



Full Length Article

Effect of template δ -ferrite microstructures on the ferrite-to-austenite phase transformation and interphase boundary character distributions in additively manufactured duplex stainless steels

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ABSTRACT

Due to the high cooling rate, laser powder bed fusion of 2205 duplex stainless steels results in non-equilibrium microstructures with $>98\%$ δ -ferrite. A closer-to-equilibrium austenite fraction can be obtained through a solid-state phase transformation during a short heat treatment. However, an understanding of how the microstructures of the solidified δ -ferrite control this phase transformation is currently missing. Here, we discuss the δ -ferrite to austenite phase transformation in two distinct as-solidified microstructures using three-dimensional electron back-scatter diffraction, uncovering complex morphologies and interface crystallography that control the properties. Four types of austenite (intergranular, instability-induced, sympathetic, and intragranular) are formed and their interphase habit plane and curvature distributions after the heat treatment are studied. We show that the transformed microstructure will be different if crystallographic texture, dislocations, and precipitates in the δ -ferrite are altered by tuning the printing parameters. According to the grain boundary plane and curvature distributions, changes in δ -ferrite microstructure shifts the driving force for the intergranular transformation path from interfacial energy minimization towards strain energy minimization. The material printed with a reduced laser power and scan speed has lower dislocation density in the as-solidified state, which promotes a higher fraction of Kurdjumov-Sachs and Nishiyama-Wassermann δ -ferrite/austenite interfaces. This microstructure exhibits a weaker δ -ferrite texture, which limits the planar growth of austenite allotriomorphs and redirects thermal energy towards developing protrusions. These findings highlight the capability of additive manufacturing as a tool for microstructure engineering, via influencing the phase transformation product and the interfaces formed.

1. Introduction

Additive manufacturing (AM) offers unprecedented opportunities for the production of net-shaped or near net-shaped components with complex geometries [1,2]. Among various metal AM technologies, laser powder bed fusion (LPBF) has already been implemented in various industries due to its capability to process a wide range of metallic alloys [3]. The digitally controlled layer-by-layer manufacturing process provides unparalleled design freedom, and significantly reduces raw material wastage compared to conventional manufacturing routes based on subtractive strategies [1,2]. Nevertheless, the highly localized melting and solidification events as well as the cyclic rapid heating and cooling

during LPBF result in steep thermal gradients and complex thermal cycles, potentially impacting each material voxel differently [1,4]. This encourages the formation of highly non-equilibrium microstructures with dispersed precipitates, high dislocation densities, and sharp crystallographic texture in the as-built state [5–7]. The complex effects of non-equilibrium microstructures are significant in alloy grades that go through subsequent solid-state phase transformations during and/or after AM in particular, for instance, steels that undergo δ -ferrite to austenite and/or austenite to bainite/ferrite transformations [8–10].

Duplex stainless steels (DSSs) are known for their exceptional mechanical (strength and toughness) properties and corrosion resistance, which are achieved through the synergy of the properties of austenite

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and δ -ferrite (referred to as ferrite in the following). Thus, they are widely used in harsh environments, for instance, in oil & gas infrastructures and marine applications [11]. In fundamental research on phase transformations, DSSs are suitable model systems for studying ferrite-to-austenite phase transformation pathways [12], with the added benefit of an almost entirely ferritic as-LPBF state that is not available in conventional processing routes. DSS can be used to provide insights on how different transformation routes affect the microstructural and interfacial characteristics. During solid-state phase transformations, fine and geometrically complicated morphologies form, such as austenite protrusions inter-penetrated and/or inclined in three dimensions (3D).

3D electron back-scatter diffraction (EBSD) has been shown to be capable of capturing the complex spatial distribution of morphological and crystallographic information that is required to better understand the various phase transformation pathways in 3D printed DSSs, e.g., as demonstrated in a recent work on LPBF 2205 DSS by the current authors [13]. Such information provides insights into the different driving forces for phase transformations [14–17]. The large amount of data captured by 3D-EBSD also offers vast opportunities to explore materials behaviors focusing on different aspects [18–20]. The significant grain refinement and a series of austenite deformation mechanisms achieved in heat-treated LPBF manufactured DSS 2205 have been shown to contribute to its excellent combination of mechanical properties, and this has been related to the ferrite to austenite phase transformation [21]. However, there is currently no systematic 3D study available elucidating how the change in AM parameters and thus non-equilibrium microstructural characteristics of the as-built state will alter the subsequent solid-state phase transformation pathways and, thus, impact characteristics such as the 3D grain morphology and interface character during and/or after a post-AM thermal treatment.

Previous studies have demonstrated that printing parameters such as laser power and scan speed significantly impact the texture [22], precipitate formation [23–25], and dislocation density [26–28] of LPBF processed components. These non-equilibrium microstructures have shown substantial impacts on the morphology and crystallography of phases and interfaces formed during post-AM heat treatments, presenting opportunities for microstructure engineering. The nature of dislocations and precipitates in the parent matrix control the character of the interphase boundary formed during phase transformation, primarily by influencing interface structures [29]. It has been reported that nucleation on dislocations and/or precipitates that have coherent interfaces with the parent matrix phase usually leads to the formation of rational interfaces upon phase transformations [29]. In an interstitial free steel, texture has been shown to affect the grain boundary character distribution of grain boundaries [30]. The significant impact of the thermal history on the interface evolution has been demonstrated in different alloy systems with various crystal structures including face-centered-cubic (FCC) [31], hexagonal close packed [32], and body-centered-cubic (BCC) [12]. However, there is little information available concerning how the change in AM process parameters alters the FCC-BCC interfaces formed during LPBF and solid-state phase transformations. In more detail, there is a knowledge gap regarding how certain characteristics of the non-equilibrium microstructures affect the solid-state phase transformation from ferrite to austenite. Filling these gaps through systematic studies of the ferrite-austenite interface character will contribute towards the control of solid-state phase transformation pathways and products, via tuning crystallographic texture, dislocation structures, and precipitation in the ferrite matrix.

This study compares two distinct DSS microstructures in their as-LPBF and heat-treated states based on 3D orientation datasets collected using serial sectioning and EBSD. One of them was introduced in our recent work [13] aimed at revealing the various types of austenite morphology (intergranular, instability-induced, sympathetic, and intragranular), and their interphase habit plane and curvature distributions. Comparing this dataset with a new dataset, provides the opportunity to unravel how the phase transformation pathways,

microstructure morphology, and the interface crystallography are controlled by crystallographic texture, and characteristics of dislocations and precipitates in the δ -ferrite which, in turn, can be tailored through adjustments to AM parameters. Considering the formerly reported impacts of ferrite-austenite interface character distributions on hot working [33] and deformation mechanisms [34] in these classes of steels, such knowledge would ultimately guide the use of additive manufacturing as a tool for microstructural engineering. This can be achieved not only in the as-built state, but also upon subsequent thermal processing and solid-state phase transformation, to unlock desirable properties.

2. Experimental

The DSS 2205 powder with a chemical composition of 22.60 Cr, 5.90 Ni, 0.60 Si, 0.18 N, 3.20 Mo, 1.10 Mn, 0.02 C, 0.01 S, 0.02 P (in wt. %), and a balance of Fe was obtained from Sandvik Osprey Ltd. The powder particles were mainly of spherical morphology and exhibited diameters of 15 – 53 μm based on measurements by sieve and laser diffraction. An SLM Solution 125^{HL} machine furnished with a 400 W continuous wavelength fiber laser was utilized to print the samples under an Ar shielding atmosphere where the O_2 concentration was kept lower than 200 ppm. Sample AB200-700 was printed using the printing parameters of 200 W laser power, 700 mm/s scan speed, 60 μm hatch spacing, 30 μm layer thickness, and a 67° incremental rotation between layers. Laser power and scan speed were altered to 100 W and 300 mm/s for sample AB100-300, all other parameters were kept the same. A 316 L stainless steel plate served as a substrate and was pre-heated to 200 °C. Cubic samples with edge lengths of 15 mm were printed using LPBF. The porosity of AB200-700 and AB100-300 was 0.04 % and 0.08 %, respectively, as examined by optical microscopy. Based on our previous research [8,35], a post-LPBF thermal treatment in a tube furnace under Ar shielding gas was performed on selected samples at 1000 °C for 10 minutes, followed by water quenching to form a fine and more balanced duplex microstructure. The heat treated AB200-700 and AB100-300 samples were accordingly named HT200-700 and HT100-300.

2D-EBSD on as-built and heat-treated samples was performed in a JEOL 7001F scanning electron microscope coupled with a Hikari 31 Super EBSD Camera and operating at 20 kV, 6×6 binning, 13 nA probe current, and a step size of 0.1 – 1 μm . Data acquisition and processing were carried out using the TSL-OIM software. The 2D-EBSD specimens were extracted from the X-Z surface at the central areas along the sample BD and were subjected to a final fine polishing using 0.04 μm OPU colloidal silica. Scanning transmission electron microscopy (STEM) on as-built AB200-700 and AB100-300 samples was conducted in a JEOL JEM-F200 (S)TEM equipped with a cold field emission gun operated at 200 kV. STEM specimens were prepared using a standard plasma focused ion beam (FIB) lift-out method [36] in a Thermo Fisher Helios G4 Xe plasma FIB. Experimental convergent beam electron diffraction (CBED) patterns were acquired by tilting the samples to align with the [100] zone axis (ZA) and applying a convergence angle of 5 mrad. CBED patterns for ferrite were simulated using the JEMS (v 4.9) electron microscopy software. The simulations spanned a sample thickness range from 75 to 200 nm, with 2 nm increments. Higher order Laue zone deficiency lines were excluded [37]. The simulated patterns that most closely matched the experimental results were selected. These simulations were based on the ideal crystal structure of BCC iron. The dislocation density was quantified using the line intercept method [38].

Using the same ThermoFisher Helios G4 PFIB Xe⁺ DualBeam system, the 3D-EBSD datasets for the samples HT200-700 and HT100-300 were collected via block-face serial sectioning. Details of specimen preparation and setup can be found in [39]. The EBSD collection software Aztec4.0 and the Auto Slice and View software were employed for operational control. Rocking milling was performed to ensure minimal curtaining effects. The slice thickness was 100 nm while the EBSD data was collected on every other layer, yielding a step size of 200 nm. The

3D orientation map was established from parallel stacks of 2D EBSD orientation maps. The volume of the collected dataset was $\sim 100 \times 75 \times 92 \mu\text{m}^3$ and $\sim 100 \times 75 \times 71 \mu\text{m}^3$ for HT200-700 and HT100-300, respectively.

The DREAM.3D software [40] and MTEX toolbox in Matlab [41] were used to conduct 3D reconstruction and data analysis. The ParaView software [42] was used to generate visualizations. Generally, the reconstruction procedure includes alignment, clean-up, segmentation, and meshing. The readers are referred to [13] for the detailed description and parameters. Using a grain boundary misorientation threshold of 1° , the processed HT200-700 dataset contained approximately 70 million voxels and was segmented as 65,146 unique grains alongside 297,075 grain boundaries and 57,343,510 grain boundary mesh triangles. Using the same processing settings, the processed HT100-300 dataset contained approximately 57 million voxels and was segmented as 38,538 unique grains alongside 157,850 grain boundaries and 45,240,517 triangles.

The grain boundary (mean) curvature was measured locally for the triangles via executing the previously developed cubic-order algorithm detailed in [43] and implemented in DREAM.3D [40,44]. In this algorithm, a parabola fitting is applied on a patch of contiguous triangles surrounding the triangle of interest. The crystallographic properties of the surrounding triangle patch involving the 2nd and 3rd nearest neighbors of the triangle of interest are measured. After a least-squares method is used to perform parabola fitting, the Weingarten matrix of this triangle patch is acquired. The eigenvalues of the Weingarten matrix correspond to the principal curvatures k_1 and k_2 , hence the triangle mean curvature is $(k_1 + k_2)/2$, with a convex interface having a positive value, and a concave interface having a negative value. In this study, only the absolute values of the mean curvatures were considered during quantification of the grain boundary curvature distributions [14].

Upon completion of the 3D orientation data reconstruction, the microstructure is composed of discrete grains with unique ID numbers, that are wrapped by a triangular mesh. Every triangle within the mesh corresponds to information involving the grain IDs on two sides, the grain orientations on two sides, the disorientation across the triangle, area of the triangle, the surface normal, and the curvature. This enables local mapping of these characteristics and quantification of their distributions over morphological, volumetric, and crystallographic variables. Because the duplex microstructures in this work have two distinct phases, the reconstructed triangles/boundaries were grouped into three

different categories: ferrite-ferrite, ferrite-austenite, and austenite-austenite interfaces. Based on the austenite morphology, the ferrite-austenite interfaces were further classified into interfaces associated with ferrite boundaries neighboring intergranular, sympathetic, and intragranular austenite morphologies. More in-depth analyses were then conducted on each of these categories of boundaries to calculate the grain boundary plane and curvature distributions utilizing the approach outlined in [14]. The interphase plane character distribution for dissimilar phases was calculated separately, reflecting the ferrite and austenite habit planes.

3. Results

3.1. Microstructure evolution

3.1.1. Microstructures in as-built state

Fig. 1 illustrates typical as-built microstructures of the AB200-700 and AB100-300 LPBF DSS 2205 samples via 2D-EBSD maps, representing non-equilibrium phase fractions (Fig. 1a, e) and columnar grain structures (Fig. 1b, c, f, g) extending along the BD. From the EBSD phase maps in Fig. 1a, e, ferrite (colored in red) is seen to occupy the majority of areas, accounting for $>98\%$ of the mapped region in both AB200-700 and AB100-300, with occasional small austenite particles observed in green color as shown in the enlarged EBSD maps in Fig. 1s. The inverse pole figure (IPF) maps along the Y-direction (out-of-page) in Fig. 1b, f and the IPF maps along the BD in Fig. 1c, g show columnar grain structures elongated towards the BD. From the IPF map in Fig. 1g and the pole figures in Fig. 1h, ferrite does not display a sharp texture along the BD in AB100-300 as opposed to the strong $\langle 001 \rangle$ ferrite fiber texture developed in AB200-700 (Fig. 1c, d).

Fig. 2 displays typical dislocation structures and oxide dispersions in AB200-700 (Fig. 2a) and AB100-300 (Fig. 2b) observed in STEM along the BCC $[001]$ ZA. Insets show CBED patterns with comparable features in both conditions, implying a similar TEM specimen thickness in both AB200-700 and AB100-300. Hence, a valid comparison can be made regarding the morphology, density, and/or number of dislocations in these samples. High dislocation densities, dislocation entanglement, and amorphous Si-O dispersions with spherical morphology are observed in both samples. The dislocations align with each other and display a grid-like pattern in both samples. To quantify the dislocation density in each condition, the TEM foil thickness was estimated by comparing the

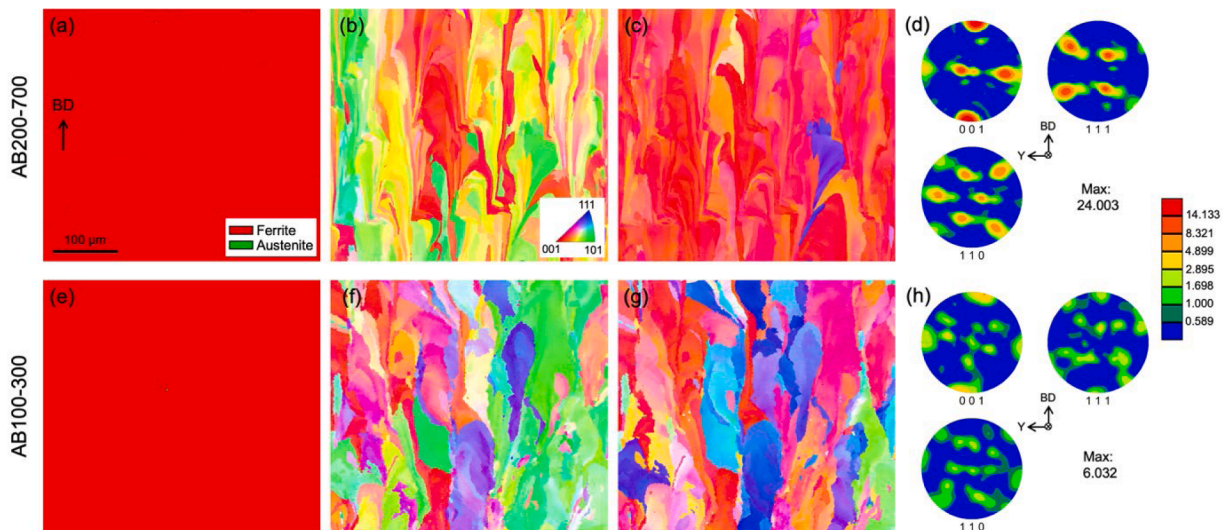


Fig. 1. Overall microstructures of the (a) – (c) AB200-700 and (e) – (g) AB100-300 LPBF DSS 2205 samples via 2D-EBSD. (a, e) EBSD phase maps illustrating $>98\%$ ferrite, with BD along the vertical direction. (b, f) IPF maps on the same areas along the Y-direction (out-of-page) demonstrating columnar grain structure in ferrite. (c, g) IPF maps along the BD showing a strong $\langle 100 \rangle // \text{BD}$ ferrite texture in AB200-700 and the absence of a sharp ferrite texture in AB100-300. (d, h) Ferrite pole figures collected on larger areas showing significantly higher texture intensities in AB200-700 compared to those in AB100-300.

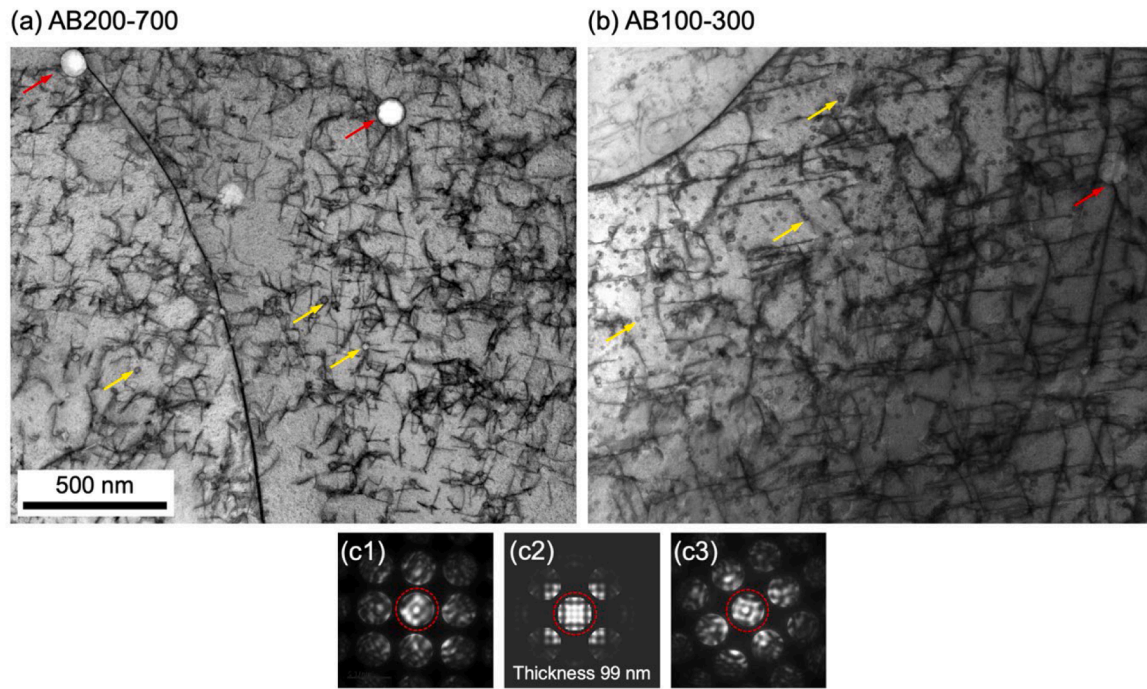


Fig. 2. STEM micrographs displaying dislocation structures and precipitates in ferrite along the [001] ZA in samples (a) AB200-700 and (b) AB100-300. Red arrows point to examples of large Si-O inclusions, while yellow arrows indicate examples of smaller Si-O inclusions. (c) CBED patterns along the ZA [100] obtained from (c1) AB200-700, (c2) simulation, and (c3) AB100-300, demonstrating comparable TEM specimen thicknesses for AB200-700 and AB100-300 at around 99 nm.

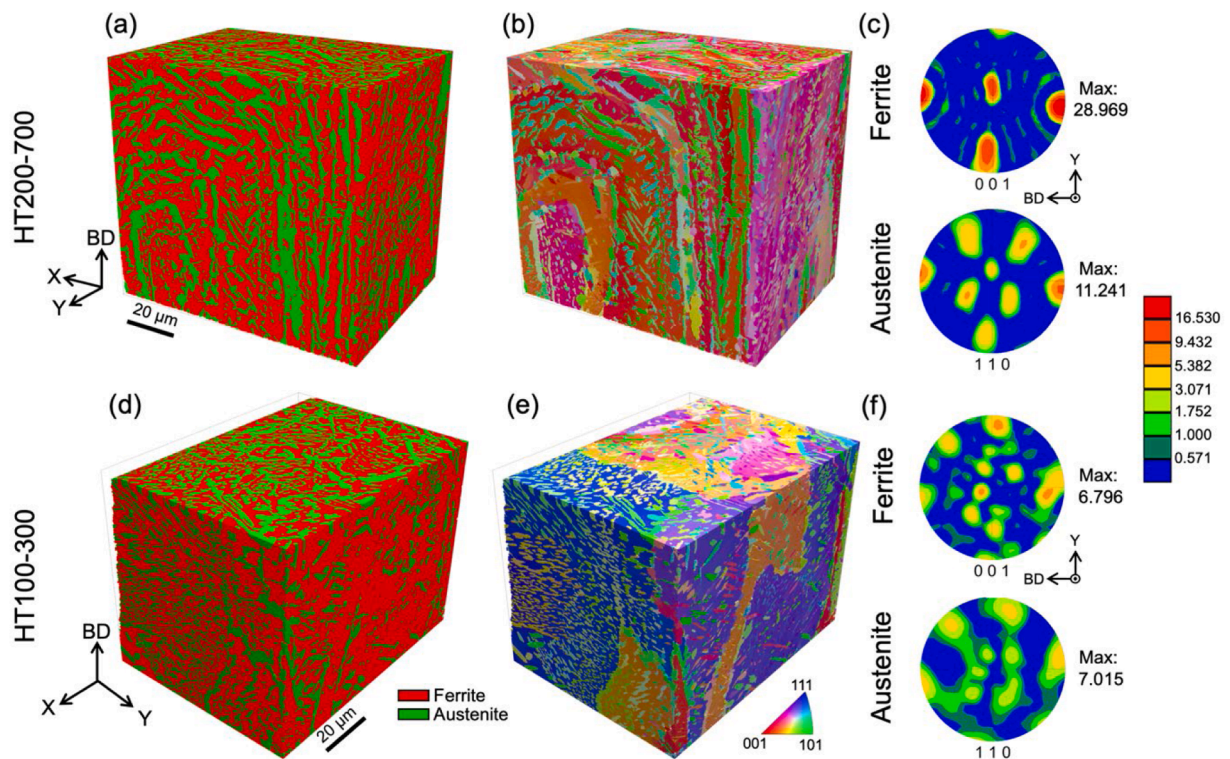


Fig. 3. Overall microstructures from the 3D-EBSD data of (a) – (b) HT200-700 and (d) – (e) HT100-300 samples in volumetric view. (a, d) EBSD phase maps displaying more balanced ferrite/austenite fractions, BD is along the vertical direction. (b, e) IPF map along the BD shown in the coordinate systems in (a, d), demonstrating a columnar grain structure of the parent ferrite phase besides fine austenite grains developed during the post-LPBF thermal treatment, and showing the presence of a texture in both ferrite and austenite in HT200-700 while the texture in both ferrite and austenite in HT100-300 is more random. (c, f) Pole figures for ferrite and austenite, displaying significantly stronger texture intensities in HT200-700 compared to HT100-300, and more closely correlated peaks between ferrite (001) and austenite (110) in HT100-300 compared to HT200-700. Pole figures were generated from large 2D EBSD scans to include significant numbers of unique grains.

simulated and experimental CBED patterns. For visual comparison between the observation and simulation, only the central disk (denoted by the dashed red circle) of the CBED pattern is utilized. The dislocation density is higher in AB200-700 ($2.25 \pm 0.43 \times 10^{14} \text{ m}^{-2}$) compared to that in AB100-300 ($1.72 \pm 0.29 \times 10^{14} \text{ m}^{-2}$) as measured using the line intercept method [38]. This represents a 30 % increase of dislocation density in AB200-700 compared to that in AB100-300. At least 15 lines were examined in each micrograph. Si-O precipitates with various sizes are also observed in both samples. In particular, Si-O inclusions with larger sizes as indicated by red arrows are more often noticed in AB200-700 (Fig. 2a), whereas finer inclusions as pointed out by yellow arrows are more frequently present in AB100-300 (Fig. 2b). Moreover, visual inspection indicates that AB100-300 seems to possess a much higher density of fine-sized Si-O inclusions compared to AB200-700.

3.1.2. Microstructures after the heat treatment

The AB200-700 and AB100-300 samples were then exposed to a heat treatment at 1000 °C for a duration of 10 min to allow nucleation and growth of austenite. Overall 3D microstructures of the thermally treated HT200-700 and HT100-300 conditions in volumetric view are illustrated in EBSD phase maps (Fig. 3a, d), IPF maps projected along the BD (Fig. 3b, e), and pole figures in the ferrite and austenite lattice frames (Fig. 3c, f), respectively. It should be noted that, the microstructure of HT200-700 was shown in a previous manuscript [13], it is shown here from a different perspective for comparison purpose. The volumetric microstructures displayed in EBSD IPF maps along the Y-direction (Fig. 2S(a and c)) and further pole figures in the ferrite and austenite lattice frames (Fig. 2S(b and d)) are illustrated in the Supplementary. As measured using the entire data volume, a 31/69 austenite/ferrite phase fraction is achieved for HT100-300 by post-LPBF heat treatment as presented in Fig. 3d, which contains a slightly lower austenite volume compared to that of HT200-700 (39/61 phase fraction) shown in Fig. 3a. In Fig. 3e and f, both ferrite and austenite in the HT100-300 exhibit random crystallographic orientations without any significant texture along neither the Y-direction nor the BD. This is opposed to the strong

$\langle 001 \rangle$ ferrite texture and $\langle 110 \rangle$ austenite texture along the BD as indicated in the pole figures for HT200-700 (Fig. 3c). There are 164 and 82 ferrite grains with a volume of $> 50 \mu\text{m}^3$ in the entire captured volume of HT200-700 and HT100-300 condition, respectively. Observing from the three exterior surfaces of the collected data volumes in Fig. 3a, b, d, e, austenite is noticed to be mainly found along the prior ferrite-ferrite interfaces with finer intragranular austenite particles dispersed in the ferrite grain interiors. However, it should be noted that part of these intragranular islands may have indeed nucleated on the ferrite-ferrite interfaces underneath the three exterior surfaces. In addition, they may exhibit different shapes, sizes, and aspect ratios in 3D. Therefore, 2D characterization is not sufficient to comprehensively categorize and analyze different modes of austenite nucleation and growth. This highlights the importance of 3D orientation datasets in facilitating the comprehensive and unambiguous classification and comparison of distinct types of austenite developed through different ferrite-to-austenite phase transformation pathways. As will be shown in the following, 3D-EBSD facilitates the understanding of correlations between AM parameters and subsequent solid-state phase transformations.

3.2. Daughter austenite types classified by phase transformation pathway

The classification of ferrite-to-austenite phase transformations in the HT200-700 and HT100-300 conditions is illustrated in Fig. 4, where austenite grains are depicted in IPF colors projected along the Y-direction. Here, austenite grains are grouped into their different types based on their nucleation and growth mechanisms. A detailed description of the austenite categorization criteria and relevant schematics can be found in [13]. Intergranular austenite grains displayed in Fig. 4a, d are those that nucleate and grow along prior ferrite-ferrite boundaries. It can be observed that the intergranular austenite contains abundant protrusions without identifiable grain boundaries between them and the allotriomorphs. The growth of these protrusions is driven by the instability mechanism [12]. On the other hand, austenite protrusions with

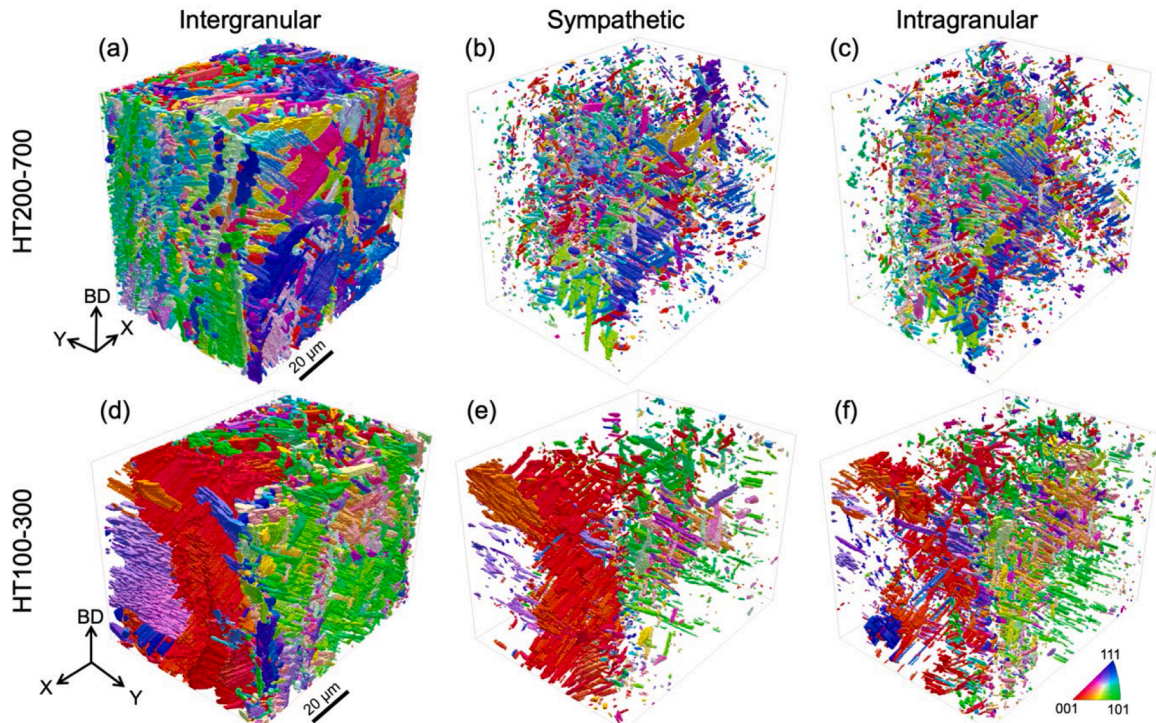


Fig. 4. Daughter austenite phase transformation products classified by different ferrite-to-austenite phase transformation pathways shown in IPF maps along the Y-direction. (a, d) Intergranular, (b, e) sympathetic, (c, f) intragranular austenite in (a) – (c) HT200-700 and (d) – (f) HT100-300.

low-angle grain boundaries (LAGBs) between them and intergranular austenite are characterized as being driven by the sympathetic nucleation as can be seen in Fig. 4e [45,46]. Lastly, intragranular austenite grains in Fig. 4f are those that do not touch any ferrite-ferrite interfaces and that do not establish an LAGB with any intergranular austenite [47]. It can be noticed that intergranular austenite (Fig. 4a, d) inhabits the highest volume fraction among all potential austenite types. Sympathetic (Fig. 4b, e) and intragranular (Fig. 4c, f) austenite occupy much lower volume fractions owing to their needle-shaped morphologies and smaller grain sizes. Among all austenite types in HT100-300, intergranular, sympathetic, and intragranular austenite occupy volume fractions of 86 %, 7 %, and 7 %, respectively, showing a slight decrease in intergranular volume and a slight increase in both sympathetic and intragranular volume compared to HT200-700 (89 %, 5 %, and 6 %, respectively, Table 1). At least five 3D-EBSD volumes were used to produce the statistics including error bars for each sample. Although present in comparatively low volume fractions, both sympathetic and intragranular austenite contribute significantly to grain refinement following the post-LPBF heat treatment. Their presence is crucial for achieving the desirable mechanical properties, as detailed in [21]. The growth of instability-induced protrusions is observed to appear much more pronounced in HT100-300 compared to HT200-700. Videos showing the full microstructure overviews and different austenite types in 3D volumes are included in the Supplementary.

Examples of representative grain structures for each austenite type are displayed in Fig. 5a–f for comparison. In the intergranular mechanism, the formation of austenite allotriomorphs is easily distinguishable from instability-induced protrusions in HT200-700 (Fig. 5a), whereas the significant instability growth in HT100-300 covers almost the entire allotriomorph (Fig. 5d). It can also be seen that instability-induced austenite protrusion growth in HT100-300 takes place when only minimal allotriomorph growth has occurred (Fig. 5d), as opposed to the substantial allotriomorph volume in HT200-700 (Fig. 5a). In both the sympathetic and intragranular mechanisms, austenite protrusions tend to extend over longer distances in HT200-700 (Fig. 3S), with a lower aspect ratio than those of HT100-300 (Table 2). The visual impression of higher number of longer sympathetic and intragranular protrusions in HT100-300 (Fig. 4b, c, e, f) could be explained by its significantly higher fraction of grains that fall between the lengths of 20 and 40 μm compared to HT200-700 (Fig. 3S). However, it is clear from Fig. 3S that HT200-700 displays higher number fraction of protrusions longer than 40 μm in both the sympathetic and intragranular transformation pathways.

3.3. Character of interfaces

3.3.1. Statistics and visualizations

The overall misorientation angle distributions across all types of interfaces in both HT200-700 and HT100-300 are plotted in Fig. 6a. To understand these interfaces in more detail, they were classified into two groups, namely austenite-ferrite, and austenite-austenite interfaces. Since most of the prior ferrite-ferrite grain boundaries were consumed by austenite nucleation and growth, further investigation was not conducted on these interfaces due to limited available interfacial areas, obscuring meaningful conclusions. The misorientation distributions for ‘all’ boundaries and ferrite-austenite boundaries in Fig. 6a and b both present a broad peak between 42° and 46° while the entire range spans from 1° to 62.8° . This suggests the potential existence of a

crystallographic orientation relationship (OR) at the ferrite-austenite boundaries. It also shows that the overall misorientation profiles for the post-LPBF thermally treated LPBF DSS 2205 samples are dominated by the ferrite-austenite boundaries, meaning that the majority of interfaces are the ones between ferrite and austenite. The misorientation distributions for austenite-austenite grain boundaries presented in Fig. 6c display a strong peak at 60° , implying that $\Sigma 3$ twin boundaries dominate within austenite grains. In addition, it can be observed that HT100-300 presents a larger peak at $42\text{--}46^\circ$ compared to HT200-700, implying a potentially stronger alignment with specific ORs in HT100-300.

The peak between 42° and 46° in the misorientation profiles for the ferrite-austenite boundaries (Fig. 6b) corresponds well to the misorientation angles predicted by the established OR models that characterize the BCC to FCC (and reverse) phase transformations. These include Kurdjumov-Sachs (K-S) [48], Nishiyama-Wasserman (N-W) [49], Greninger-Troiano [50], Pitsch [51], and Bain [52]. Among them, the K-S and N-W ORs (Table 3) are the most suitable models for the observed phase transformations in this work [12]. For further investigations, the interfaces were classified into three types, K-S, N-W, and Others, according to their misorientation angles and axes. To be grouped into the K-S or N-W classification, the angular deviation from the perfect misorientation angle and axis must be $<5^\circ$ for that specific OR. Interfaces within 5° of both ORs (which are separated by only 5.26°) are assigned to the OR with the minimum misorientation. All other interfaces were grouped into Others.

Fig. 7a–f illustrates the overall visualizations of the austenite-ferrite interfacial ORs in terms of austenite types including intergranular, sympathetic, and intragranular in HT200-700 (Fig. 7a–c) and HT100-300 (Fig. 7d–f). It should be noted that the microstructure of HT200-700 was shown in a previous manuscript [13], it is shown here from a different perspective for comparison purposes. Table 4 lists the ferrite-austenite boundary area fractions of different austenite types in HT200-700 and HT100-300. From Fig. 7a, d and Table 4, it is apparent that the majority of interphase boundaries in both conditions are associated with intergranular austenite, accounting for $>85\%$ of the total austenite-ferrite interfacial areas. In contrast, sympathetic and intragranular each occupies $<10\%$ of the interphase boundary areas in either condition (Table 4). Table 5 shows ferrite-austenite boundary area fractions of different OR categories under respective austenite types. Intergranular austenite displays a much higher area fraction of non-KS/NW interfaces compared to sympathetic and intragranular in both conditions. A consistently lower non-KS/NW interface area fraction is also seen in HT100-300, regardless of the austenite type (Table 5). Both sympathetic and intragranular austenite exhibit higher area fractions of K-S/N-W interfaces in HT100-300 compared to those in HT200-700. Interface crystallography such as variant selection and inter-variant selection was reported to affect hardness in the same class of DSSs as detailed in a previous study [53].

Using the 3D-EBSD datasets collected in this work and the approaches developed in [14,54], the interphase habit plane and curvature distributions are determined for boundaries corresponding to intergranular, sympathetic, and intragranular austenite in the crystal lattice frames of ferrite and austenite (Fig. 7a1–f1). When intersecting, the ferrite and austenite tend to terminate on the (110) and (111) planes, respectively, regardless of austenite type or sample. These planes have been reported to be energetically and crystallographically favorable for BCC and FCC crystal structures, respectively [55–58]. In both HT200-700 and HT100-300, the highest relative areas in both lattice frames are observed on phase interfaces belonging to the sympathetic austenite, followed by intragranular and intergranular austenite. When compared among the same austenite type and lattice frame, both the maximum relative areas of the habit plane distributions and curvatures are always larger in HT100-300 compared to those in HT200-700. Subjected to the same comparison, the minimum relative areas of the habit plane distributions are always lower in HT100-300. These

Table 1
Volume fractions of different austenite types in HT200-700 and HT100-300.

Austenite type	HT200-700	HT100-300
Intergranular	$89 \pm 1\%$	$86 \pm 3\%$
Sympathetic	$5 \pm 1\%$	$7 \pm 2\%$
Intragranular	$6 \pm 1\%$	$7 \pm 1\%$

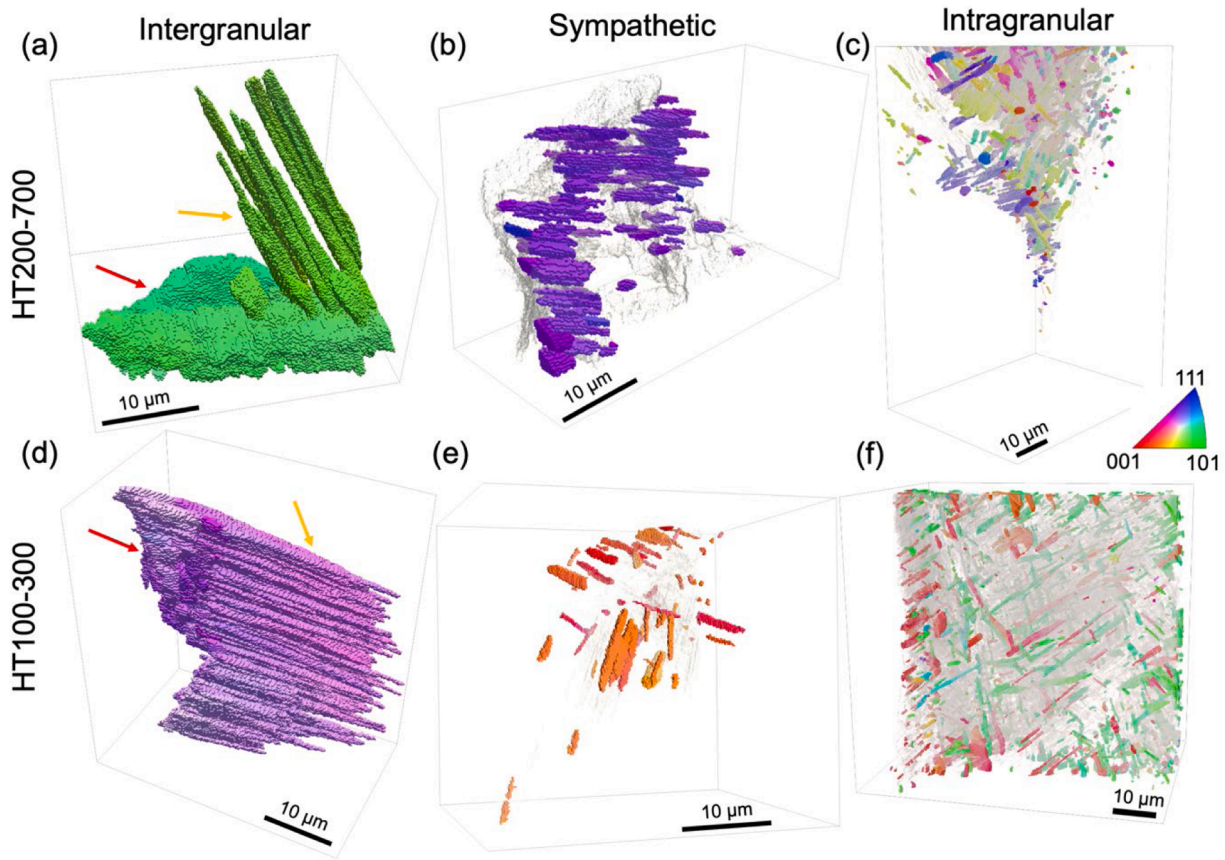


Fig. 5. Typical grain morphology of (a, d) intergranular, (b, e) sympathetic, and (c, f) intragranular austenite in sample (a) – (c) HT200-700 and (d) – (f) HT100-300 shown in IPF colors along the sample Y-direction. The allotriomorph and instability-induced protrusions within (a, d) intergranular type are pointed out by red and yellow arrows, respectively. The connected intergranular austenite and surrounding ferrite grains for (b, e) sympathetic and (c, f) intragranular types, respectively, are displayed in semi-transparent solid color.

Table 2

Average aspect ratio for sympathetic and intragranular austenite types in HT200-700 and HT100-300.

Austenite type	HT200-700	HT100-300
<i>Sympathetic</i>	$3.3 \pm 0.3 \times 10^{-2}$	$4.9 \pm 0.3 \times 10^{-2}$
<i>Intragranular</i>	$3.0 \pm 0.2 \times 10^{-2}$	$3.7 \pm 0.4 \times 10^{-2}$

distributions imply a stronger tendency for interfaces in HT100-300 to terminate on energetically and crystallographically favorable planes than those in HT200-700, despite exhibiting more curved interfaces compared to those in HT200-700. In addition, the grain boundary curvature distributions exhibit slightly different behaviors across different

austenite types and within different samples. In HT200-700, the maximum/minimum locations of the curvature distributions inversely correspond to the minimum/maximum locations of the plane distributions in both lattice reference frames for the phase interfaces belonging to the intergranular austenite type. However, the minima of the curvature distributions start to deviate away from the maxima of the plane distributions on phase interfaces belonging to sympathetic austenite in HT200-700. This deviation becomes stronger in the intragranular type, where the minimum curvatures in both ferrite and austenite lattice frames do not locate on the planes with maximum relative areas. In HT100-300, the minimum curvatures in both lattice frames are not located at planes with the maximum relative area, regardless of austenite type.

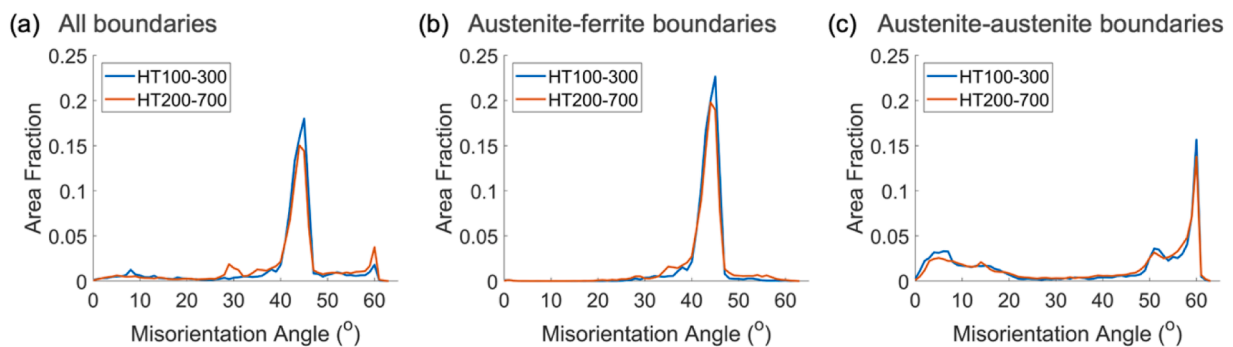


Fig. 6. Distribution of misorientation angles for (a) all boundaries (including austenite-ferrite, ferrite-ferrite, and austenite-austenite), (b) austenite-ferrite interfaces, and (c) austenite-austenite interfaces in HT200-700 and HT100-300.

Table 3

K-S and N-W orientation relationships between FCC and BCC defined by plane and direction parallelisms and misorientation angle-axis pairs.

Orientation relationship	Parallelism	Angle-axis pair
Kurdjumov-Sachs (K-S) [48]	$\{111\}_{\text{fcc}}//\{110\}_{\text{bcc}}$ $\langle 110 \rangle_{\text{fcc}}//\langle 111 \rangle_{\text{bcc}}$	$42.85^\circ \langle 0.968$ $0.178 \ 0.178 \rangle$
Nishiyama-Wassermann (N-W) [49]	$\{111\}_{\text{fcc}}//\{110\}_{\text{bcc}}$ $\langle 112 \rangle_{\text{fcc}}//\langle 110 \rangle_{\text{bcc}}$	$45.98^\circ \langle 0.976$ $0.083 \ 0.201 \rangle$

3.3.2. Interface visualizations by austenite type

Fig. 8 illustrates the interfaces between intergranular austenite and ferrite in HT200-700 (Fig. 8a–c) and HT100-300 (Fig. 8d–f), distinguished by unique interface colors. These interfaces are classified into three groups based on the austenite-ferrite ORs, i.e., K-S (Fig. 8a, d)), N-W (Fig. 8b, e), and Others (Fig. 8c, f). For intergranular austenite in HT200-700, a mix of allotriomorphic and instability-induced interfaces is present in all categories of ORs. Instability-induced interfaces here refer to the ferrite-austenite interfaces on the instability-induced austenite grains. For intergranular austenite in HT100-300, the majority of the undulating austenite-ferrite interfaces that correspond to instability-induced protrusions are of K-S/N-W ORs as shown in Fig. 8d and e. Nevertheless, a substantial number of austenite-ferrite interfaces for intergranular austenite still fall into the Others OR category, as can be seen in Fig. 8f.

Fig. 9 illustrates the interfaces between sympathetic austenite and ferrite in HT200-700 (Fig. 9a–c) and HT100-300 (Fig. 9d–f), distinguished by unique interface colors. These interfaces are classified into three groups based on the austenite-ferrite ORs of K-S (Fig. 9a, d), N-W (Fig. 9b, e), and Others (Fig. 9c, f). Fig. 10 shows the interfaces between intragranular austenite and ferrite in the same format as in Fig. 9. For both sympathetic and intragranular austenite in both conditions, needle-shaped boundaries are mostly categorized as K-S/N-W (Fig. 9a, b, d, e) and Fig. 10a, b, d, e), whereas the flatter boundaries mainly fall into the Others category (Figs. 9c, f) and 10c, f). Some needle-shaped interfaces are observed in the Others category of HT200-700 and concentrate near the lower corner in both sympathetic (Fig. 9c) and intragranular (Fig. 10c) types. This can be attributed to a sample edge effect, which creates artefacts by falsely categorizing selective parts of the grains that originated from outside of the sample edges. On the other hand, some dispersed needle-shaped interfaces are observed deeper inside the probed volume in the Others category of HT100-300 for both sympathetic (Fig. 9f) and intragranular (Fig. 10f). These occurrences are much more evenly distributed and dispersed spatially while being relatively far away from the sample edges. Hence, they are deemed valid events and not artefacts.

4. Discussion

The current work presents a systematic study of the effects of AM parameters on the subsequent solid-state phase transformation from δ -ferrite to austenite in terms of microstructure and interface characteristics. The results reveal that tuning AM parameters affects the transformation pathways during the post-AM thermal treatment, altering the texture (Fig. 3), grain structure (Figs. 4 and 5), interface morphologies (Figs. 8–10), and the interphase boundary plane and curvature distributions (Fig. 7). The ferrite-austenite interface characters considerably impact various properties of DSSs including their deformation behavior [34] and restoration processes [33], and are therefore crucial for property optimization. The following will first discuss the observed microstructure characteristics in the as-built condition, followed by analyzing the effects of non-equilibrium microstructures on the driving forces of different phase transformation pathways and the interfaces formed.

4.1. Microstructure characteristics

The crystallographic texture and grain morphology of the as-built ferrite are substantially altered by the change in LPBF parameters (Fig. 1b, d, f, h). AB200-700 exhibits a much stronger ferrite texture and grain growth more aligned to the BD compared to AB100-300. This is mainly caused by the higher laser power in AB200-700. As reported in [22], with a laser power <2000 W, the penetration depth is shallow, and a near-parabolic melt pool geometry with a small curvature is achieved. Such melt pool characteristics are consistent with what has been reported in our previous study on the same DSS 2205 [8]. In this case, as solidification progresses during LPBF, the solidification front between the solid and liquid phases moves in BD, while the grains continue to grow epitaxially [59]. As laser power increases, both thermal gradient and the migration direction of the solidification front become more parallel to the BD [22]. Since the preferred epitaxial growth direction of BCC materials tends to be $\langle 100 \rangle$ [60], a stronger ferrite texture forms when the laser power increases from 100 W in AB100-300 to 200 W in AB200-700.

The formation of Si-O inclusions could have been caused by the rapid cooling rates typically experienced during LPBF, as shown in previous studies on both duplex and single-phase stainless steels manufactured via both conventional and AM routes [8,61–63]. During rapid solidification in laser welding, it has been reported that Si has a tendency to evaporate. Si vapor recondenses on the wall of the keyhole in the form of particles [61]. Si-O inclusions may originate from either the powder feedstock or the laser melting process, as Si has a strong tendency to bond with O [8]. The differences in size and density of Si-O inclusions (Fig. 2) are mainly a result of the different laser power, whereas the scan speed plays a less important role here. During LPBF, a lower laser power is reported to induce higher cooling rates due to less heat input into the melt pool [64]. A higher cooling rate during solidification in steels has been associated with a decrease in inclusion particle size and an increase in particle number densities [23], consistent with the observations in AB100-300.

The crystallographic texture of the parent ferrite and transformed austenite is altered by the change in LPBF parameters. HT200-700 exhibits a strong $\langle 100 \rangle$ and $\langle 110 \rangle$ texture in ferrite and austenite along the BD, respectively, whereas HT100-300 presents a much weaker crystallographic texture in both phases (Fig. 3d, h). While the ferrite texture is largely inherited from the as-built microstructure, the austenite texture is mainly dictated by the parent ferrite texture due to the high tendency towards K-S/N-W ORs between ferrite and austenite as shown in Table 5. LPBF parameters also significantly affect the characteristics of austenite developed during subsequent thermal processing. While the exact same thermal treatment was applied to both AM samples and similar austenite morphologies were obtained in both, a lower austenite phase fraction and more significant instability-induced protrusion growth are seen in the HT100-300 sample, which can be related to the ferrite-austenite interface character. This will be discussed in the next section. The common morphological types of austenite can be attributed to the austenite nucleation at the ferrite-ferrite grain boundary, as sympathetic protrusions on an intergranular austenite grain, and within the ferrite grain (Figs. 4 and 5). While a very low austenite fraction is present in the as-built microstructures regardless of LPBF parameters (Fig. 1a, e), we reported for the same DSS 2205 in our previous study that many of those austenite particles formed at the high-angle ferrite-ferrite grain boundaries during LPBF without pronounced further growth [8,65]. This is not sufficiently resolved in the phase maps in the current study due to the step size used. Upon thermal treatment, these grain boundary austenite particles grow and become allotriomorphs, followed by instability-induced protrusions (Fig. 5a, d) [47,66]. The sympathetic nucleation [45,46] forms additional protrusions that have an LAGB with the allotriomorph (Fig. 5b, e). Within the ferrite grains, preferential austenite nucleation along the ferrite sub-structures [47] and precipitates [29] results in intragranular austenite growth (Fig. 5c, f).

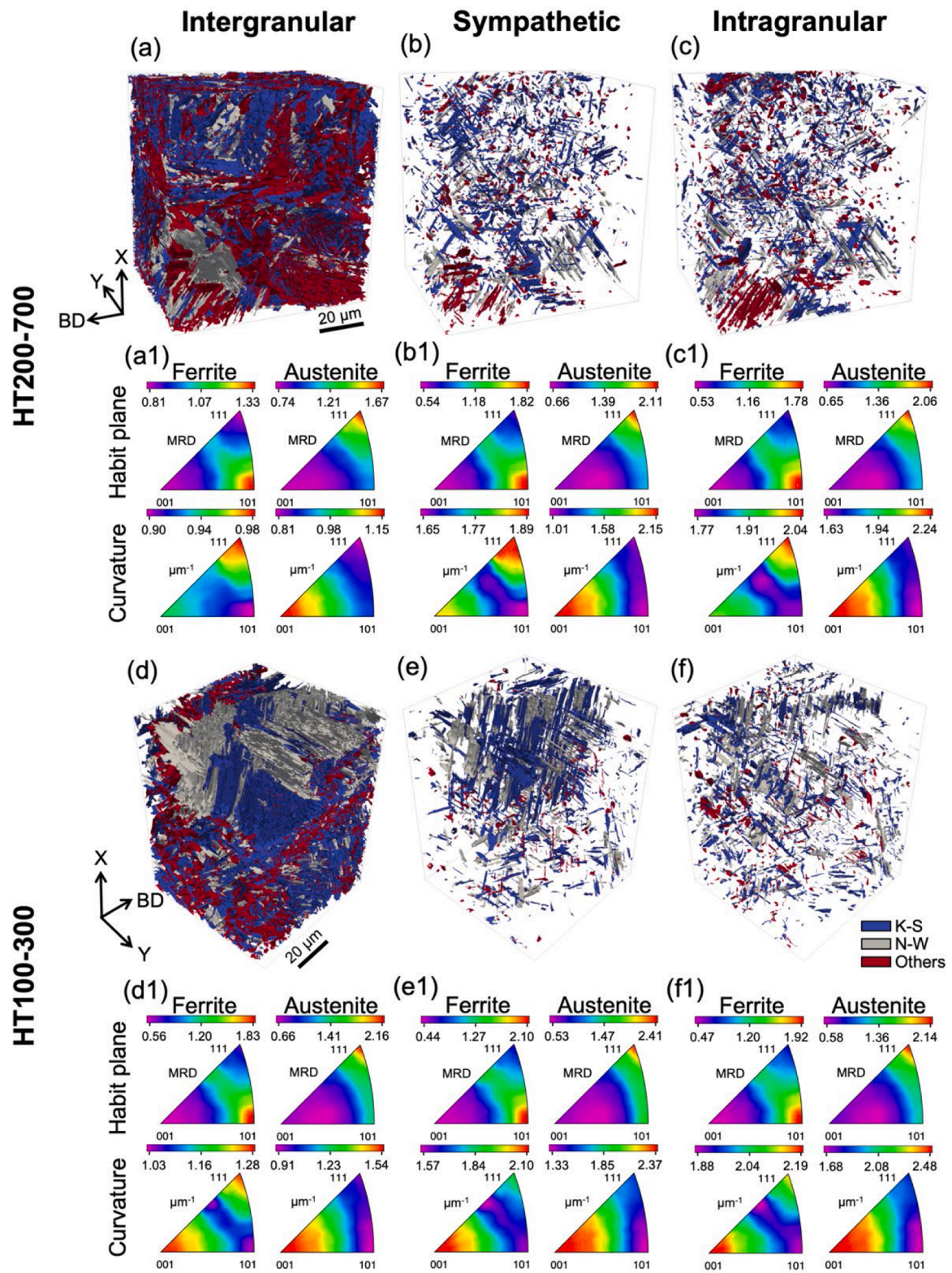


Fig. 7. Interface visualizations and character distributions. Visualizations of austenite-ferrite interfaces for (a, d) intergranular austenite, (b, e) sympathetic austenite, and (c, f) intragranular austenite in (a) – (b) HT200-700 and (d) – (f) HT100-300. Grain boundary plane and curvature distributions in the ferrite and austenite lattice frames for (a1, d1) intergranular austenite, (b1, e1) sympathetic austenite, and (c1, f1) intragranular austenite in (a1) – (b1) HT200-700 and (d1) – (f1) HT100-300. The colour gradient represents multiples of random distribution (MRD) in the plane distributions and curvature in (μm^{-1}) in the curvature distributions.

Table 4

Ferrite-austenite boundary area fractions of different austenite types in HT200-700 and HT100-300.

Austenite type	HT200-700	HT100-300
Intergranular	87±2 %	86±1 %
Sympathetic	7 ± 2 %	6 ± 2 %
Intragranular	6 ± 1 %	8 ± 2 %

Table 5

Ferrite-austenite boundary area fractions of different OR categories for different austenite types in HT200-700 and HT100-300. Measurements were taken from the entire volume of $\sim 100 \times 75 \times 92 \mu\text{m}^3$ and $\sim 100 \times 75 \times 71 \mu\text{m}^3$ for HT200-700 and HT100-300, respectively.

Boundary type	OR	HT200-700	HT100-300
Intergranular	K-S	49±4 %	43±7 %
	N-W	10±4 %	33±8 %
	Others	41±2 %	24±7 %
Sympathetic	K-S	58±6 %	59±8 %
	N-W	24±6 %	33±7 %
	Others	18±7 %	8 ± 3 %
Intragranular	K-S	53±8 %	52±8 %
	N-W	26±8 %	35±4 %
	Others	21±6 %	13±5 %

4.2. Interface characteristics

The phase transformation pathways influence the austenite-ferrite interface characteristics. This can be related to the distinct starting microstructures of HT200-700 and HT100-300, exhibiting different characteristics of intergranular and intragranular defects in the parent ferrite matrix. The crystallographic texture, dislocation density, and precipitate size and density in the parent ferrite affect the type and magnitude of the phase transformation driving force. This leads to changes in interface morphology and crystallography, area fractions of different interface

categories, and interphase plane and curvature distributions.

4.2.1. The roles of interfaces and parent ferrite microstructures on the volume fractions of austenite types

The change in starting microstructures (i.e., LPBF parameters) significantly impacts the austenite volume fraction. For instance, HT100-300 displays a lower volume fraction of austenite (31 %) compared to HT200-700 (39 %). This is caused by the phase interface configuration of the austenite allotriomorph and the change in ferrite texture, which influence the interface characteristics and growth rate of austenite during the phase transformation. When austenite nucleates on the parent ferrite grain boundary and eventually develops into an austenite allotriomorph, three distinct interface configurations around the austenite nuclei may form. Specifically, the nuclei could form rational interfaces with all ferrite neighboring grains, a rational interface with only one ferrite neighbor, or irrational interfaces with all neighboring ferrite grains.

Generally, the high cooling rate (and high undercooling) during LPBF encourages the development of austenite nuclei with at least one rational interface to neighboring ferrite grains. Entirely irrational interfaces with all neighboring ferrite grains are likely developed during isothermal holding at 1000 °C in the post-LPBF thermal treatment since there is less difference in the energy barrier for austenite nucleation with a rational or irrational interface at such a high temperature [12]. Among these configuration options, austenite nuclei with rational K-S/N-W interfaces may be more prone to further growth since they are more energetically favorable and require less energy to grow compared to non-KS/NW ones [67]. In the case of HT200-700, a strong texture in ferrite (Fig. 3d) implies a greater chance for ferrite grains to have close orientations, which could facilitate the formation of rational interfaces with all ferrite neighboring grains for austenite allotriomorphs. Or, when a rational interface is only formed with one of the ferrite neighbors, the other irrational interfaces may have a higher likelihood to align closer with the K-S/N-W range. In contrast, the more random ferrite

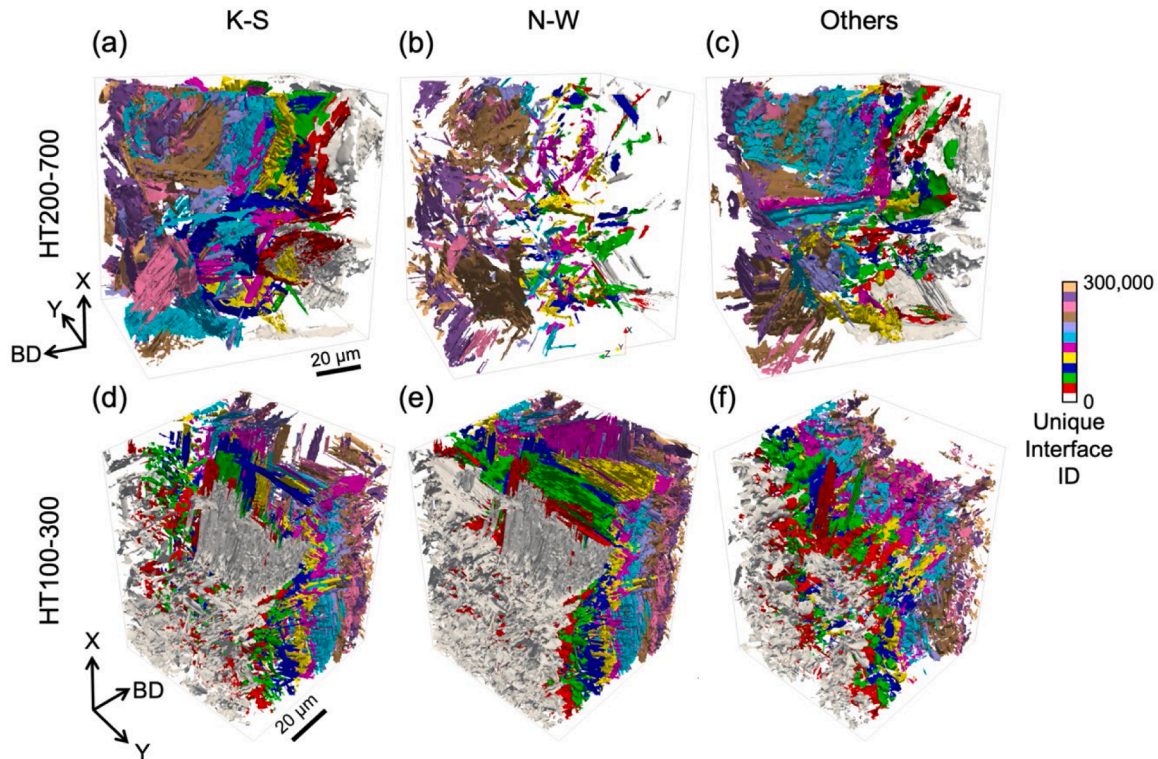


Fig. 8. Interfaces between intergranular austenite and ferrite categorized into various ORs, namely (a, d) K-S, (b, e) N-W, and (c, f) Others, in (a) – (c) HT200-700 and (d) – (f) HT100-300. Unique random colors are assigned to each distinct interface ID.

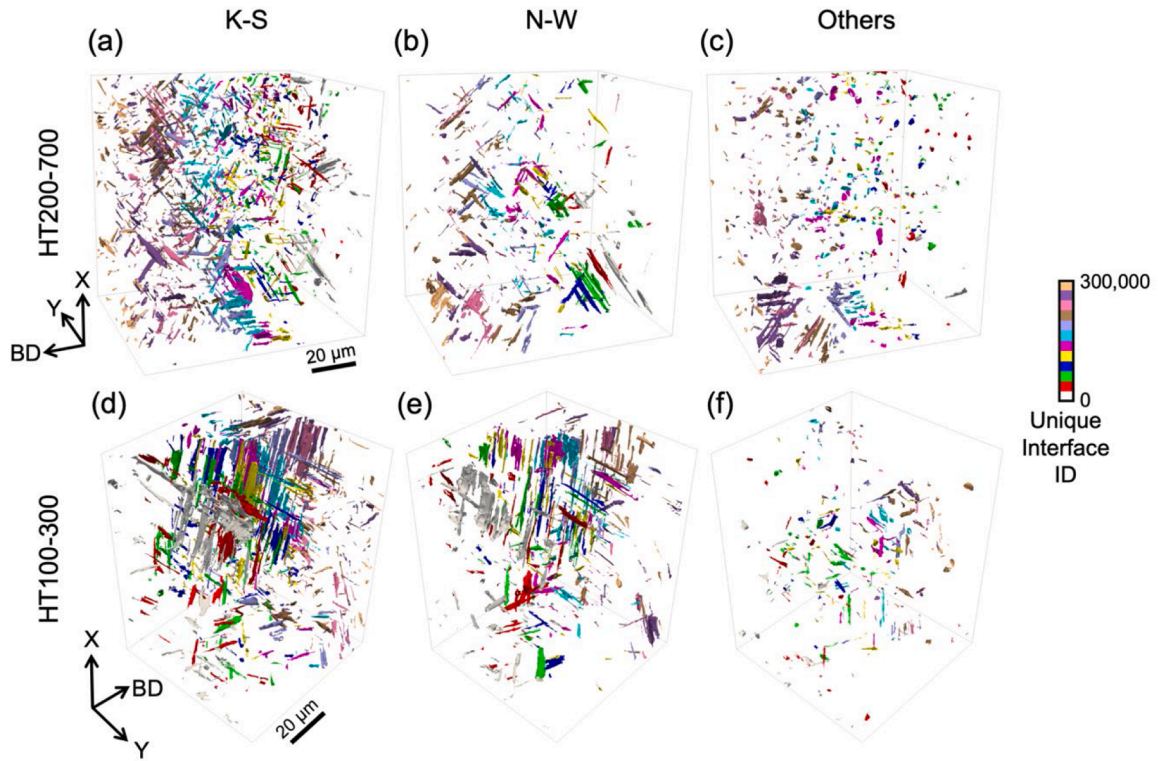


Fig. 9. Interfaces between sympathetic austenite and ferrite categorized into various ORs, namely (a, d) K-S, (b, e) N-W, and (c, f) Others, in (a) – (c) HT200-700 and (d) – (f) HT100-300. Unique colors are assigned to each distinct interface ID.

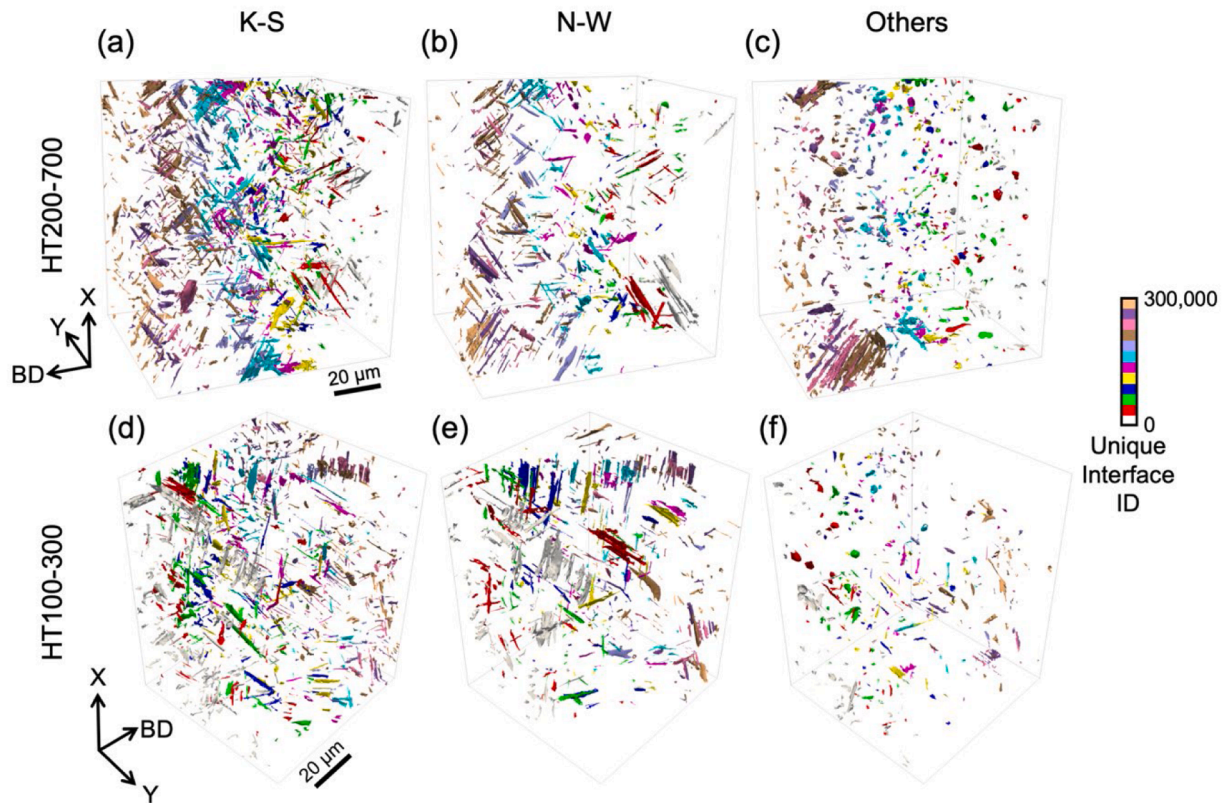


Fig. 10. Interfaces between intragranular austenite and ferrite categorized into various ORs, namely (a, d) K-S, (b, e) N-W, and (c, f) Others, in (a) – (c) HT200-700 and (d) – (f) HT100-300. Unique colors are assigned to each distinct interface ID.

texture in HT100-300 (Fig. 3h) suggests a less likely interface configuration where the austenite allotriomorph maintains rational interfaces with all ferrite neighbors due to the larger orientation spread among ferrite grains. When a rational interface is only formed with one of the ferrite neighbors, the other irrational interfaces may be more likely to deviate significantly from the K-S/N-W range. Consequently, the austenite allotriomorphs in HT200-700 could migrate their interfaces more easily compared to the ones in HT100-300 due to preferential growth along low-energy K-S/N-W interfaces [67]. This ultimately leads to the lower total austenite volume fraction in HT100-300 compared to that in HT200-700, since austenite allotriomorphs tend to be of larger size compared to instability-induced, sympathetic, and intragranular ones. Although no significant change in compositions was observed between the two conditions, Cr, Ni, and Mn vaporization was reported to be promoted by the higher laser energy input during the LPBF process [68].

The austenite growth is dictated by the phase transformation pathways. The growth of austenite allotriomorphs in HT100-300 may be impeded sooner than that in HT200-700, leading to a decreased volume fraction of intergranular austenite in HT100-300 (Table 1). Consequently, the activation energy provided by the post-LPBF thermal treatment may instead be directed to other phase transformation pathways including instability-induced, sympathetic, and intragranular types. This generates more abundant instability-induced protrusions (Fig. 5d) and increased volume fraction of sympathetic and intragranular austenite (Table 1).

Instability-induced austenite protrusions are grouped within the intergranular transformation path due to the absence of any detectable grain boundary in this study. Nevertheless, it is clear that such protrusions become more significant in HT100-300 compared to HT200-700 in terms of both length and density (Figs. 4a, d and 5a, d). Instability-induced growth is initiated by interface perturbations [47, 66]. However, its subsequent development could have been motivated by the tendency of further grain growth of the intergranular austenite. Therefore, the instability-induced growth in HT100-300 mainly occurs on the side of the allotriomorph that follows the K-S/N-W ORs (Fig. 8d–f) since they are energetically preferred, and the growth therefore proceeds more easily.

4.2.2. The effects of dislocations and precipitates on the interface evolution

The lower dislocation density in AB100-300 compared to AB200-700 is associated to the lower aspect ratio (Table 2) of the sympathetic and intragranular austenite in HT100-300 compared to HT200-700. This is because dislocations are the main nucleation sites for sympathetic and intragranular austenite, and assist austenite growth by accelerating elemental diffusion and thus promoting phase transformations [69,70]. The heterogeneous nucleation on dislocation structures favors the formation of rational interfaces [29], in this case, the K-S/N-W ones. The stronger tendency of these sympathetic and intragranular austenite-ferrite interfaces to follow K-S/N-W ORs in HT100-300 compared to HT200-700 may be attributed to the limited nucleation sites provided by its lower dislocation density. Under this limitation, only interfaces with energetically favorable K-S/N-W ORs are more prompt to further growth. This is particularly reflected by the higher K-S/N-W interface fractions of sympathetic and intragranular austenite in HT100-300 (Table 5), which mainly nucleate heterogeneously [12, 47] on existing intergranular austenite-ferrite boundary [71] and ferrite substructures [29,72], respectively, and then grow along ferrite defects such as dislocations.

The relatively smaller and flat K-S/N-W interfaces formed through sympathetic and intragranular mechanisms could be caused by sympathetic nucleation on existing austenite-ferrite interfaces decorated with Cr₂N precipitates and heterogeneous nucleation on Cr₂N precipitates, respectively [29]. These precipitates usually exhibit coherent interfaces with the ferrite matrix [73] and thus contribute to the formation of a rational interface with low interfacial energy [29] between the austenite

and ferrite. These interfaces do not extend as much as the ones that nucleate on dislocation structures. This is because plastic deformation and dislocations usually accelerate elemental diffusion and thus phase transformations, leading to larger and/or more elongated interphase boundary areas [69,70]. On the other hand, the smaller and relatively flat non-KS/NW interfaces formed through sympathetic and intragranular mechanisms could be from the sympathetic nucleation on existing austenite-ferrite interfaces decorated with Si-O inclusions and heterogeneous nucleation on Si-O inclusions, respectively. Such amorphous inclusions usually have incoherent interfaces with the ferrite matrix [8] and, thus, lead to the formation of irrational interfaces between austenite and ferrite [29].

The non-KS/NW interface morphology resulting from the sympathetic and intragranular transformation pathways presents differences between HT200-700 and HT100-300. While only those artefacts that are located near sample edges exhibit needle-shaped morphologies in HT200-700 due to identification of partial grains that may originate from outside the sampled volume, such morphology is also seen inside the probed HT100-300 volume, meaning it is unlikely that these features are artefacts caused by the edge effect. This may be because the denser and smaller Si-O inclusions in AB100-300 contributed to non-KS/NW [29] elongated growth via providing additional opportunities for coalescence of nuclei [74] where adjacent nuclei combine and, hence, form longer grains. The thermal energy spared by the reduced allotriomorph growth may also facilitate such energetically non-favorable growth.

4.2.3. Driving forces of various phase transformation pathways

It should be noted that a decrease in curvature, i.e., a flatter interface, is associated with a decrease in grain boundary energy when assuming a constant mean field chemical potential, as elaborated in [14]. Additionally, in microstructures where grain growth is governed by minimizing interfacial grain boundary energy, an inverse relationship often exists between the distributions of grain boundary energies and grain boundary planes, as noted in [15–17]. Lower-energy boundaries tend to be more common than higher-energy boundaries, minimizing the overall system energy. Consequently, it is expected that there will be an inverse correlation between the distributions of grain boundary planes and curvatures if grain growth is primarily driven by minimizing the interfacial grain boundary energy.

Sympathetic and intragranular transformation paths exhibit similar interface characteristics in HT200-700 and HT100-300, primarily due to their similar driving forces. In addition to interfacial energy minimization, intragranular austenite's heterogeneous nucleation on intersections of ferrite sub-structures (e.g., sub-boundaries and dislocations) [47] or inclusions [12], and its subsequent growth along these features, could be motivated by strain energy minimization, similar to what has been reported before in Ti-alloys [75]. Sympathetic austenite's nucleation on ferrite-austenite interfaces decorated with these features, and its subsequent growth along them, could also be motivated by strain energy minimization. A combination of interfacial energy and strain energy minimization criteria for the sympathetic and intragranular routes is supported by the observation of minimum curvatures not necessarily locating at the planes with the maximum relative areas in the plane distributions. There is only partially inverse correlation between the maxima and minima of the habit plane and curvature distributions in both ferrite and austenite lattice frames (Fig. 7b1, c1, e1, f1). In addition, the minimum curvatures are more concentrated towards planes that do not have the maximum populations in HT100-300. This implies an increasingly strain energy minimization driven growth in HT100-300, likely induced by the stronger tendency to grow on dislocations to accommodate strain compared to those in HT200-700. Nevertheless, it is evident that a change in starting microstructures does not significantly alter the driving forces that govern the phase transformation mechanisms in the sympathetic and intragranular routes.

Despite the similarity in phase interface characteristics of

sympathetic and intragranular austenite, the presence of an LAGB between the sympathetic austenite protrusion and the austenite allotriomorph unavoidably increases the grain boundary energy within the sympathetic transformation path compared to that in the intragranular path, causing a consistently higher area fraction of K-S/N-W interfaces in the sympathetic route (Table 5). This is because K-S/N-W interfaces are known for lower interfacial energies [67], and this may lead to a combined interfacial and strain energy minimization. The strongest alignment with the crystallographically and energetically favorable planes is presented in the sympathetic path, as indicated by the consistently highest maximum and lowest minimum relative areas in the plane distributions for the sympathetic austenite among all austenite types in both samples.

On the contrary, the intergranular transformation path exhibits different interface characteristics in HT200-700 and HT100-300. It shifts from a fully inverse correlation between the curvature and habit plane distributions in both lattice frames in HT200-700 to a partially inverse correlation in HT100-300. This suggests an increasingly strain energy minimization driven phase transformation for the intergranular austenite in HT100-300, whereas the intergranular austenite transformation from ferrite in HT200-700 is mainly interfacial energy minimization driven. This may be caused by the more abundant protrusion growths in the intergranular austenite in HT100-300, where interface perturbations initiated by the instability mechanism may have grown on the defects assisted by strain energy minimization. This could also be attributed to the smaller size of the austenite allotriomorph and the more significant instability-induced protrusions in HT100-300, which effectively make the grain morphology of the intergranular austenite more closely mimic the sympathetic one, in addition to the small allotriomorphs that the sympathetic would need to attach to. This then renders the interfacial characteristics more similar between the intergranular and sympathetic austenite in HT100-300.

5. Conclusions

The effects of AM parameters on the subsequent solid-state phase transformation from δ -ferrite to austenite in an LPBF-manufactured DSS 2205 in terms of microstructure, texture, and interface characteristics were systematically investigated using 3D-EBSD. The as-built ferritic microstructure of the sample printed using 100 W and 300 mm/s contains lower dislocation density, more random texture, smaller precipitates, and higher precipitate density compared to those of the one printed using 200 W and 700 mm/s. The main findings are summarized in the following:

1. The crystallographic texture of the as-built ferrite is substantially altered by AM parameters, mainly due to the change in laser power that affects the cooling rate within the melt pool. Such parent ferrite texture remains after subsequent thermal treatment and dictates the crystallographic texture of the transformed austenite due to the dominance of K-S/N-W OR between ferrite and austenite.
2. The change in AM parameters does not interfere with the morphological types of austenite based on phase transformation pathways, which can be classified as intergranular, sympathetic, and intragranular austenite.
3. A change in parent ferrite texture significantly affects the size of austenite allotriomorphs. A weaker parent ferrite texture suppresses allotriomorph growth via limiting the likelihood of forming K-S/N-W interfaces with neighboring ferrite grains during nucleation at ferrite-ferrite grain boundaries. Due to the typically large volume occupied by allotriomorphic austenite, such a growth suppression effect introduced by a change in texture ultimately leads to a lower total volume fraction of austenite.
4. Higher dislocation density in the as-built ferrite matrix facilitates denser and more elongated austenite growth in the sympathetic and intragranular phase transformation pathways. This is because plastic

deformation and dislocations usually accelerate elemental diffusion and thus phase transformations, leading to more elongated inter-phase boundary areas

5. Under the constraints of limited nucleation sites provided by ferrite dislocations, interfaces with energetically favorable K-S/N-W ORs are more likely to undergo further growth. This leads to higher fractions of K-S/N-W ORs in the sympathetic and intragranular phase transformation pathways.
6. The higher density and smaller size of Si-O inclusions may provide more opportunities for nuclei to coalesce and lead to elongated non-KS/NW interfaces in the sympathetic and intragranular transformation paths.
7. For both samples and among different austenite types, elongated phase interfaces often possess stronger alignment with the crystallographically and energetically favorable planes in ferrite and austenite lattice frames. This lowers the system energy during phase transformation driven by both interfacial energy and strain energy minimizations.
8. A change in starting microstructure shifts the type of driving force that governs the intergranular transformation route from primarily interfacial energy minimization towards a combination of interfacial and strain energy minimization, accompanied by a sharp increase in K-S/N-W interface area fractions.

New insights are provided demonstrating the capability of AM as a microstructural engineering approach for duplex stainless steels to encourage or suppress different phase transformation pathways. This can be achieved by manipulating crystallographic texture, dislocations, and precipitation characteristics through tuning AM parameters. This knowledge can also be applied to other steels or alloy systems that undergo solid-state phase transformations and will ultimately facilitate the design of tailored duplex microstructures with desirable properties.

CRedit authorship contribution statement

X.Y. He: Writing – original draft, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **Z.P. Xu:** Writing – review & editing, Methodology, Formal analysis, Conceptualization. **G. S. Rohrer:** Writing – review & editing, Methodology, Formal analysis, Conceptualization. **C. Kong:** Writing – review & editing, Methodology, Formal analysis. **S. Primig:** Writing – review & editing, Supervision, Methodology, Funding acquisition, Conceptualization. **N. Haghdadi:** Writing – review & editing, Supervision, Methodology, Funding acquisition, Conceptualization.

Declaration of competing interest

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Supplementary materials

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