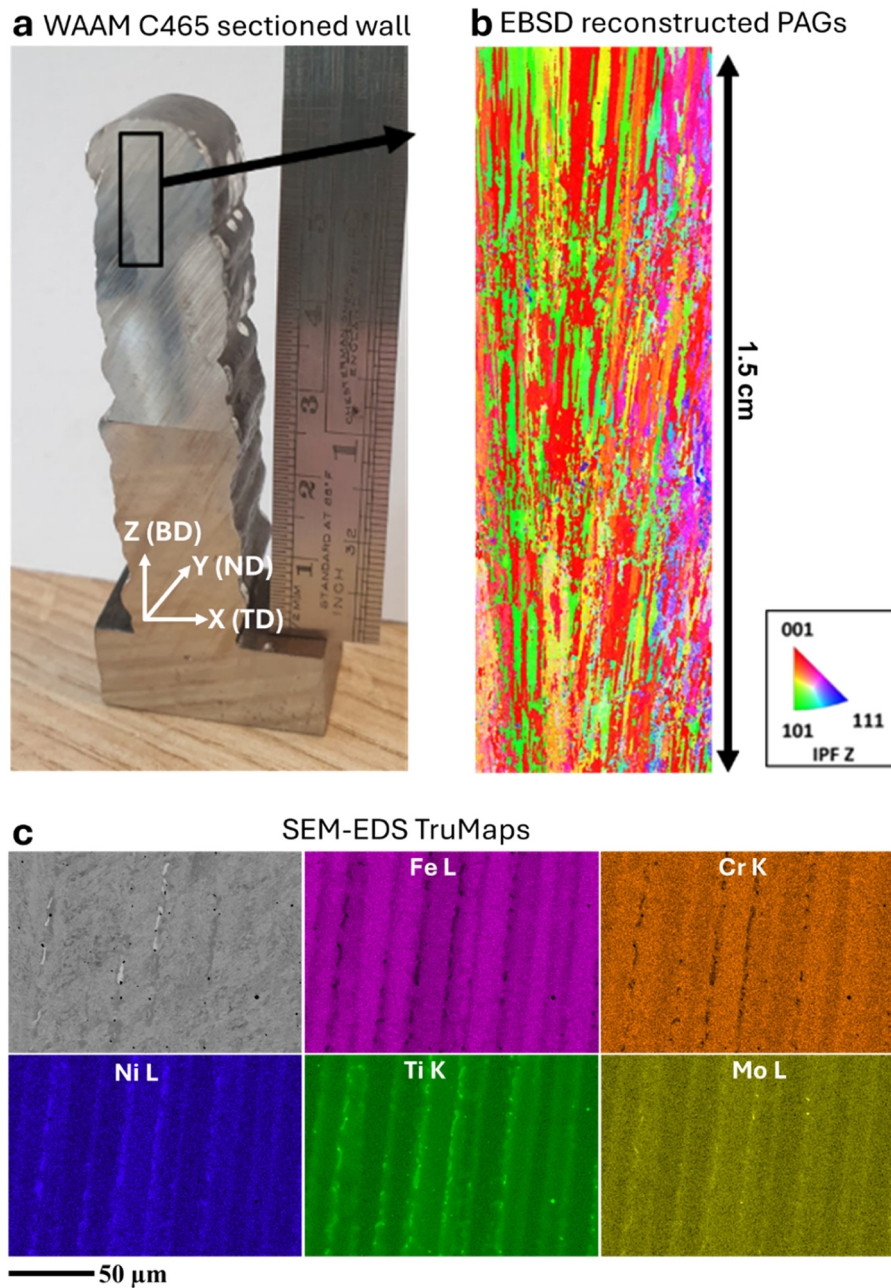


Fig. 3. Microstructure of the as-printed WAAM C465. (a) Cross-section view of the WAAM wall along the Z-X plane, (b) Montaged EBSD IPF map of the reconstructed PAGs from the marked region in (a). (c) SEM-EDS elemental TruMaps highlighting the chemical inhomogeneity across the dendritic and inter-dendritic regions.

after SST (Fig. 5(b)). Elemental analysis (Fig. 5(c)) showed considerable amounts of Ni and Ti segregation in the inter-dendritic regions contributing to the stabilisation of the retained austenite and Ti-rich Laves particles. For comparison, the microstructure of wrought C465 bar subjected to the same SST (WR_SST) was also investigated, as shown in Fig. 6. The wrought WR_SST condition revealed an equiaxed grain structure with an average PAG size of $54 \pm 16 \mu\text{m}$ (Fig. 6(a)), which was comparable to the WAAM_SST condition (Fig. 5(a)). Unlike the WAAM_SST condition, the wrought WR_SST condition contained very low retained austenite fraction ($\sim 0.2\%$), as seen in Fig. 6(b). The EDS analysis also revealed a much more homogeneous distribution of the key alloying elements in the wrought WR_SST condition (Fig. 6(c)), in addition to a few coarser TiN inclusions, which are typical of wrought C465 material [14].

3.4. Predicting homogenisation behaviour using DICTRA simulations

To design a homogenisation heat treatment for the WAAM C465, kinetic simulations using a DICTRA model were performed [30]. To enable this, the simulation cell was treated as a single-phase FCC (austenite) system, rather than a two-phase moving boundary model (FCC + Laves), due to the unavailability of a commercial mobility database for the Laves phase. The dissolution behaviour of the Laves phase, confined to the retained-austenitic regions, was approximated through the redistribution of its constituent elements (mainly Ti and Mo) from the inter-dendritic regions into the surrounding FCC matrix, an approach expected to have a negligible impact on the predicted homogenisation behaviour due to the very low Laves fraction. Key DICTRA simulation inputs were directly extracted from the SEM-EDS line scan that captured the most severe segregation profile across multiple dendritic and inter-dendritic regions, as shown in Fig. 7. These included the

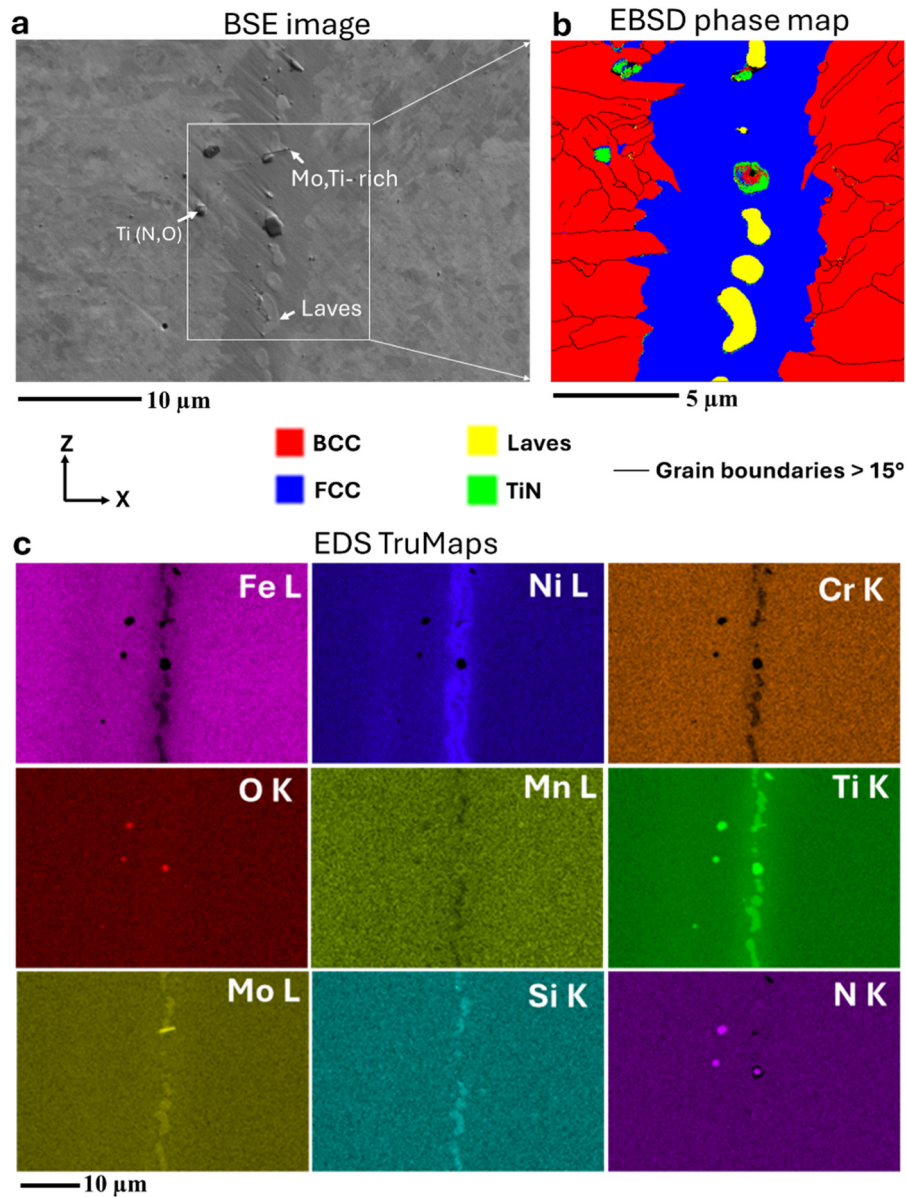


Fig. 4. High-resolution characterisation of the WAAM C465 solidified microstructure. (a) BSE image showing secondary particles within the inter-dendritic regions. (b) High-resolution EBSD phase map, collected using a step size of 25 nm, from the marked region in (a). Phase colour codes are shown alongside. Black lines indicate high-angle grain boundaries with misorientation angles $> 15^\circ$. (c) EDS TruMaps corresponding to the BSE image in (a), highlighting the chemical composition of the secondary phases and oxides.

total width of the simulation cell (418 μm), corresponding to the full length of the EDS line scan crossing approximately 20 dendritic arms, the number and spacing of each grid point (800 points spaced at 0.5 μm intervals), and the local composition at each point. This EDS step size was chosen in order to resolve microsegregated inter-dendritic regions that were approximately 2–8 μm wide, ensuring reliable representation of compositional variations in the simulation domain. The DICTRA simulations were performed at 982°C for 1 h (based on the SST condition, Fig. 1), as well as at elevated temperatures of 1100°C, 1150°C and 1200°C for time intervals of 4 h. The predicted profiles of the key alloying elements (Cr, Ti, Ni and Mo) as a function of temperature and time are shown in Fig. 8. Under the SST condition (982°C for 1 h), the predicted compositional profiles across the dendritic and inter-dendritic regions remained largely unchanged (Fig. 8(a)) compared to the as-printed WAAM condition (Fig. 7). This simulation result was consistent with the experimental SEM-EDS analysis on the WAAM microstructure after SST (WAAM_SST, Fig. 5(c)), thus further confirming the

limited effectiveness of SST in achieving homogenisation in the WAAM C465. Simulations at 1100°C and 1150°C for 4 h resulted in a gradual reduction in the Ti, Cr, and Mo segregation, however, pronounced Ni segregation in the inter-dendritic regions persisted, as evident from the simulated profiles in Fig. 8(b,c). Predictions for heat treatment at a higher temperature of 1200°C for 4 h produced nearly flat compositional profiles for all key elements, apart from Ni which was much more dispersed, suggesting effective homogenisation, with elimination of retained austenite and dissolution of the Laves particles in the inter-dendritic regions.

3.5. Assessment of homogenisation treatments in WAAM C465

Following the conclusion from the DICTRA simulations, the microstructure of WAAM C465 after homogenisation at 1200°C for 4 h (WAAM_H) is shown in Fig. 9. For brevity, only this suitable condition is shown here; detailed results for the other lower temperature

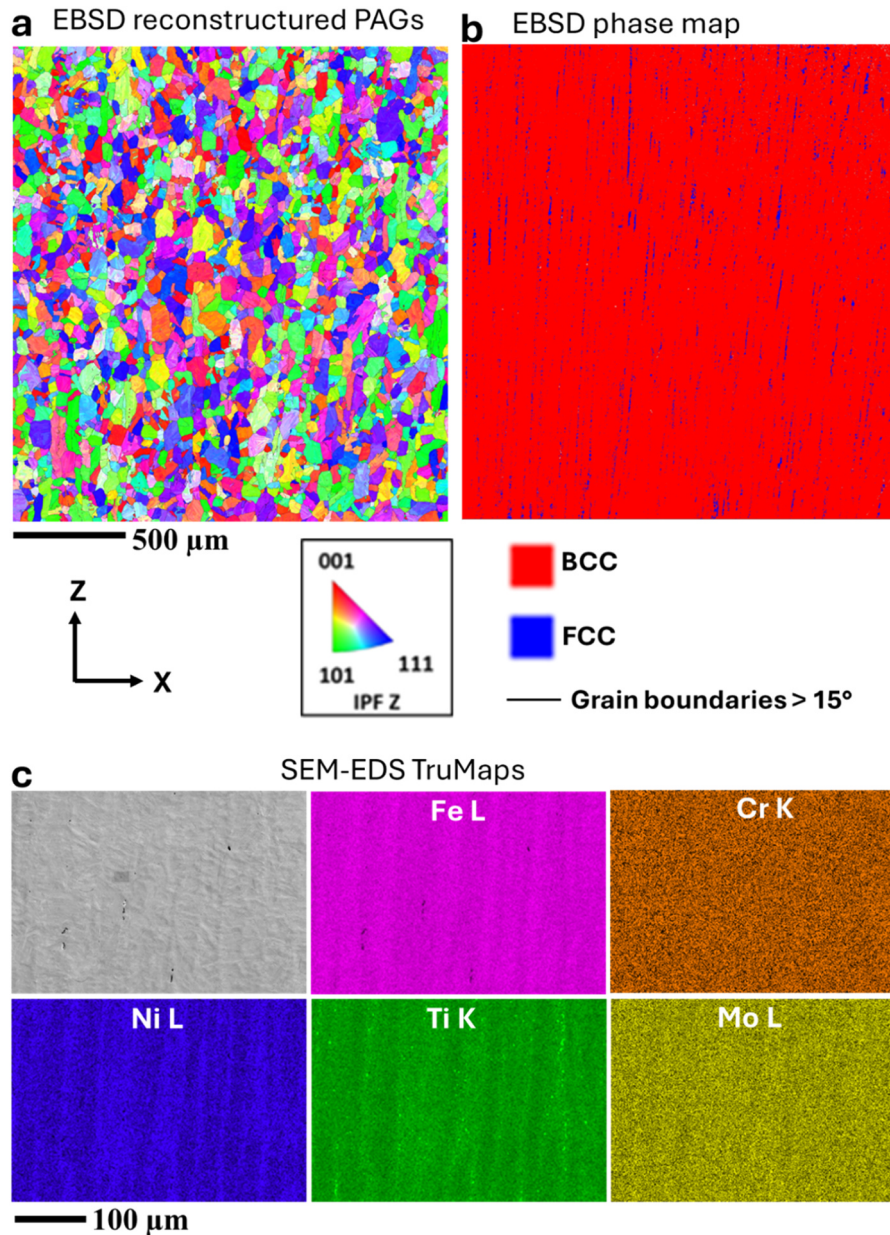


Fig. 5. Microstructure of WAAM C465 after a standard solution treatment (WAAM_SST). (a) EBSD IPF map of the reconstructed PAGs and (b) corresponding phase map. Grain boundary overlays were omitted in (b) to clearly visualise the distribution of retained austenite (blue colour). (c) SEM-EDS TruMaps showing the distribution of key alloying elements Fe, Cr, Ni, Ti and Mo.

homogenisation conditions studied (1100°C and 1150°C for 4 h) are provided in the supplementary file (Fig. S2 and Fig. S3, respectively). The average PAG size after homogenisation at 1100°C and 1150°C was $97 \pm 36 \mu\text{m}$ and $108 \pm 31 \mu\text{m}$, respectively; whereas homogenisation at 1200°C, had the disadvantage that it led to a marked increase in PAG size to $198 \pm 58 \mu\text{m}$ (Fig. 9(a)), indicating more pronounced grain growth at this elevated temperature. However, the EBSD phase map revealed no detectable retained austenite after homogenisation at 1200°C (Fig. 9(b)), indicating that the high temperature treatment resulted in complete transformation to austenite, which then fully converted to martensite upon cooling. Elemental analysis also confirmed that homogenisation at 1100°C and 1150°C for 4 h was insufficient to remove the Ni-rich regions along the inter-dendritic boundaries (Fig. S2 and Fig. S3, respectively), corroborating well with the DICTRA simulation results in Fig. 8(b,c). Since enrichment in Ni (and Ti) locally lowers the martensitic start (M_s) temperature and stabilises retained austenite [44], the lower-temperature conditions (1100 °C and 1150 °C) were not

expected to fully eliminate inter-dendritic retained austenite. In comparison, homogenisation at 1200°C, resulted in a nearly homogeneous distribution of the alloying elements Cr, Ti, and Mo, with significantly reduced inter-dendritic microsegregation and Laves particles (Fig. 9(c)). However, much less concentrated, Ni was still slightly enriched near the inter-dendritic regions. Overall, this demonstrates that homogenisation at 1200°C for 4 h was effective in eliminating inter-dendritic microsegregation to the extent that it prevented significant retained austenite, consistent with the DICTRA simulation results in Fig. 8(d). Fine Ti-rich nitrides and oxide particles, were also still randomly distributed within the microstructure, even after homogenisation at higher temperature, reflecting the high stability of TiN and TiO, as commonly reported in other AM fabricated Ti-containing martensitic steels [18,45].

3.6. Final heat treatment conditions

To be representative of the typical in-service material condition, it is

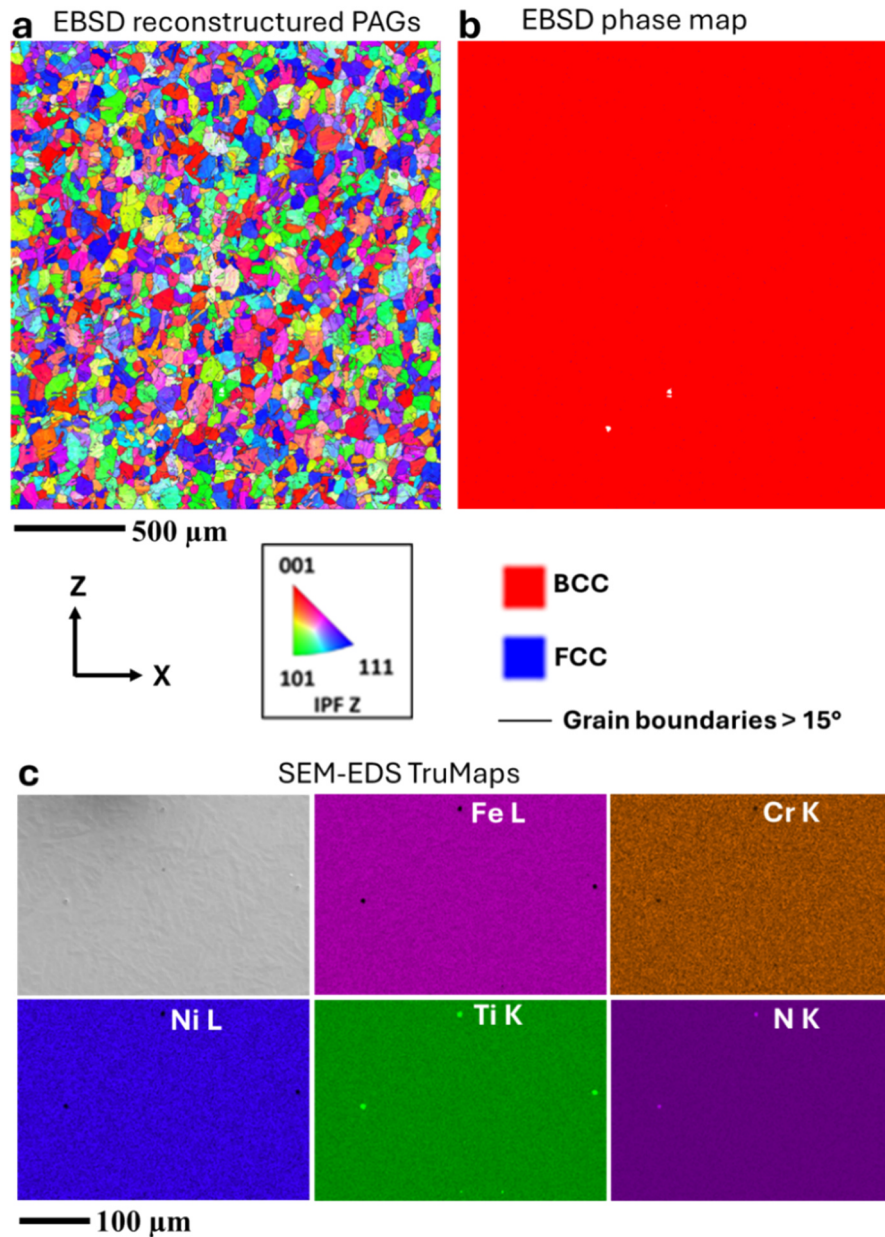


Fig. 6. Microstructure of wrought C465 bar after a standard solution treatment (WR_SST). (a) EBSD IPF map of the reconstructed PAGs and (b) corresponding phase map. (c) SEM-EDS TruMaps showing the distribution of Fe, Cr, Ni, Ti and N. Here, the reference Z-axis corresponds to the rolling direction.

important to briefly review any further microstructural changes that occur after SST and ageing H950 (A) before discussing the effect of the WAAM processing route on the localised corrosion behaviour. These observations are compared across the three main processing routes described in Section 2.3: (i) WAAM with the standard solution treatment and ageing (WAAM_SST + A), (ii) WAAM with prior homogenisation step before standard solution treatment and ageing (WAAM_H + SST + A), and (iii) the fully heat-treated wrought bar (WR_SST + A). The representative microstructures of the WAAM_SST + A and WAAM_H + SST + A samples are shown in Fig. 10 and wrought WR_SST + A sample in Fig. 11. The non-homogenised WAAM_SST + A condition exhibited fine recrystallised grains with an average PAG size of $69 \pm 11 \mu\text{m}$ (Fig. 10(a)), whilst the pre-homogenised WAAM_H + SST + A condition (Fig. 10(c)) exhibited slightly coarser PAG size of $85 \pm 17 \mu\text{m}$. Nevertheless, when compared to the initial homogenised condition (WAAM_H, Fig. 9(a)), it appears that subsequent application of the SST to the homogenised sample led to significant PAG refinement,

considering that ageing temperature is well below the austenitisation temperature of the alloy and will therefore have no impact on the PAG size.

The EBSD phase maps further revealed an increase in austenite fraction to $\sim 9.2\%$ on ageing the directly solution treated WAAM sample (WAAM_SST + A, Fig. 10(b)), compared to $\sim 3\%$ observed in the prior WAAM_SST condition (Fig. 5(b)), indicating that for this condition, the ageing treatment promoted austenite reversion [46]. However, very small fraction of austenite ($\sim 0.1\%$) was detectable by EBSD in the homogenised WAAM sample after full heat treatment to the same ageing treatment (WAAM_H + SST + A, Fig. 10(d)), indicating that homogenisation step prior to SST and ageing treatment suppressed the reversion transformation. For comparison, the microstructure evolution of the wrought C465 bar after SST and ageing treatment (WR_SST + A) was also investigated, as shown in Fig. 11. For the wrought C465, as expected, the average PAG size ($53 \pm 18 \mu\text{m}$) showed negligible change after the ageing treatment (Fig. 11(a)) as compared to the prior SST

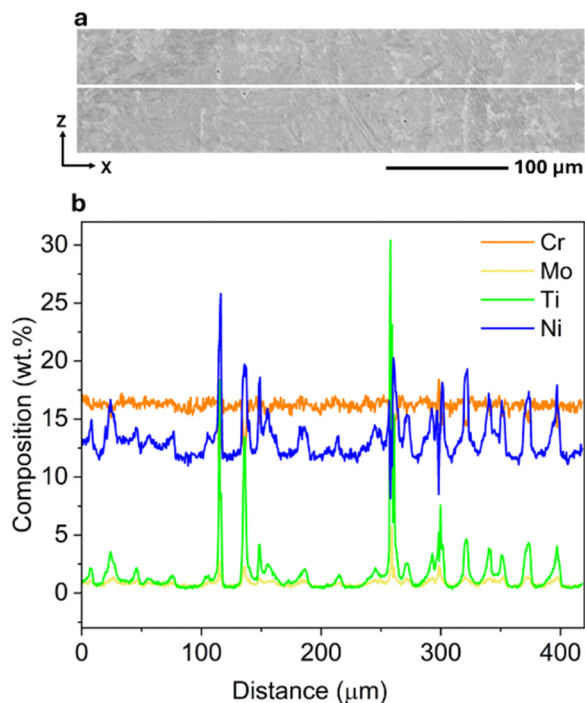


Fig. 7. (a) Low-magnification SE image of as-printed WAAM C465; (b) corresponding EDS line scan profile showing the variation of key alloying elements Cr, Ni, Ti, and Mo across dendritic and inter-dendritic regions, covering a distance of 418 μm .

condition (WR_SST, Fig. 6(a)). The EBSD phase map of wrought WR_SST + A sample (Fig. 11(b)) revealed a very low austenite content ($\sim 0.5\%$), comparable to that observed in the pre-homogenised WAAM_H + SST + A condition (Fig. 10(d)). A summary of PAG size and austenite content after different manufacturing and post-processing conditions is presented in Table 4. It should be noted that the quantitative analysis of retained/reverted austenite by EBSD technique may not be accurate due to its limited spatial resolution and the difficulty in resolving thin austenite films [46].

3.7. Electrochemical evaluation of pitting corrosion behaviour

3.7.1. CPP analysis of pitting corrosion behaviour

Cyclic potentiodynamic polarisation tests were performed in 0.6 M NaCl to compare the localised corrosion behaviour of the WAAM produced and wrought C465 materials. The WAAM produced samples were tested after two final heat treatment conditions, following: (i) the SST and ageing (WAAM_SST + A) and (ii) the prior homogenisation step combined with the SST and ageing (WAAM_H + SST + A). The results were compared to the fully heat-treated wrought C465 sample (WR_SST + A). The CPP curves are shown in Fig. 12(a), and the corresponding threshold potentials are summarised in Fig. 12(b). As is evident from the CPP curves, all the material conditions developed a distinct passive region, which eventually breaks down at higher potentials, resulting in a positive hysteresis loop indicative of pitting corrosion [32]. No significant difference in E_{corr} was observed among the tested samples after the same immersion time. However, the E_{pit} and E_{rep} potentials varied significantly with processing route, as shown in Fig. 12(b), indicating differences in resistance to localised corrosion. For example, the non-homogenised WAAM_SST + A sample exhibited the lowest E_{pit} (0.12 ± 0.01 V) and E_{rep} (-0.01 ± 0.02 V), reflecting higher susceptibility to pit initiation and limited repassivation capability. In contrast, the addition of the homogenisation step to the WAAM processing route (WAAM_H + SST + A) significantly improved the pitting resistance, with both E_{pit} (0.2 ± 0.02 V) and E_{rep} (0.08 ± 0.006 V) values shifting

towards more noble potentials. The fully heat-treated wrought WR_SST + A sample exhibited the highest E_{pit} (0.23 ± 0.008 V) and E_{rep} (0.07 ± 0.009 V) values, indicating superior pitting resistance among all the tested materials. Interestingly, the pre-homogenised WAAM_H + SST + A sample exhibited E_{pit} and E_{rep} values that were comparable to those of the wrought counterpart, indicating the effectiveness of homogenisation in enhancing pitting corrosion resistance. Additionally, the CPP curves revealed metastable pitting events during the anodic scan across all the tested conditions (Fig. 12(a)). The characteristic of metastable events are often related to a material's susceptibility to stable pitting [33]. Therefore, a comparative analysis of these transients is critical for understanding the transition from metastable to stable pit growth and its role in governing localised corrosion behaviour, as these insights cannot be fully obtained from CPP curves.

3.7.2. ECN-based evaluation of metastable pitting activity

To address the limitations of CPP in resolving short-lived transients, more sensitive ECN measurements were employed to evaluate the metastable pitting events by capturing the natural fluctuations of current and potential under free-corrosion conditions, without the need for external perturbation [38]. ECN measurements were compared for the same processing conditions as above, namely: WAAM_SST + A, WAAM_H + SST + A, and WR_SST + A conditions. Fig. 13(a-f) shows the ECN record obtained from coupled specimens and the corresponding frequency of metastable events. In all cases, an initial rise in potential was observed, while the current tends towards zero. This is normally associated with the formation and thickening of the passive film and indicates that there were no substantial galvanic effects. After this transitory phase, metastable pitting events can be observed as sharp drops in the potential, which are followed by a sudden repassivation (restoration of the potential). Metastable pitting events can also be observed as peaks in the current, the sign of which indicates where the metastable event occurred, with the positive current peaks corresponding to the WE and the negative peaks to the G electrode. Hence, by counting positive and negative pitting events, it is possible to understand distinctly how many events occurred on the WE and G electrodes. A clear distinction in the frequency of current transients was observed across the pairings for the different processing conditions. When non-homogenised WAAM_SST + A sample was compared with the pre-homogenised WAAM_H + SST + A sample (Fig. 13(a,b)), a substantially higher frequency of current transients was observed in the former. Similarly, comparison with wrought WR_SST + A sample (Fig. 13(c,d)) revealed that non-homogenised WAAM_SST + A sample again exhibited a higher transient frequency. These observations suggest that, among all the C465 samples tested, the non-homogenised WAAM_SST + A sample had inferior passive film stability and a greater tendency towards stable pit formation. In contrast, the pairing of pre-homogenised WAAM_H + SST + A sample with wrought WR_SST + A sample (Fig. 13(e,f)), revealed nearly comparable transient activities. These results indicate that the homogenisation treatment prior to SST and ageing significantly suppressed metastable pitting events in WAAM produced C465 material, yielding a response comparable to that of the wrought counterpart.

3.7.3. Identification of pit initiation sites after anodic polarisation

Post-exposure surface analysis, as revealed by high resolution, low-kV SEM-EDS analyses in Fig. 14, demonstrated that pit initiation was strongly influenced by microstructural features of C465 that could be correlated to the samples respective processing routes. Post anodic polarisation, the WAAM_SST + A sample corroded surface revealed that the pits were formed along the (Ni, Ti)-enriched and Cr-depleted inter-dendritic networks (Fig. 14(a)). A magnified image from the inter-dendritic region further showed nano-scale pit nucleation sites near the Ti-rich Laves/ matrix interface (Fig. 14(b)). Additionally, pit nucleation sites were observed near the interfaces of TiN inclusion

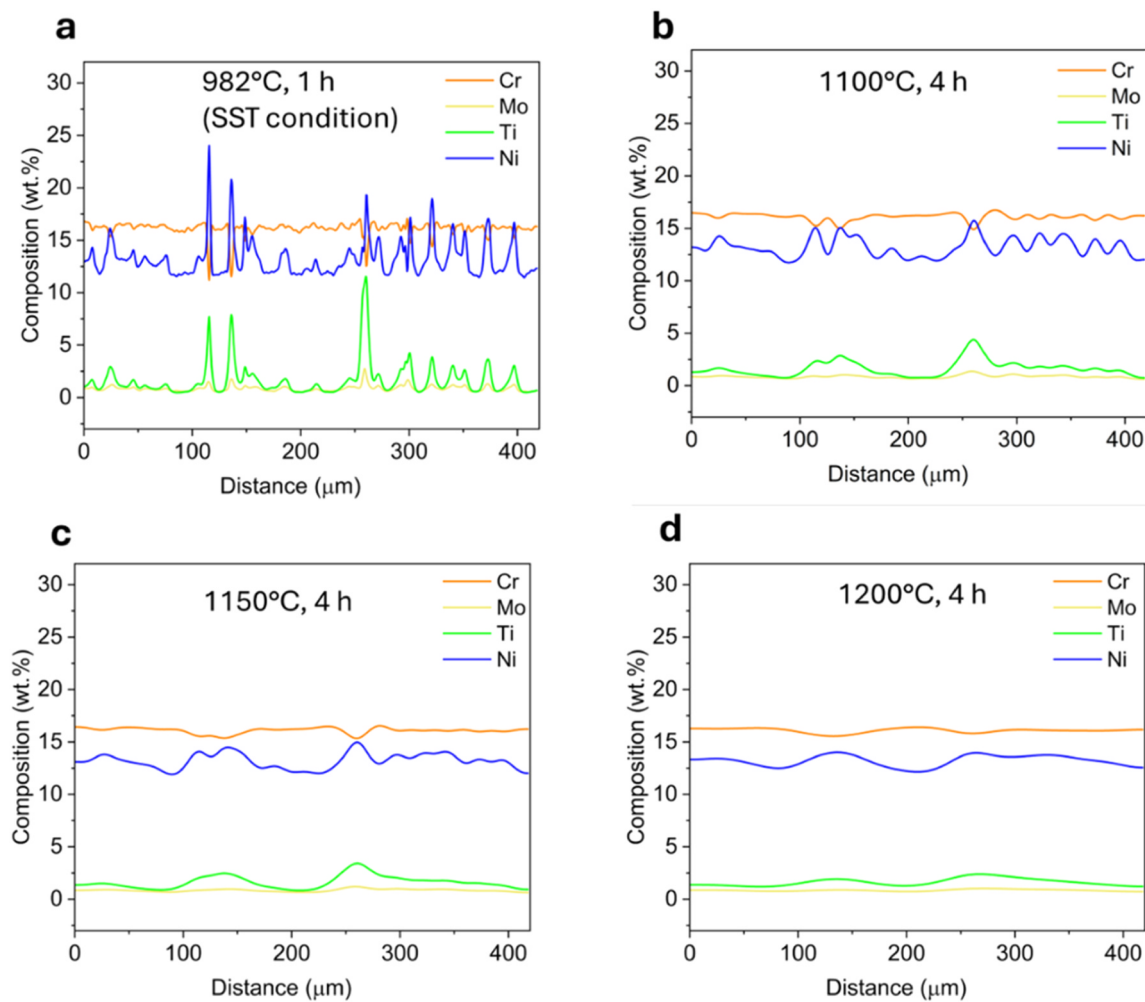


Fig. 8. TC-Python automated DICTRA simulations based on the SEM-EDS line scan data in Fig. 7, evaluated under various thermal conditions: (a) 982°C for 1 h (SST condition), (b) 1100°C for 4 h, (c) 1150°C for 4 h and (d) 1200°C for 4 h.

present in both WAAM and wrought microstructures (Fig. 14(c,d), respectively). Notably, due to the much higher solidification rate, the TiN inclusions in the WAAM sample were significantly smaller (~ 250 nm) than in wrought sample (~ 2 μ m), which resulted in smaller pit formation around these inclusions in the WAAM produced sample.

4. Discussion

This study presents, for the first time, a comprehensive investigation into the effect of WAAM processing on the corrosion performance of an ultra-high strength precipitation-hardened martensitic stainless steel C465. Particular emphasis was paid to the potential of post-build heat treatments to improve key microstructural features that govern susceptibility to pitting corrosion. WAAM produced material was systematically benchmarked against a conventionally manufactured and fully heat treated wrought C465 material. The electrochemical techniques, including CPP and ECV were employed to elucidate the role of microstructural modifications in governing pitting corrosion behaviour. These findings provided critical insights into the interplay between WAAM process, post-printing heat treatment strategies, and corrosion resistance, offering a pathway toward performance-driven processing of WAAM C465 components.

4.1. Effect of the WAAM process on the C465 as-printed microstructure

The as-printed WAAM C465 material developed a microstructure

distinctly different from its wrought counterpart, due to the markedly different heat input and solidification characteristics between the two processing routes and the lack of post thermo-mechanical processing in additive manufacturing. During WAAM deposition, a small melt pool (relative to billet casting) is translated across the workpiece with the material added using a wire feed system. Each new melt track partially remelts the top of the previously deposited one, and solidification proceeds at the rear of the melt pool predominantly upward but tilted slightly towards the deposition direction due to the curvature of the melt pool surface [47]. As the melt pool is heated, the thermal gradient in the liquid is positive and constitutional undercooling is often insufficient for free nucleation ahead of the solidification front [48]. Consequently, the solidification grain structure is typically dominated by columnar grain growth of the primary solid solution phase during solidification, which grows epitaxially from the grains melted back at the fusion boundary. Repeated passes with directional heat extraction (predominantly down the wall into the substrate) thus results in elongated columnar PAGs, as seen in the large area EBSD IPF map in Fig. 3(b). However, because the solidification rate is much higher than in billet casting, the dendrite spacings and any primary particles generated in the liquid are substantially refined [17]. For example, the typical dendrite arm spacing in a cast forging billet is ~ 200 μ m [49], whereas in WAAM it is in the range of 2–8 μ m (Fig. 3(c)).

In addition to the columnar grain morphology, the solidified WAAM C465 microstructure exhibited pronounced microsegregation across its dendritic and inter-dendritic regions, as evident from the elemental

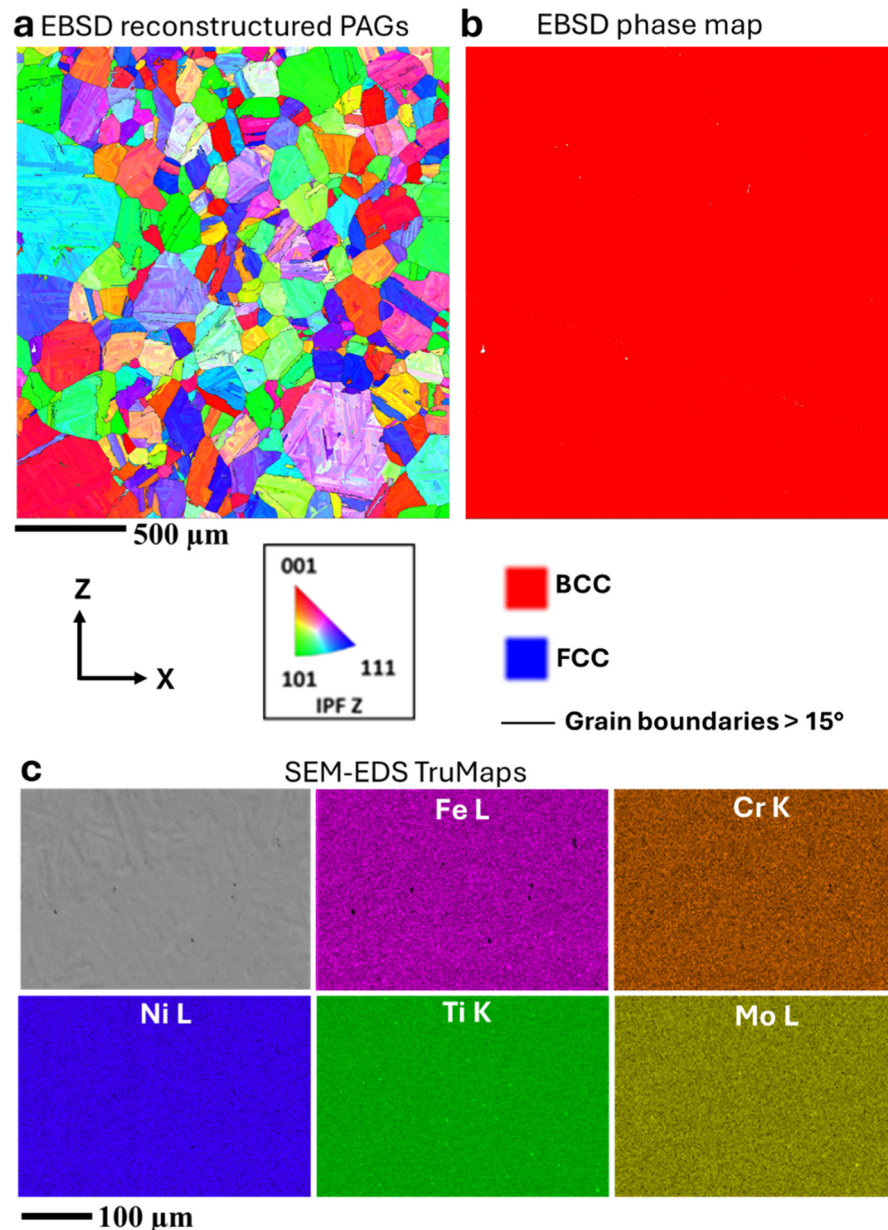


Fig. 9. Microstructure of WAAM C465 after homogenisation at 1200°C for 4 h (WAAM_H). (a) EBSD IPF map of the reconstructed PAGs and (b) corresponding phase map. (c) SEM-EDS TruMaps showing the distribution of key alloying elements Fe, Cr, Ni, Ti and Mo.

mappings in Fig. 3(c). However, the scale of this is much finer than in a cast billet, as noted above. The WAAM fabricated microstructure also revealed the presence of retained austenite in the inter-dendritic networks, confirmed by the high-resolution EBSD phase map in Fig. 4(b), and these regions were predominantly enriched in Ni and Ti (Fig. 4(c)). This localised enrichment of solute elements explains the preferential stabilisation of retained austenite along the inter-dendritic networks, as CALPHAD-based thermodynamic assessments have shown that Ti (which is typically classified as a ferrite stabiliser) can lower the M_s temperature and reduce the Gibb's free energy difference between austenite and martensite [44]. The co-location of retained austenite within a martensitic microstructure with the Ti-enriched inter-dendritic network has also been reported in other martensitic steels produced by AM [45,46,50].

Furthermore, (Ti, Mo, and Si)-rich Laves and few (Mo, Ti)-rich particles were also observed in the inter-dendritic networks (Fig. 4(b-c)). Scheil-Gulliver solidification simulations in Fig. 2(a) indicate solidification initiates with austenite formation, while solute elements such as

Ti, Ni, Mo, and Si are progressively rejected into the remaining liquid, promoting Laves formation during the final stages of solidification [45]. Discrete TiN inclusions with O dissolved in them were also observed within both dendritic and inter-dendritic regions, consistent with the Scheil-Gulliver solidification prediction. These Ti-based inclusions have also been reported in C465 fabricated by PBF-L [18], where they are believed to originate from oxygen contained in the feedstock powder particles. Accordingly, the feedstock wire or other sources of atmospheric exposure could introduce inclusions in the WAAM C465. Microstructural characterisation of the WAAM feedstock wire, as shown in the supplementary file (Fig. S4), revealed presence of TiN ($\sim 8 \mu\text{m}$) and nano-scale Al-rich oxide inclusions. During the WAAM deposition, these pre-existing highly stable inclusions can survive remelting of the wire, leading to the formation of Ti-rich inclusions in the final solidified WAAM C465 [16].

In summary, the WAAM processing route characteristically produces columnar PAG structures, with pronounced microsegregation, as well as retained austenite, Laves particles and Ti-rich inclusions, leading to a

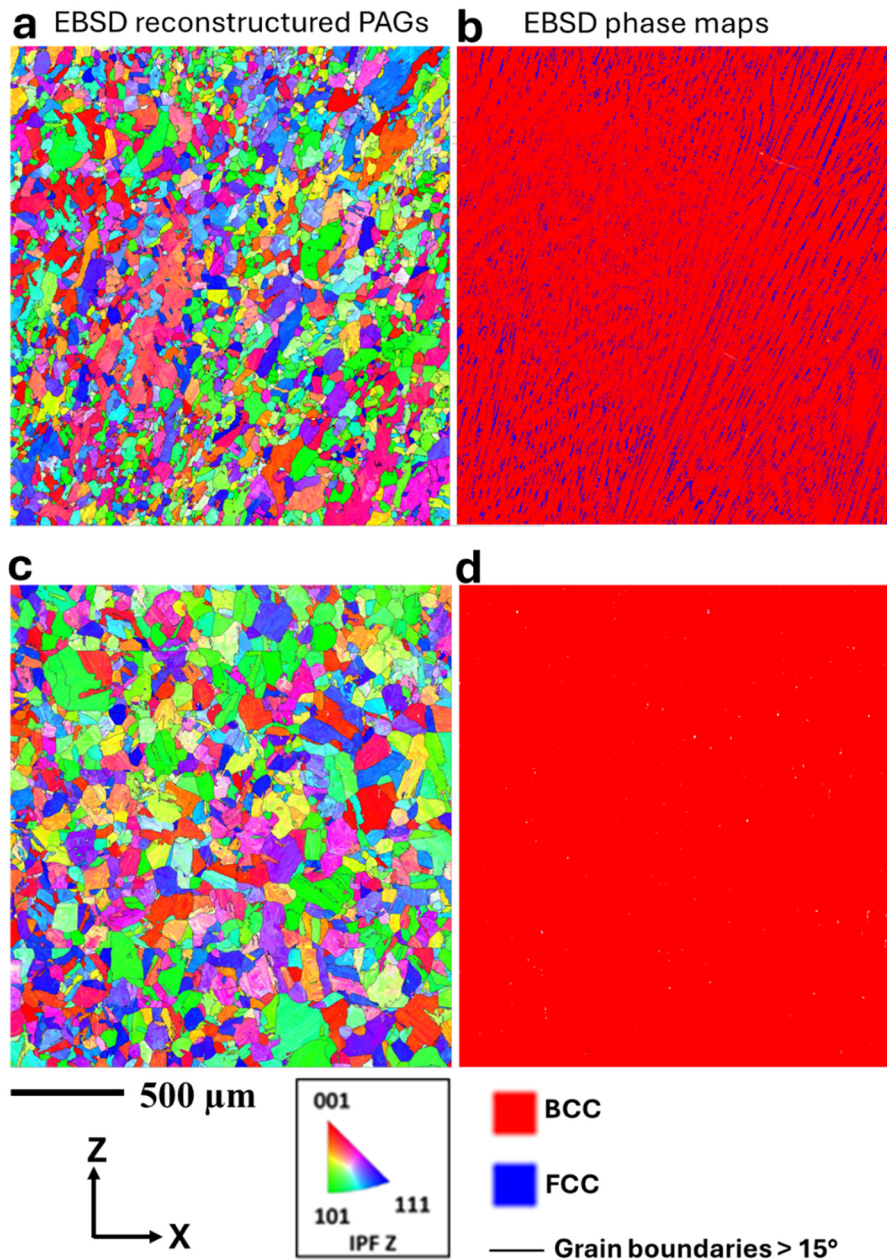


Fig. 10. Microstructure of WAAM C465 after standard solution treatment and H950 ageing. (a,c) EBSD IPF map of the reconstructed PAGs and (b,d) corresponding phase maps. (a,b) Non-homogenised WAAM_SST + A and (c, d) Pre-homogenised WAAM_H + SST + A conditions.

heterogeneous microstructure. Such microstructural features are expected to impart anisotropic properties and critically influence the reliability and performance of the WAAM fabricated C465 components in demanding applications. However, fortunately, the solidification microstructure is also much finer in scale than in a cast forging billet and would be expected to be more amenable to improvement by post-build heat treatment, without resorting to additional thermo-mechanical processing.

4.2. Effect of post-build heat treatments on the WAAM C465 microstructure

4.2.1. Grain refinement

The recommended standard solution treatment for C465 forged billets (AMS 5936 C) produced significant PAG refinement in the WAAM produced microstructure WAAM_SST (Table 4). This refined grain structure was comparable to that seen in the wrought bar microstructure

(WR_SST), demonstrating that in this alloy SST alone was able to produce sufficient PAG refinement through recrystallisation of the austenite that reforms on heating above the austenite phase boundary. This is an interesting phenomenon which does not always occur on re-heating martensitic steels. For many alloys, a memory of PAG structure is reproduced by phase transformation during re-heating, resulting in limited grain refinement, as has been discussed at length in [42,51,52]. However, more rapid heating rates and heavily alloyed steels tend to provide sufficient stored energy for the martensite-to-austenite transformation to drive recrystallisation [52].

4.2.2. Chemical homogenisation

The standard solution treatment recommended for wrought products was insufficient to reduce inter-dendritic microsegregation and eliminate retained austenite and Laves phase in the WAAM processed material (Fig. 5). This indicates that the applied SST was inadequate to enable sufficient diffusion of solute elements such as Ni and Ti away from the

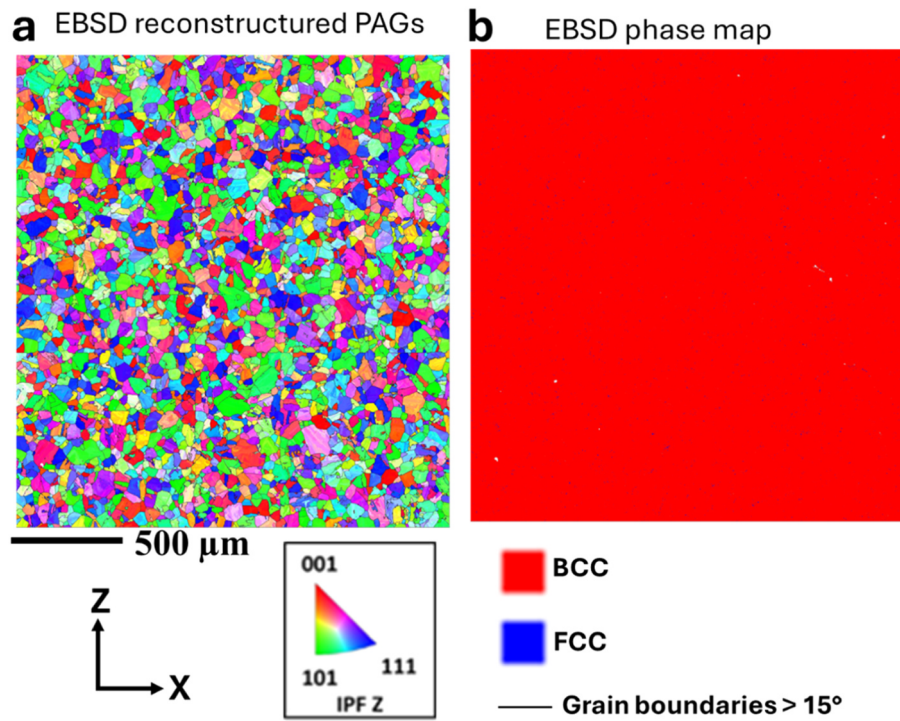


Fig. 11. Microstructure of wrought C465 bar after standard solution treatment and H950 ageing (WR_SST + A). (a) EBSD IPF map of the reconstructed PAGs and (b) corresponding phase map. Here, the Z-axis corresponds to the rolling direction.

Table 4

Summary of PAG size and austenite content after different manufacturing and post-processing conditions.

Sample conditions	PAG size (μm)	Austenite content %
WAAM	in millimetres	8.4
WAAM_SST	62 ± 22	3
WR_SST	54 ± 16	0.2
WAAM_H	198 ± 58	0
WAAM_SST + A	69 ± 11	9.2
WAAM_H + SST + A	85 ± 17	0.1
WR_SST + A	53 ± 18	0.5

inter-dendritic regions which consequently suppressed the M_s temperature sufficiently and enabled inter-dendritic austenite to persist even after deep cryogenic cooling at -73°C . In contrast, the wrought bar, which had undergone a much more extensive homogenisation and

thermo-mechanical processing history [10], exhibited a near-homogeneous distribution of key alloying elements after the same heat treatment (WR_SST), although it contained some coarser TiN inclusions (Fig. 6(c)). A smaller fraction of retained austenite was observed in wrought WR_SST sample (Table 4), indicating a near-complete transformation of austenite to martensite upon deep cryogenic cooling. These findings suggest that, while the standard heat treatment is effective for wrought C465 material, a tailored homogenisation strategy is required for WAAM produced C465 to mitigate microsegregation and secondary phases formed during solidification.

The DICTRA simulations provided insight into the homogenisation behaviour of the WAAM C465 and were able to predict suitable conditions to remove the chemical gradients present in the solidified material. Experimental homogenisation treatment (at 1200°C for 4 h) showed good agreement with the DICTRA predictions, revealing a nearly

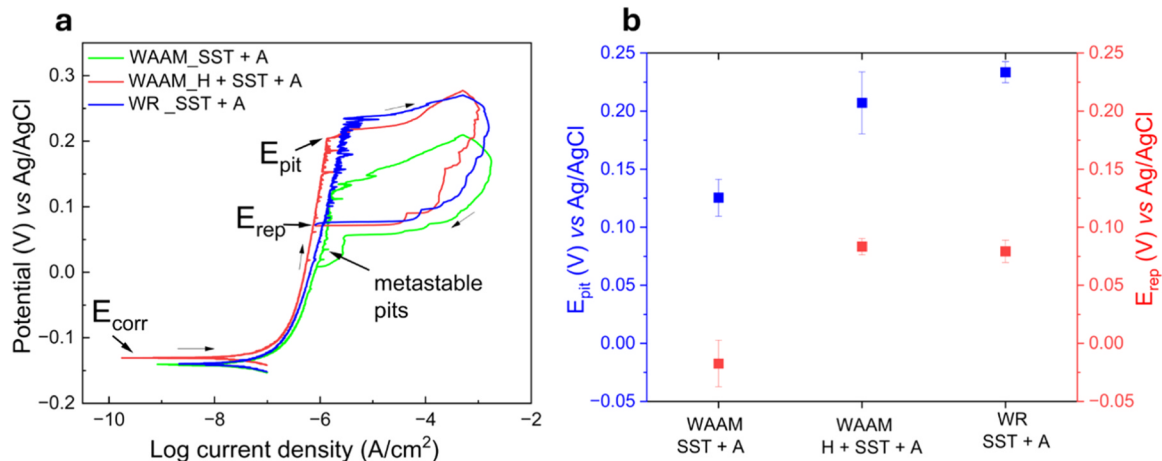


Fig. 12. (a) Representative CPP curves of WAAM produced and wrought C465 samples in 0.6 M NaCl solution at room temperature and (b) Summary of E_{pit} and E_{rep} potentials obtained from the CPP curves.

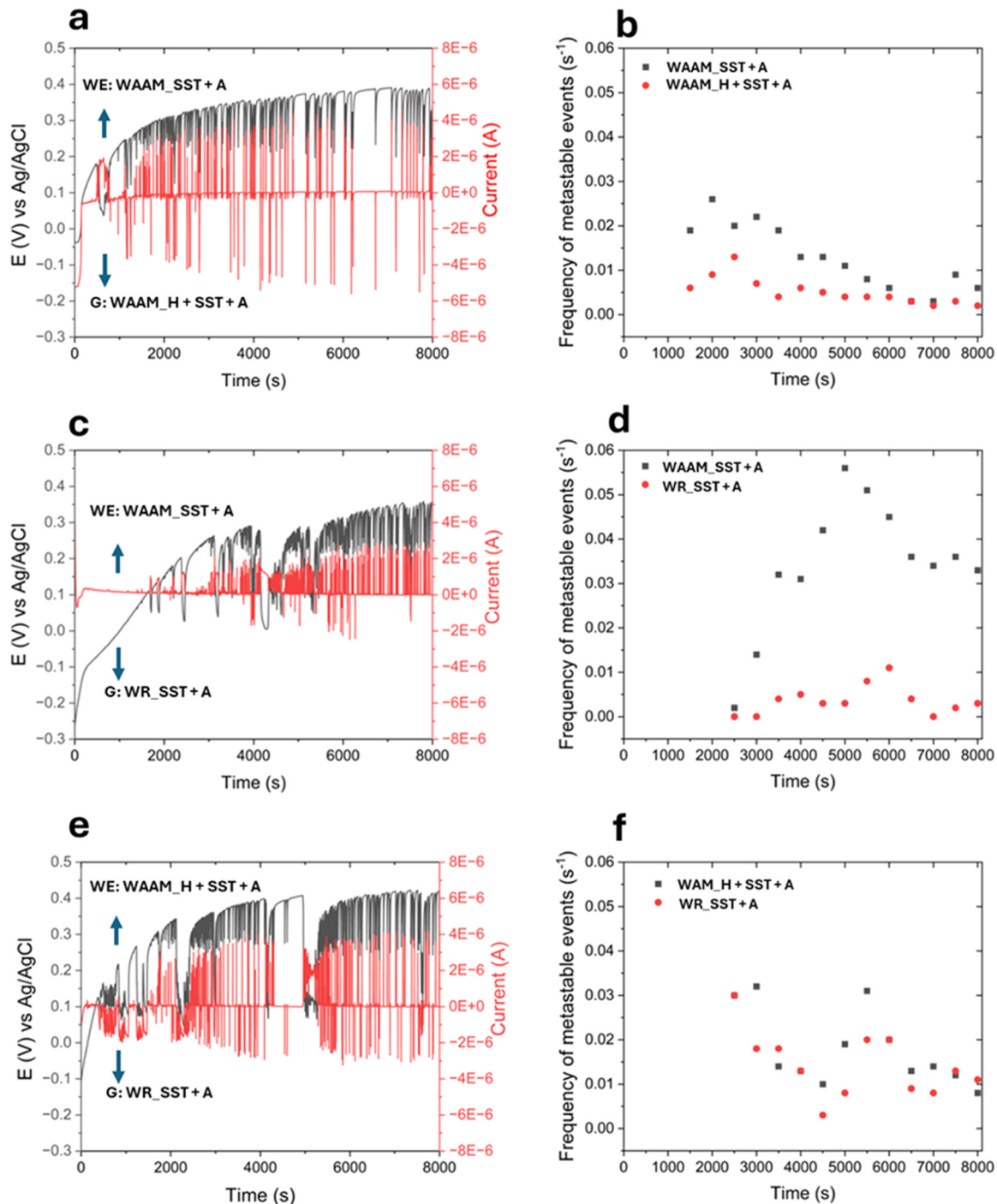


Fig. 13. (a,c,e) Electrochemical noise records showing potential (black) and current (red) transients of C465 tested in 4.6 mM FeCl₃ + NaCl solution to reach a total chloride ions concentration of 0.6 M. In each pairwise test, one specimen was designated as working electrode (WE) and the other as the ground electrode (G), as indicated. (b,d,f) Frequency of metastable events obtained from the electrochemical noise records. Comparisons are shown for (a,b) WAAM_SST + A (WE) vs WAAM_H + SST + A (G); (c,d) WAAM_SST + A (WE) vs WR_SST + A (G); and (e,f) WAAM_H + SST + A (WE) vs WR_SST + A (G).

uniform microstructure largely free from inter-dendritic micro-segregation and Ti-rich Laves particles (Fig. 9). The enhanced diffusion and redistribution of Ni and Ti at high homogenisation temperatures reduced the localised inter-dendritic enrichment that would otherwise stabilise the retained austenite. As a result, homogenised WAAM_H sample was almost free from inter-dendritic retained austenite (Fig. 9(b) and Table 4).

4.2.3. Grain growth

Unfortunately, high temperature heat treatments have the potential

to promote PAG growth which is undesirable in C465 aerospace products [13], so a compromise with the level of homogenisation that can be achieved is required. For example, homogenisation at the more effective temperature of 1200°C led to a pronounced influence on the PAG size. The average PAG size remained relatively stable at lower temperatures, but increased sharply at higher temperature of 1200°C, indicating significant grain coarsening. This abrupt change was likely associated with dissolution of the Laves phase, which reduced the Zener grain boundary pinning and consequently enabled rapid grain growth at higher temperatures. Lian et al. [13] have reported a similar behaviour in wrought

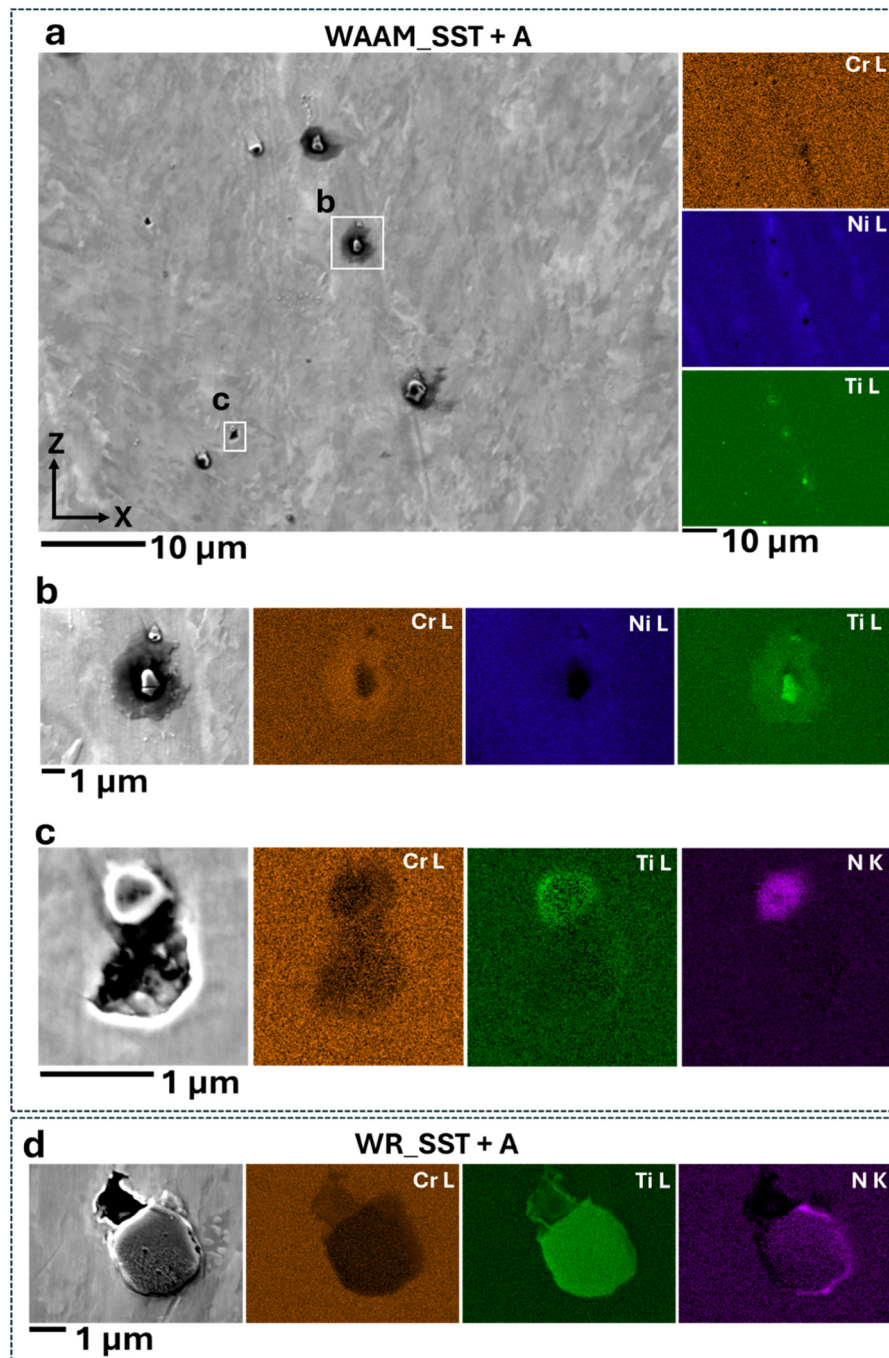


Fig. 14. Corroded surfaces of (a-c) non-homogenised WAAM_SST + A sample and (d) wrought WR_SST + A sample after anodic polarisation at + 1.0 V (vs Ag/AgCl) for 2 s in 0.6 M NaCl solution. High resolution, low-kV SEM-EDS TruMaps show nano-to microscale pit nucleation sites: (a) along the (Ni, Ti)-enriched and Cr-depleted inter-dendritic regions, (b) at the Ti-rich Laves/matrix interface (c) at TiN inclusion/matrix interface in non-homogenised WAAM_SST + A sample and (d) at TiN inclusion/matrix interface in wrought WR_SST + A sample.

C465 material, where grain growth became significant above 1000°C due to dissolution of the Laves particles. Notably, the Laves dissolution temperature in wrought C465 is substantially lower than observed in the present WAAM C465 material. This difference can be attributed to the elemental segregation which led to localised enrichment of Laves-forming elements, particularly Ti, Mo, and Si in the inter-dendritic regions of the WAAM material (Fig. 4(c)). This enrichment increases the thermal stability of the Laves, delaying its dissolution and necessitating higher homogenisation temperatures. While the dissolution behaviour of Laves in other AM fabricated Ti-rich maraging steels has not been systematically studied, similar trends have been reported in AM Inconel

718, where Laves dissolution occurs between 1100–1200°C [53].

4.2.4. Reverted austenite

After the final heat-treatment step required to achieve full in-service properties, the aged microstructure of WAAM C465 is critically governed by both the prior heat treatment history and the compositional heterogeneity inherited from the WAAM process. Pronounced phase transformation behaviour was observed in the non-homogenised WAAM_SST + A condition, which exhibited a notable increase in austenite fraction as compared to its pre-aged condition (WAAM_SST, Table 4). This increase is due to the martensite to austenite reversion

transformation during the ageing process [46,54]. The austenite reversion typically occurs through the diffusion-controlled decomposition of martensite and is strongly influenced by the localised segregation of Ni which provides favourable conditions for stabilising reverted austenite [54]. It is worth noting that the pre-aged condition (WAAM_SST) contained retained austenite along the Ni-enriched inter-dendritic regions (Fig. 5(b) and Table 4), which can provide both the crystallographic and chemical conditions necessary for the formation of reverted austenite [55]. Pre-existing retained austenite has been widely recognised as a preferential nucleation site owing to its identical FCC crystal structure, which significantly reduces the energy barrier for austenite reversion [54,56]. Ni enrichment resulting from solidification-induced microsegregation and Ni rejection during precipitation is also known to be important to the promotion of austenite reversion in AM alloys [54,57].

In contrast, the WAAM_H + SST + A condition, which underwent homogenisation at 1200°C prior to SST, exhibited a very limited extent of austenite reversion after ageing (Table 4). The suppression of austenite reversion in this case further reinforces the fact that pre-existing retained austenite and localised Ni-segregation play a critical role during the reversion transformation. A comparable behaviour has been reported in PBF-L maraging steels, where a higher temperature solution treatment also eliminated both retained and reverted austenite [54]. In comparison, the wrought WR_SST + A sample which contained a small amount of retained austenite before ageing exhibited only a marginal increase in austenite fraction after ageing (Table 4), consistent with its greater chemical homogeneity and negligible retained austenite content prior to the ageing treatment, as seen in Fig. 6.

Fig. 15 schematically summarises the microstructure evolution in C465 resulting from various manufacturing routes and subsequent heat treatment steps. The WAAM process (Fig. 15(a)) exhibited a heterogeneous microstructure characterised by elongated columnar structure, pronounced solute-enriched (primarily Ni and Ti) inter-dendritic regions associated with retained austenite, (Ti, Mo)-rich Laves phase, and the presence of Ti-rich oxide and nitride inclusions. Although the standard heat treatment sequence (solution treatment and ageing) developed for the wrought material could significantly refine the grain structure, it was insufficient to redistribute Ni and Ti from the inter-dendritic regions and as a result failed to eliminate inter-dendritic retained/ reverted austenite and Laves phases. This limitation necessitated a suitable homogenisation step prior to the standard heat treatment to reduce the WAAM induced microstructural heterogeneity. The homogenisation behaviour in WAAM C465 is governed primarily by the diffusion kinetics of solute elements (mainly Ni and Ti) which otherwise stabilise the retained austenite in the inter-dendritic networks by locally lowering the M_s temperature, along with dissolution of the Ti-rich Laves phase and associated grain coarsening. In the present study, homogenisation at 1200°C for 4 h effectively reduced inter-dendritic segregation and suppressed retained austenite and dissolved the Laves phase albeit at the expense of grain coarsening. Following this homogenisation step, the subsequent standard solution treatment and ageing produced a microstructure that was comparable to the fully heat treated wrought C465 (Fig. 15(b)).

4.3. Pitting corrosion behaviour

Given that pitting often serves as the onset of SCC in chloride environments, this section focuses on how WAAM specific microstructural features like microsegregation, Laves phase, retained or reverted austenite and inclusions influence the pitting corrosion behaviour of C465 material.

The electrochemical response was evaluated after application of the ageing treatment, as this represents the final processing step employed for aerospace applications of the C465 material. However, the benefits of using a prior homogenisation step with the WAAM material were compared to simply applying a post-build standard solution treatment and ageing. The pitting corrosion behaviour of WAAM-produced C465 is

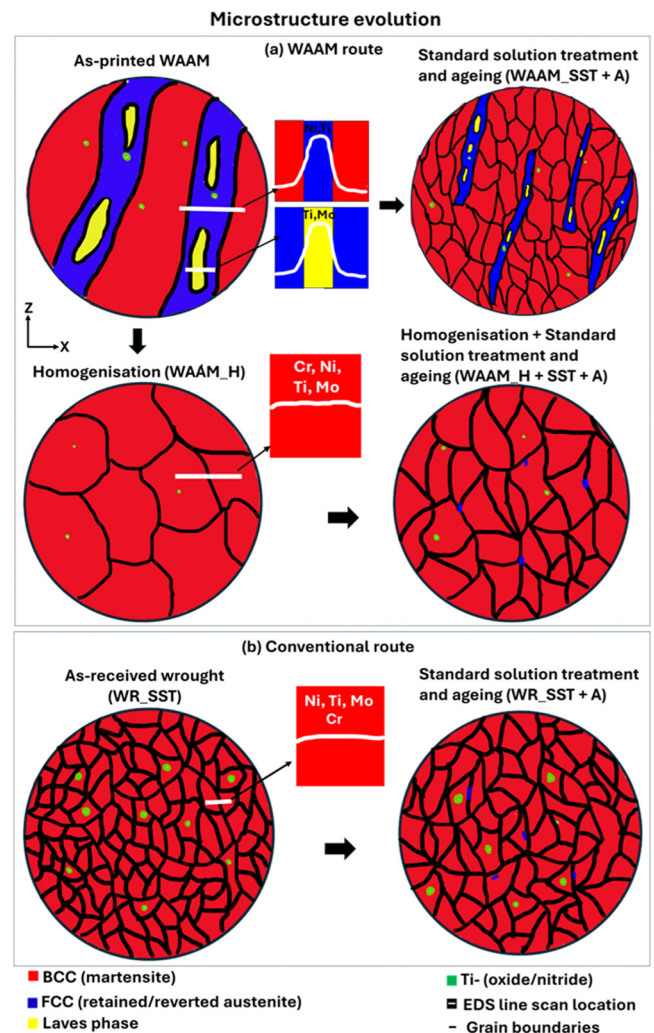


Fig. 15. Schematic illustration of microstructure evolution during the heat treatment sequence in C465 produced by (a) WAAM and (b) conventional manufacturing routes.

clearly dictated by the microstructural state imposed by post-processing heat treatments. From the CPP and ECN results, it can be deduced that while the pitting resistance of the WAAM material (WAAM_SST + A condition) is inferior to its wrought counterpart (WR_SST + A condition), the application of an effective homogenisation treatment prior to the SST and ageing treatment significantly enhanced its corrosion performance, comparable to the wrought benchmark.

The pit nucleation studies provided further useful insights into the observed pitting corrosion behaviour in WAAM produced and wrought C465 (Fig. 14). In WAAM_SST + A condition, pits were predominantly nucleated along (Ni, Ti)-enriched and Cr-depleted inter-dendritic networks. These (Ni, Ti)-enriched inter-dendritic regions aligned with columnar grain growth direction, correspond to the retained/reverted austenite phase (see Figs. 4(b) and 10(b)). Given that Cr is a ferrite stabiliser, it is likely to segregate preferentially in the martensitic regions, leading to local Cr deficiency in the inter-dendritic austenite regions (Fig. 4(b) and (c)). Since Cr plays a pivotal role in passive film formation and stability in stainless steels, this localised depletion possibly compromised the integrity of the passive film, rendering the retained/reverted austenite regions more susceptible to passive film breakdown under chloride exposure. Consequently, the inter-dendritic austenite acted as a preferential site for nucleation of metastable pits. This observation was consistent with the recent findings by Keogh et al. [58] who reported that reverted austenite in aged 15–5 PHMSS resulted

in inhomogeneous distribution of Cr in the microstructure, compromising the integrity of the passive film and increasing susceptible sites for localised corrosion attack. In AM fabricated martensitic steel, during ageing treatment the reversion transformation from martensite to austenite resulted in accumulating a high dislocation density in the interphase boundaries between them due to the volume change of the lattice transformation, creating favourable sites for pit nucleation [59]. Therefore, it can be reasonably inferred that insufficient diffusion of Ni and Ti from the inter-dendritic regions during SST resulted in higher fraction of austenite in non-homogenised WAAM_SST + A sample, which degraded the passive film stability and reduced pitting corrosion resistance. In contrast, homogenisation at 1200°C for 4 h effectively reduced the enrichment of austenite-stabilising elements (Ni and Ti) in the inter-dendritic regions. As a result, the pre-homogenised WAAM_H + SST + A sample exhibited a limited amount of reverted austenite, comparable to the benchmark wrought WR_SST + A sample. This microstructural refinement led to fewer metastable pitting events during the ECN test and improved passive film stability, yielding corrosion performance comparable to wrought WR_SST + A sample.

In addition, the presence of Ti-rich Laves particles embedded in the inter-dendritic austenite regions also contributed to the inferior pitting resistance of the non-homogenised WAAM_SST + A sample. High-resolution SEM-EDS analysis in Fig. 14(b), revealed a nano-scale pit at the interface of Laves particles and the surrounding austenite matrix. Similar pit initiation mechanisms have been reported in AM fabricated IN718 [53], where micro-galvanic coupling between the Laves and matrix has been proposed to promote preferential dissolution of the matrix following passive film breakdown. Liu et al. [60] demonstrated that Laves phase possesses lower Volta potential than the surrounding ferrite matrix, leading to selective dissolution and pitting in the matrix. These findings suggest that the Ti-rich Laves phase in WAAM produced C465 likely promoted micro-galvanic corrosion activity with the surrounding austenite, thereby facilitating pit nucleation at the Laves/matrix interface. This mechanism was consistent with the higher frequency of metastable events recorded for the WAAM_SST + A sample as compared to WAAM_H + SST + A sample (Fig. 13(a) and (b)), as homogenisation treatment dissolved the Laves particles (Fig. 9(c)), thereby significantly reducing the number of energetically favourable pit nucleation sites. As a result, the pre-homogenised WAAM_H + SST + A sample exhibited fewer ECN transients, a nobler E_{pit} and enhanced repassivation capabilities. In line with this understanding, the wrought WR_SST + A sample, which lacked Laves particles, demonstrated the lowest metastable activity and highest pitting resistance.

Inclusions were another factor that influenced the pitting behaviour of C465 steel. The TiN inclusions acted as preferential pit nucleation sites for both WAAM (Fig. 14(c)) and wrought C465 samples (Fig. 14(d)). The role of inclusions on the localised corrosion behaviour of stainless steel has been extensively studied [22,23]. Inclusion-induced passive film breakdown has been attributed to Cr-depleted zones around the inclusions, as well as the micro-galvanic coupling effect arising from the potential difference between the inclusion and surrounding matrix. Previous studies have suggested that pitting nucleation at smaller size inclusions (< 500 nm) rarely evolves into stable pits due to the rapid formation of a stable passivation film over the inclusion surface [25,61]. In the present study, the Ti-rich inclusions in WAAM C465 steel were observed to be significantly finer (~ 250 nm) compared to those in the wrought C465 (~ 2 µm). Accordingly, while nanoscale TiN inclusions in WAAM C465 are likely to contribute to pit initiation, they are unlikely to lead to stable pits. In contrast, larger micron-scale TiN inclusions in the wrought WR_SST + A sample were possibly the primary factor influencing its localised corrosion, despite its higher overall pitting resistance, which is consistent with reported observations for other wrought martensitic stainless steels [62].

To summarise, the effect of manufacturing routes and post-printing heat treatments on the pitting corrosion behaviour of the C465 material is schematically illustrated in Fig. 16. The WAAM fabricated C465

exhibited inferior pitting resistance compared to its wrought counterpart (Fig. 16a), primarily due to the WAAM-specific solidification features such as inter-dendritic microsegregation, retained/reverted austenite, and Laves particles. These microstructural characteristics were not sufficiently mitigated by the standard heat treatment protocols developed for wrought C465 material. Consequently, these features compromised passive film stability and increased pit nucleation susceptibility in WAAM C465 when exposed to a chloride environment. A suitable homogenisation treatment prior to the standard heat treatment protocols effectively reduced compositional gradients, dissolved Laves particles, and suppressed austenite formation. These microstructural improvements led to enhanced pitting corrosion resistance, and restored corrosion performance to levels comparable with wrought C465 (Fig. 16(b) and (c)).

Overall, the findings highlight the need for WAAM-specific post-processing strategies, as standard heat-treatments developed for wrought alloys cannot be directly applied to additively manufactured material, even when compositions are identical, because the thermal history and solidification path are fundamentally different. This work establishes a clear process–microstructure–corrosion relationship in WAAM C465 and provides a practical framework to mitigate solidification-induced pitting susceptibility in high-strength precipitation-hardened martensitic steels.

5. Conclusions

This study investigated the influence of WAAM processing and post-printing heat treatments on the microstructure and corrosion performance of C465, benchmarked against its wrought counterpart. Based on experimental results and DICTRA-guided simulations, the following conclusions were drawn:

- WAAM produced a heterogeneous microstructure, characterised by coarse columnar PAGs, pronounced microsegregation, retained austenite, Laves phase and Ti-rich inclusions.
- Standard solution treatment refined grains but failed to remove microsegregation or retained austenite due to limited Ni and Ti diffusion, whereas wrought C465 achieved chemical homogeneity with negligible retained austenite.
- Homogenisation at 1200°C for 4 h, guided by DICTRA simulations, substantially reduced microsegregation and eliminated retained austenite and Laves particles, but significant grain coarsening occurred.
- Ageing after SST only partially promoted martensite-to-austenite reversion in WAAM C465, likely driven by the pre-existing retained austenite and Ni segregation. In contrast, ageing after homogenisation and SST suppressed this reversion, yielding a microstructure comparable to wrought C465.
- Electrochemical tests showed that non-homogenised WAAM_SST + A condition had the lowest pitting resistance and poorest repassivation, whereas homogenisation step prior to SST and ageing (WAAM_H + SST + A) enhanced pitting resistance to levels comparable with the fully heat-treated wrought C465.
- Pits initiated preferentially along Cr-depleted inter-dendritic austenite and at Laves/matrix interfaces in WAAM C465, while larger TiN inclusions dominated pit nucleation in wrought C465.
- Although homogenisation at 1200°C effectively mitigates microsegregation, the need for such high-temperature treatment and subsequent quenching may limit industrial feasibility for large WAAM components due to cost, thermal stress, and distortion risks. These findings nonetheless provide a transferable methodology—DICTRA-guided heat treatments combined with CPP-ECN benchmarking, that can be applied to qualify WAAM-printed martensitic stainless steel.
- Future work should explore lower-temperature or shorter-time homogenisation schedules, inter-pass thermal management, and alloy/

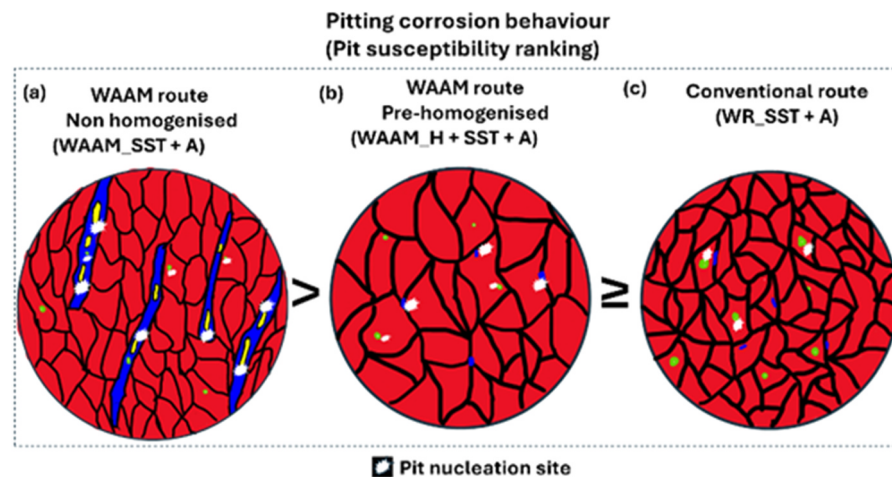


Fig. 16. Schematic linking manufacturing route and heat treatment condition to the corresponding pitting corrosion performance and susceptibility ranking of C465 steels. The colour coding follows the same legend as in Fig. 15.

process modifications to reduce microsegregation and Laves formation, thereby minimising reliance on extreme post-processing. Given that pitting often precedes SCC in chloride environments, improving pitting resistance through microstructural control is a critical step toward enhancing overall durability. Future studies should extend these findings to evaluate SCC susceptibility under service-relevant conditions.

CRediT authorship contribution statement

Mozumder Y. H.: Writing – review & editing, Writing – original draft, Visualization, Validation, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **F. Scenini:** Writing – review & editing, Supervision, Funding acquisition, Conceptualization. **Pickering E. J.:** Writing – review & editing, Supervision, Funding acquisition, Conceptualization. **Prangnell P. B.:** Writing – review & editing, Supervision, Funding acquisition, Conceptualization. **S. Williams:** Writing – review & editing, Methodology. **J. Wang:** Writing – review & editing, Methodology. **Y. Lu:** Writing – review & editing, Methodology. **V. Bongiorno:** Writing – review & editing, Visualization, Investigation, Formal analysis. **M. Taylor:** Writing – review & editing, Visualization, Investigation, Formal analysis.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Appendix A. Supporting information

Supplementary data associated with this article can be found in the online version at [doi:10.1016/j.addma.2026.105196](https://doi.org/10.1016/j.addma.2026.105196).

Data availability

Data will be made available on request.

References

- [1] B. Blakey-Milner, P. Gradl, G. Snedden, M. Brooks, J. Pitot, E. Lopez, M. Leary, F. Berto, A. du Plessis, Metal additive manufacturing in aerospace: a review, *Mater. Des.* 209 (2021) 110008, <https://doi.org/10.1016/j.matdes.2021.110008>.
- [2] R. Huang, M. Riddle, D. Graziano, J. Warren, S. Das, S. Nimbalkar, J. Cresko, E. Masanet, Energy and emissions saving potential of additive manufacturing: the case of lightweight aircraft components, *J. Clean. Prod.* 135 (2016) 1559–1570, <https://doi.org/10.1016/j.jclepro.2015.04.109>.
- [3] S.W. Williams, F. Martina, A.C. Addison, J. Ding, G. Pardal, P. Colegrove, Wire + arc additive manufacturing, *Mater. Sci. Technol.* 32 (2016) 641–647, <https://doi.org/10.1179/1743284715Y.0000000073>.
- [4] W.M. Garrison, Ultrahigh-strength steels for aerospace applications, *JOM* 42 (1990) 20–24, <https://doi.org/10.1007/BF03220942>.
- [5] K.F. Best, High strength materials for aircraft landing gear, *Aircr. Eng. Aerosp. Technol.* 58 (1986) 14–24, <https://doi.org/10.1108/eb036313>.
- [6] Lakshmi Gopal, Tirumalai Sudarshan, Cadmium replacement in coatings, *Surf. Eng.* 41 (2025) 295–301, <https://doi.org/10.1177/02670844251336526>.
- [7] B.T. Daymond, N. Binot, M.L. Schmidt, S. Preston, R. Collins, A. Shepherd, Development of Custom 465® corrosion-resisting steel for landing gear applications, *J. Mater. Eng. Perform.* 25 (2016) 1539–1553, <https://doi.org/10.1007/s11665-015-1830-5>.
- [8] Carpenter Technology, Custom 465®, (2022). (<https://www.carpentertechnology.com/alloy-finder/custom-465>) (accessed November 13, 2025).
- [9] SAE International, Steel, Corrosion-Resistant, Bars, Wire, and Forgings 12Cr - 11Ni - 1.7Ti - 1Mo Vacuum Induction Plus Vacuum Consumable Electrode Remelted Solution Heat Treated, Precipitation Hardenable, (2012). <https://doi.org/10.4271/AMS5936C>.
- [10] Carpenter Technology, C465 Datasheet, (n.d.). (https://www.carpentertechnology.com/hubfs/Data%20Sheets/Custom_465_Datasheet.pdf) (accessed November 15, 2025).
- [11] H. Zhang, X. Ji, D. Ma, M. Tong, T. Wang, B. Xu, M. Sun, D. Li, Effect of aging temperature on the austenite reversion and mechanical properties of a Fe-10Cr-10Ni cryogenic maraging steel, *J. Mater. Res. Technol.* 11 (2021) 98–111, <https://doi.org/10.1016/j.jmrt.2020.12.096>.
- [12] L. Wang, C. Dong, J. Yao, Z. Dai, C. Man, Y. Yin, K. Xiao, X. Li, The effect of η -Ni₃Ti precipitates and reversed austenite on the passive film stability of nickel-rich Custom 465 steel, *Corros. Sci.* 154 (2019) 178–190, <https://doi.org/10.1016/j.corsci.2019.04.016>.
- [13] Y. Lian, M. Ma, J. Zhang, J. Huang, W. Gao, Z. Zhang, C. Zhao, Influence of austenitizing temperature on the microstructure and mechanical properties of an Fe-Cr-Ni-Mo-Ti maraging stainless steel, *J. Mater. Eng. Perform.* 28 (2019) 5466–5475, <https://doi.org/10.1007/s11665-019-04279-6>.
- [14] S. Ifergane, E. Sabatani, B. Carmeli, Z. Barkay, V. Ezersky, O. Beeri, N. Eliaz, Hydrogen diffusivity measurement and microstructural characterization of Custom 465 stainless steel, *Electrochim. Acta* 178 (2015) 494–503, <https://doi.org/10.1016/j.electacta.2015.08.016>.

- [15] S. Ifergane, M. Pinkas, Z. Barkay, E. Brosh, V. Ezersky, O. Beeri, N. Eliaz, The relation between aging temperature, microstructure evolution and hardening of Custom 465® stainless steel, *Mater. Charact.* 127 (2017) 129–136, <https://doi.org/10.1016/j.matchar.2017.02.023>.
- [16] X. Xu, S. Ganguly, J. Ding, S. Guo, S. Williams, F. Martina, Microstructural evolution and mechanical properties of maraging steel produced by wire+arc additive manufacture process, *Mater. Charact.* 143 (2018) 152–162, <https://doi.org/10.1016/j.matchar.2017.12.002>.
- [17] K. Treutler, V. Wesling, The current state of research of wire arc additive manufacturing (WAAM): a review, *Appl. Sci.* 11 (2021), <https://doi.org/10.3390/app11188619>.
- [18] Z. Liu, Z. Wei, X. Zou, E. Brodu, D. Said, C. Furtado, J. Xie, K. Vanmeensel, Microstructural evolution and mechanical behavior of Custom 465 precipitation hardening stainless steel fabricated via laser powder bed fusion, *Mater. Sci. Eng. A* 892 (2024) 146069, <https://doi.org/10.1016/j.msea.2023.146069>.
- [19] Z. Sun, B. Sun, V. Soh, C. Lee, D. Lau, F. Wei, A.K. da Silva, M. Lin, C.C. Tan, P. Wang, U. Ramamurty, Laser powder bed fusion of crack-susceptible stainless maraging steel undergoing solid-state phase transformations, *Acta Mater.* 263 (2024) 119534, <https://doi.org/10.1016/j.actamat.2023.119534>.
- [20] W.E. Frazier, Metal additive manufacturing: a review, *J. Mater. Eng. Perform.* 23 (2014) 1917–1928, <https://doi.org/10.1007/s11665-014-0958-z>.
- [21] E.J. Schindelholz, M.A. Melia, J.M. Rodelas, Corrosion of additively manufactured stainless steels—process, structure, performance: a review, *Corrosion* 77 (2021) 484–503, <https://doi.org/10.5006/3741>.
- [22] M. Laleh, A.E. Hughes, W. Xu, I. Gibson, M.Y. Tan, A critical review of corrosion characteristics of additively manufactured stainless steels, *Int. Mater. Rev.* 66 (2021) 563–599, <https://doi.org/10.1080/09506608.2020.1855381>.
- [23] M.H. Ghoncheh, A. Shahriari, N. Birbilis, M. Mohammadi, Process-microstructure-corrosion of additively manufactured steels: a review, *Crit. Rev. Solid State Mater. Sci.* 49 (2024) 607–717, <https://doi.org/10.1080/10408436.2023.2255616>.
- [24] K. Hariharan, X. Guo, H.-L. Huang, N. Sridhar, J. Srinivasan, J. Hwang, G. S. Frankel, E.J. Schindelholz, Inter-melt pool corrosion and repassivation of SS316L stainless steel processed by laser powder bed fusion, *Corros. Sci.* 226 (2024) 111668, <https://doi.org/10.1016/j.corsci.2023.111668>.
- [25] A. Hemmasian Ettefagh, S. Guo, J. Raush, Corrosion performance of additively manufactured stainless steel parts: a review, *Addit. Manuf.* 37 (2021) 101689, <https://doi.org/10.1016/j.addma.2020.101689>.
- [26] K. Wang, Q. Chao, M. Annasamy, P.D. Hodgson, S. Thomas, N. Birbilis, D. Fabijanic, On the pitting behaviour of laser powder bed fusion prepared 316L stainless steel upon post-processing heat treatments, *Corros. Sci.* 197 (2022) 110060, <https://doi.org/10.1016/j.corsci.2021.110060>.
- [27] A. Vahedi Nemani, M. Ghaffari, S. Salahi, A. Nasiri, On the microstructural characteristics and corrosion performance of as-printed and heat-treated PH 13–8Mo martensitic stainless steel fabricated by wire arc additive manufacturing, *Mater. Today Commun.* 34 (2023) 105477, <https://doi.org/10.1016/j.mtcomm.2023.105477>.
- [28] M.R. Stoudt, R.E. Ricker, E.A. Lass, L.E. Levine, Influence of postbuild microstructure on the electrochemical behavior of additively manufactured 17-4 PH stainless steel, *JOM* 69 (2017) 506–515, <https://doi.org/10.1007/s11837-016-2237-y>.
- [29] L.K. Zhu, Y. Yan, L.J. Qiao, A.A. Volinsky, Stainless steel pitting and early-stage stress corrosion cracking under ultra-low elastic load, *Corros. Sci.* 77 (2013) 360–368, <https://doi.org/10.1016/j.corsci.2013.08.028>.
- [30] J.O. Andersson, T. Helander, L. Höglund, P. Shi, B. Sundman, Thermo-Calc & DICTRA, computational tools for materials science, *Calphad Comput. Coupling Phase Diagr. Thermochem* 26 (2002) 273–312, [https://doi.org/10.1016/S0364-5916\(02\)00037-8](https://doi.org/10.1016/S0364-5916(02)00037-8).
- [31] M.G. Pujar, U. Kamachi Mudali, S.S. Singh, Electrochemical noise studies of the effect of nitrogen on pitting corrosion resistance of high nitrogen austenitic stainless steels, *Corros. Sci.* 53 (2011) 4178–4186, <https://doi.org/10.1016/j.corsci.2011.08.027>.
- [32] G.S. Frankel, Pitting corrosion of metals: a review of the critical factors, *J. Electrochem. Soc.* 145 (1998) 2186–2198, <https://doi.org/10.1149/1.1838615>.
- [33] J.J. Perdomo, P.M. Singh, Electrochemical noise (Ecn) measurements as a corrosion monitoring tool: a review, *Corros. Rev.* 20 (2002) 359–378, <https://doi.org/10.1515/CORRREV.2002.20.4.359>.
- [34] E. Scheil, Bemerkungen zur Schichtkristallbildung 34 (1942) 70–72, <https://doi.org/10.1515/ijmr-1942-340303>.
- [35] R.C. Newman, M.A.A. Ajjawi, H. Ezuber, S. Turgoose, An experimental confirmation of the pitting potential model of galvele, *Corros. Sci.* 28 (1988) 471–477, [https://doi.org/10.1016/0010-938X\(88\)90069-8](https://doi.org/10.1016/0010-938X(88)90069-8).
- [36] N.J. Laycock, R.C. Newman, Localised dissolution kinetics, salt films and pitting potentials, *Corros. Sci.* 39 (1997) 1771–1790, [https://doi.org/10.1016/S0010-938X\(97\)00049-8](https://doi.org/10.1016/S0010-938X(97)00049-8).
- [37] J.R. Galvele, Transport processes and the mechanism of pitting of metals, *J. Electrochem. Soc.* 123 (1976) 464, <https://doi.org/10.1149/1.2132857>.
- [38] R.A. Cottis, Interpretation of electrochemical noise data, *Corrosion* 57 (2001) 265–285, <https://doi.org/10.5006/1.3290350>.
- [39] B.G. Pound, Potentiostatic pulse technique to determine the efficiency of hydrogen absorption by alloys, *J. Appl. Electrochem* 21 (1991) 967–972, <https://doi.org/10.1017/BF01077581>.
- [40] V. Vignal, V. Richoux, E. Suzon, S. Thiébaud, K. Tabaleiv, The use of potentiostatic pulse testing to study the corrosion behavior of welded stainless steels in sodium chloride solution, *Mater. Des.* 88 (2015) 186–195, <https://doi.org/10.1016/j.matdes.2015.09.001>.
- [41] C.Y. Chou, N.H. Pettersson, A. Durga, F. Zhang, C. Oikonomou, A. Borgenstam, J. Odqvist, G. Lindwall, Influence of solidification structure on austenite to martensite transformation in additively manufactured hot-work tool steels, *Acta Mater.* 215 (2021) 117044, <https://doi.org/10.1016/j.actamat.2021.117044>.
- [42] M. Taylor, J. Wang, Y.H. Mozumder, S.W. Williams, F. Scenini, P.B. Prangnell, E. J. Pickering, Microstructure engineering during wire-arc additive manufacturing of high-strength martensitic steels, *Metall. Mater. Trans. A* (2025), <https://doi.org/10.1007/s11661-025-07988-x>.
- [43] J.R. Kennedy, A.E. Davis, A.E. Caballero, S. Williams, E.J. Pickering, P. B. Prangnell, The potential for grain refinement of Wire-Arc Additive Manufactured (WAAM) Ti-6Al-4V by ZrN and TiN inoculation, *Addit. Manuf.* 40 (2021) 101928, <https://doi.org/10.1016/j.addma.2021.101928>.
- [44] P. Kürsteiner, M.B. Wilms, A. Weisheit, B. Gault, E.A. Jäggle, D. Raabe, High-strength Damascus steel by additive manufacturing, *Nature* 582 (2020) 515–519, <https://doi.org/10.1038/s41586-020-2409-3>.
- [45] P. Kürsteiner, P. Barriobero-Vila, P. Bajaj, F. De Geuser, M.B. Wilms, E.A. Jäggle, D. Raabe, Designing an Fe-Ni-Ti maraging steel tailor-made for laser additive manufacturing, *Addit. Manuf.* 73 (2023), <https://doi.org/10.1016/j.addma.2023.103647>.
- [46] A. Santana, A. Eres-Castellanos, J.D. Poplawsky, D. San-Martin, J.A. Jimenez, E. Urones-Garrote, A.J. Clarke, C. Capdevila, F.G. Caballero, Solute and phase heterogeneous distribution at different scales and its effect on ageing physical phenomena in a laser powder bed fusion produced maraging steel, *Addit. Manuf.* 94 (2024), <https://doi.org/10.1016/j.addma.2024.104494>.
- [47] S.A. David, J.M. Vitek, Correlation between solidification parameters and weld microstructures, *Int. Mater. Rev.* 34 (1989) 213–245, <https://doi.org/10.1179/imr.1989.34.1.213>.
- [48] F. Wang, W. Chen, D. Wang, H. Hou, Y. Zhao, Phase-field modeling and Experimental investigation for rapid solidification in wire and arc additive manufacturing, *J. Mater. Res. Technol.* 28 (2024) 4585–4599, <https://doi.org/10.1016/j.jmrt.2024.01.021>.
- [49] Y. Ji, H. Tang, P. Lan, C. Shang, J. Zhang, Effect of dendritic morphology and central segregation of billet castings on the microstructure and mechanical property of hot-rolled wire rods, *Steel Res. Int.* 88 (2017) 1600426, <https://doi.org/10.1002/srin.201600426>.
- [50] E.A. Jäggle, Z. Sheng, P. Kürsteiner, S. Ocylok, A. Weisheit, D. Raabe, Comparison of maraging steel micro- and nanostructure produced conventionally and by laser additive manufacturing, *Mater. (Basel)* 10 (2017), <https://doi.org/10.3390/ma10010008>.
- [51] M. Taylor, Y.H. Mozumder, A.D. Smith, A.E. Davis, F. Scenini, P.B. Prangnell, E. J. Pickering, In-Situ EBSD Study of Austenitisation in a Wire-Arc Additively Manufactured High-Strength Steel, *IOP Conf. Ser. Mater. Sci. Eng.* 1310 (2024) 12001, <https://doi.org/10.1088/1757-899X/1310/1/012001>.
- [52] G. Ressel, D. Brandl, T. Hönigsmann, M. Lukas, A. Stark, C. Gruber, S. Lukas, M. Stockinger, E. Kozeschnik, Microstructural refinement by spontaneous recrystallization without prior deformation of a 15-5 PH steel alloy and its mechanism, *Mater. Des.* 234 (2023) 112370, <https://doi.org/10.1016/j.matdes.2023.112370>.
- [53] S. Luo, W. Huang, H. Yang, J. Yang, Z. Wang, X. Zeng, Microstructural evolution and corrosion behaviors of Inconel 718 alloy produced by selective laser melting following different heat treatments, *Addit. Manuf.* 30 (2019) 100875, <https://doi.org/10.1016/j.addma.2019.100875>.
- [54] T. Allam, K.G. Pradeep, P. Köhnen, A. Marshall, J.H. Schleifenbaum, C. Haase, Tailoring the nanostructure of laser powder bed fusion additively manufactured maraging steel, *Addit. Manuf.* 36 (2020), <https://doi.org/10.1016/j.addma.2020.101561>.
- [55] Y. Yao, L. Fan, R. Ding, C. Franke, Z. Yang, W. Liu, T. Li, H. Chen, On the role of cellular microstructure in austenite reversion in selective laser melted maraging steel, *J. Mater. Sci. Technol.* 184 (2024) 180–194, <https://doi.org/10.1016/j.jmst.2023.10.032>.
- [56] A.L.M. Feitosa, G.G. Ribamar, J. Escobar, R. Sonkusare, T. Boll, F. Coury, J. Ávila, J.P. Oliveira, A.F. Padilha, Precipitation and reverted austenite formation in maraging 350 steel: competition or cooperation? *Acta Mater.* 270 (2024) 119865, <https://doi.org/10.1016/j.actamat.2024.119865>.
- [57] M. Ali, A. Khosravifard, A. Järvenpää, J. Kömi, A. Hamada, Optimizing microstructure and mechanical properties in additively manufactured 18Ni (300) maraging steel by controlling strengthening agents, *Mater. Des.* 254 (2025), <https://doi.org/10.1016/j.matdes.2025.114114>.
- [58] A. Keogh, E. Aradi, M. Taylor, A. Cook, E. Pickering, Z. Obadia, P. Prangnell, F. Scenini, Effect of ageing temperature on the microstructure and localised corrosion of 15-5PH stainless steel, *Corros. Sci.* 248 (2025) 112800, <https://doi.org/10.1016/j.corsci.2025.112800>.
- [59] A. Shahriari, M. Ghaffari, L. Khaksar, A. Nasiri, A. Hadadzadeh, B.S. Amirkhiz, M. Mohammadi, Corrosion resistance of 13wt% Cr martensitic stainless steels: Additively manufactured CX versus wrought Ni-containing AISI 420, *Corros. Sci.* 184 (2021) 109362, <https://doi.org/10.1016/j.corsci.2021.109362>.
- [60] M. Liu, B. Liu, Z. Ni, C. Du, X. Li, Elucidating the effect of titanium alloying on the pitting corrosion of ferritic stainless steel, *J. Mater. Res. Technol.* 28 (2024) 1247–1262, <https://doi.org/10.1016/j.jmrt.2023.12.032>.
- [61] J. Stewart, D.E. Williams, The initiation of pitting corrosion on austenitic stainless steel: on the role and importance of sulphide inclusions, *Corros. Sci.* 33 (1992) 457–474, [https://doi.org/10.1016/0010-938X\(92\)90074-D](https://doi.org/10.1016/0010-938X(92)90074-D).
- [62] C.A.D. Rodrigues, R.M. Bandeira, B.B. Duarte, G. Tremiliosi-Filho, V. Roche, A. M. Jorge, Effect of titanium nitride (TiN) on the corrosion behavior of a supermartensitic stainless steel, *Mater. Corros.* 70 (2019) 28–36, <https://doi.org/10.1002/maco.201810289>.



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