

Impact of anisotropy and strengthening mechanisms on the fatigue behaviour of laser powder bed fusion processed (FeCoNi)₈₆Al₇Ti₇ high-entropy alloy

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ABSTRACT

The structural integrity of additively manufactured components is frequently limited by anisotropic fatigue properties. However, the influence of precipitation strengthening on this anisotropic behaviour in high-entropy alloys remains insufficiently understood. This study investigates the fatigue crack growth resistance of a laser powder bed fusion processed (FeCoNi)₈₆Al₇Ti₇ alloy. A pronounced anisotropy in the fatigue threshold (ΔK_{th}) is observed, samples where the crack growth direction is at a 45° angle relative to the build direction (BD) exhibit a threshold approximately 40% higher than those at 0°, parallel, and 90°, perpendicular to BD. Examination of the fatigue crack surfaces near the ΔK_{th} region has revealed a dominant step-wise intercellular crack growth with an appearance of cyclic plastic tearing in 45° samples, while transgranular and transcellular fatigue crack growth is dominant in samples with either 0° or 90° to BD. L₂₁ particles have not been found to contribute to crack propagation, but TEM analysis reveals that dense dislocations have piled up in these particles in 45° samples and this is not the case in 0° or 90° samples. It is thus proposed that the macroscopic alignment of the cell boundaries in the 45° sample obstructs the crack advance despite the favourable activation of {111} slip systems. This geometric barrier forces the crack to follow a high-energy, intermittent inter-cellular and overall a step-wise crack growth path. Ultimately, it is concluded that L₂₁ particles act as effective obstacles that induce dislocation pile-ups and force crack deflection, a mechanism activated in the 45° orientation that drives its high-energy fatigue propagation mode and enhanced damage tolerance.

1. Introduction

The demand for high-performance materials in complex geometries across sectors such as aerospace and energy requires alloys that combine high strength with exceptional damage tolerance. To manufacture these components, laser powder bed fusion (LPBF) is an additive manufacturing process used for high-performance alloys such as Ti-6Al-4V [1,2], IN718 [3], Co₂₉Cr₆Mo [4] or AlSi₁₀Mg [5]. Nevertheless, the layer-wise nature of LPBF introduces unique microstructural features. These include process-induced defects like porosity, and lack of fusion (LOF), as well as elongated grains aligned with the build direction (BD) [6–8]. The material resistance to fatigue crack growth (FCG) is

influenced by this inherent anisotropy and the presence of defects. Therefore, the relationship between these microstructural features and mechanical properties must be understood to ensure component reliability [3,9].

In this context, the (FeCoNi)₈₆Al₇Ti₇ high-entropy alloy (HEA), hereafter referred to as the 8677 alloy, is considered a promising candidate for structural applications. It offers an excellent combination of high yield strength, up to 1.1 GPa, and excellent ductility, ϵ_T 0.2–0.4 [9]. These mechanical properties are achieved through precipitation strengthening within a face-centred cubic (FCC) matrix, driven by the formation of coherent, nanoscale L₁₂-ordered multicomponent inter-metallic nanoparticles [10–12]. When processed by LPBF, a supersaturated solid solution is obtained due to the rapid cooling rates associated

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Nomenclature

$\Delta\sigma_a$	Maximum applicable stress range
$\Delta\sigma_0$	Stress range
ΔK	Stress intensity factor range
ΔK_{th}	Stress intensity factor range at threshold
BD	Direction of build
CD	Crack growing direction
CT	Compact tension
EBSD	Electron backscatter detector
EDS	Energy-dispersive X-ray spectroscopy
FCC	Face-centred cubic
FCG	Fatigue crack growth
FCGR	Fatigue crack growth rate
GND	Geometrically necessary dislocation
HRTEM	High-resolution transmission electron microscope

K-T	Kitagawa-Takahashi diagram
KAM	Kernel average misorientation map
LD	Loading direction
LEFM	Linear elastic fracture mechanics
LPBF	Laser powder bed fusion
LOF	Lack of fusion
MPB	Melt pool boundaries
HEA	High-entropy alloy
OICC	Oxide-induced crack closure
R	Load ratio
RICC	Roughness-induced crack closure
S-N	Stress-life
SAED	Selected area electron diffraction
SEM	Scanning electron microscope
TEM	Transmission electron microscope
Y	Geometrical correction factor

with the technique. Consequently, the high-temperature solutioning treatments ($>1200^\circ\text{C}$) required for conventionally manufacturing can be bypassed [13]. However, to activate this strengthening mechanism, a post-LPBF one step aging at 780°C is not an option, but a critical step designed to precipitate these L_{12} phases from the as-built supersaturated solid solution. The mechanical performance of the alloy is enhanced by this single-step thermal treatment when compared to the as-built state [9,10].

The thermal processing designed to form the desirable coherent L_{12} -type also promotes the formation of a secondary, incoherent L_{21} -ordered phase. During this aging treatment, the L_{21} phase precipitates at solidification cell boundaries and grain boundaries, which are solute-rich regions inherited from the rapid solidification during the LPBF process [10–12]. The role of the L_{21} phase in mechanical performance is complex and not understood. While some studies suggest it can increase strength and ductility [10], its incoherent nature and location at boundaries make it a potential site for damage initiation, under cyclic loading, where interface decohesion can lead to micro-crack formation [12]. There is currently no consensus on whether the L_{21} phase is beneficial or detrimental to the fatigue crack propagation resistance.

Furthermore, it is well established that the fatigue performance of LPBF-processed materials is sensitive to the crack orientation with respect to the BD. Previous research on LPBF 8677 alloy has been limited to stress-life (S-N) analysis in a single orientation, loading perpendicular to BD [14]. More importantly, the fatigue crack growth threshold, ΔK_{th} , which determines the condition below which existing defects remain inactive, has not been reported for this alloy [15]. For components subjected to high-cycle fatigue, ΔK_{th} is considered a fundamental design parameter, as it dictates whether the process-induced micro-defects will propagate [16]. In the near-threshold regime, crack propagation is known to be dependent on the local microstructural interactions, including crack deflection by precipitates or shielding induced by grain boundaries [17]. Thus, ΔK_{th} provides essential information regarding the material intrinsic resistance to crack extension [18]. This threshold is a critical parameter for defect tolerance assessments, often visualised in frameworks like the Kitagawa-Takahashi (K-T) diagram, which links fracture mechanics with the fatigue limit of the material [12]. Given the lack of ΔK_{th} data for different build orientations, the effect of microstructural anisotropy on the FCG behaviour and the defect tolerance of this alloy remains to be investigated.

Therefore, this study aims to address these critical knowledge gaps by investigating the anisotropic fatigue crack growth behaviour of a heat-treated, LPBF-fabricated 8677 HEA. The primary objectives are: (1) to determine the fatigue crack growth threshold value, ΔK_{th} , for three distinct orientations of the crack growth relative to the BD (0° , 45° , and 90°); these orientations were selected to characterise the behaviour

along the principal microstructural axes (parallel and perpendicular to grain elongation) and an intermediate condition representing a mixed-mode loading scenario for the layered structure; (2) to correlate the observed anisotropic FCG behaviour with the specific microstructural features, including the cellular structure and grain morphology; and (3) to elucidate the specific role of the L_{21} phase in the governing crack propagation mechanisms in the near-threshold regime.

2. Material and methods

2.1. Specimen fabrication and post-printing heat treatment

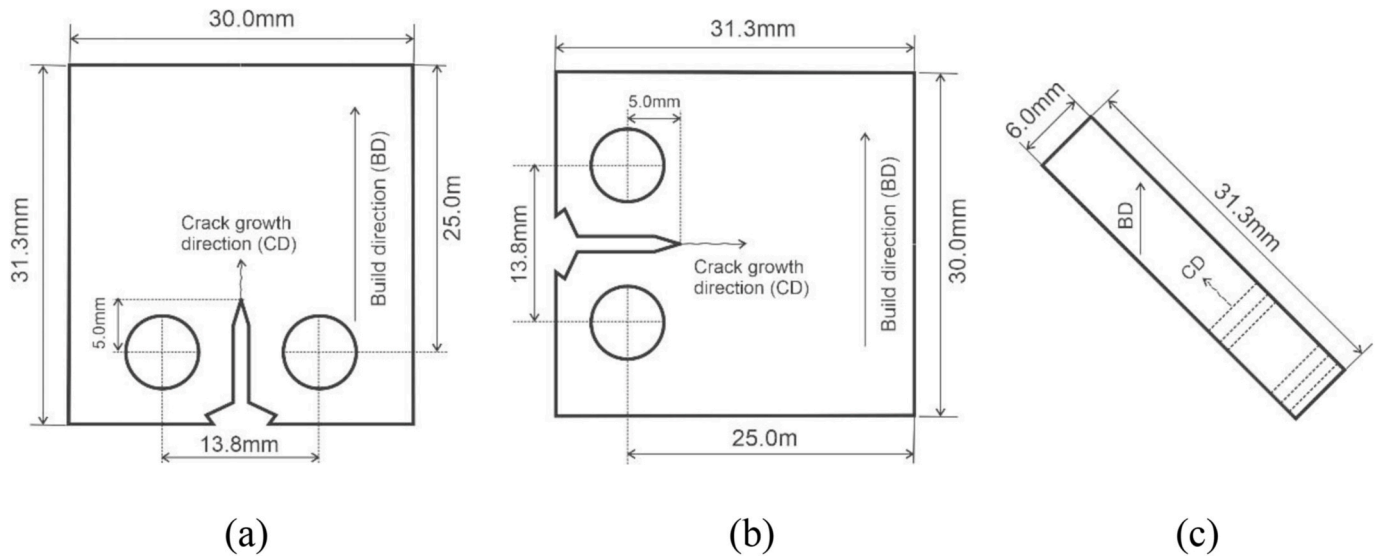
In this research, LPBF was fed with a gas-atomised powder feedstock of 8677 alloy. The actual alloy composition, as analysed by the supplier (Jiangsu Vilory Advanced Materials Technology Co.), is shown in Table 1. The chemical composition of the atomised powder was assessed using an inductively coupled plasma optical emission spectrometer (ICP-OES), along with Leco analysers for Ni, O, H, C and S. The average powder size (D50) was $30.6\ \mu\text{m}$ with D10 and D90 being $18.8\ \mu\text{m}$ and $48.2\ \mu\text{m}$, respectively, as specified by the supplier (quality check following GB/T19077). A commercial machine (HB150) equipped with a 300 W Nd:YAG fibre laser and a wavelength of $1.06\ \mu\text{m}$, and a spot diameter of $50\ \mu\text{m}$ was used. During the sample building, a continuous flow of pure Ar was used as shielding gas. The building parameters used in this work are: laser power at 180 W, scan speed at $1.000\ \text{mm/s}$, layer thickness of $30\ \mu\text{m}$, and hatching space of $100\ \mu\text{m}$. The corresponding volumetric energy density was then $60\ \text{J/mm}^3$, which is an optimal condition for LPBF of the alloy to achieve the maximum relative density with a minimum number of defects due to LOF and keyhole pores [10]. The scan strategy followed the commonly used one of incremental rotation of 67° between successive layers.

To investigate the material anisotropic fatigue behaviour, specimens were fabricated and tested in three distinct orientations, defined by the relationship between the crack direction (CD) and the BD. These orientations, shown in Fig. 1, are referred to as CD//BD (0°), $\text{CD}\perp\text{BD}$ (90°), and $\text{CD/BD} = 45^\circ$ (45°). All specimen dimensions followed the ASTM E647 standard. Each sample was individually built without a notch, which was subsequently introduced via wire cutting after the LPBF process. The samples were then aged at 780°C in a furnace air environment for 2 h, followed by quenching the samples in water. This condition has been demonstrated for the alloy after LPBF, achieving high strength and satisfactory ductility values [10]. The ageing temperature of 780°C (for 2 h) is based on the work of Yang et al. [9] for the alloy after casting, hot working, and homogenisation. The ageing temperature has been found suitable for the alloy after LPBF, achieving general bulk mechanical properties with a yield strength of $\sim 1190\ \text{MPa}$.

Table 1

Composition of the alloy (in atomic %).

Fe	Co	Ni	Al	Ti	Si	P	O	N	Mn	C	S
28.1	29.4	28.4	7.2	6.7	0.11	0.02	0.0262	0.0025	<0.005	<0.005	<0.003

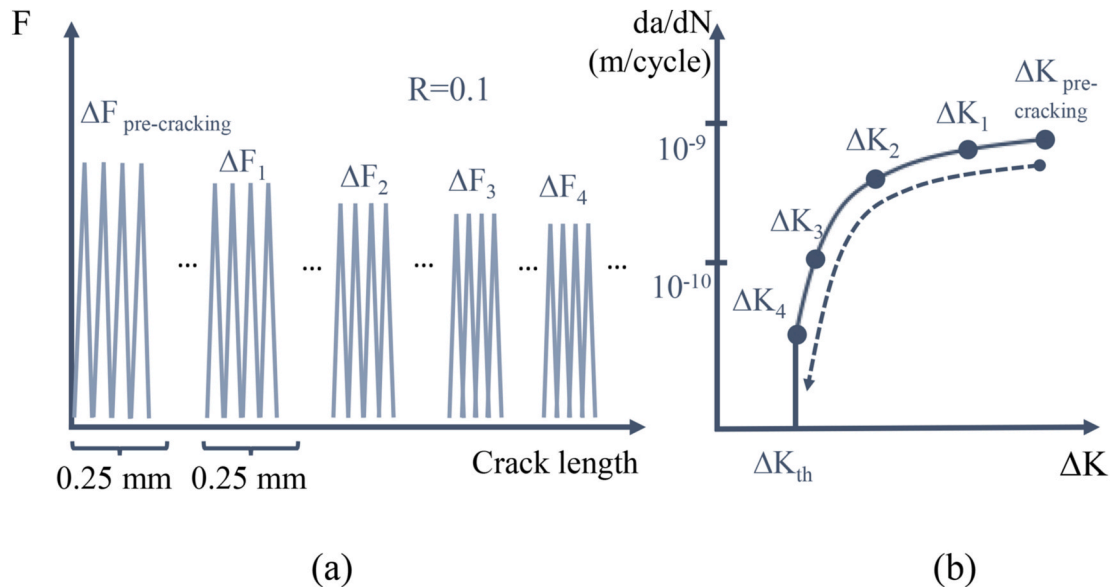
**Fig. 1.** Fatigue crack growth samples with sample dimensions, crack growth direction and build direction indicated, (a) front view of 0°, (b) front view of 90°, and (c) side view of 45° samples.

and an ultimate tensile strength of 1470–1530 MPa [11–13] after ageing for 1–4 h [10–12]. As orientation-specific tensile data are not currently available, these bulk values are used as a representative baseline for the fatigue evaluation. After heat treatment, the sample surfaces were lightly ground off before FCG testing.

2.2. Fatigue testing procedure and load shedding approach

FCG test were conducted to characterise the material behaviour near-threshold and threshold regions. A load-shedding procedure is employed in accordance with ASTM E647 [19]. The test commenced

with a pre-crack at ΔK of 14 MPa $\sqrt{m}$, as shown in Fig. 2a. Following this, the cyclic load range ΔF was decreased in a step-wise manner, initially in 10% decrements. Finer 5% reductions were then applied once the crack growth rate dropped below 10^{-8} m/cycle to ensure more precise determination as the crack approached the threshold region, Fig. 2b. A step size of 0.25 mm was employed for the load-shedding procedure. Although this is a reduction from the 0.5 mm recommended by ASTM E647, it was necessitated by the compact geometry of the LPBF specimens to ensure sufficient uncracked ligament remained for near-threshold characterisation. This approach is technically justified by the alloy's high yield strength (~ 1.1 GPa); the maximum

**Fig. 2.** (a) Force reduction profile. At each level, the crack grows 0.25 mm. ΔK_{th} is reached when da/dN achieves 10^{-10} m/cycle, with a constant R-ratio of 0.1. (b) da/dN vs ΔK strategy, the approach begins in Region II and progressively transitions towards Region I.

monotonic plastic zone size at the start of the test ($\Delta K = 14 \text{ MPa}\sqrt{\text{m}}$) is estimated to be approximately $32 \mu\text{m}$. Thus, a 0.25 mm increment allows the crack to extend nearly 8 times the size of the maximum plastic zone, effectively eliminating load interaction and transient retardation effects throughout the test. The fatigue threshold, ΔK_{th} , was determined by linear regression of the data in the near-threshold region, calculating the ΔK value corresponding to a crack growth rate of 10^{-10} m/cycle , as specified in ASTM E647. A constant load ratio of $R = 0.1$ is maintained throughout this process.

During all tests, load and actuator displacement (stroke) data were recorded every 0.01 s . The fatigue crack initiation and propagation tests were performed using the Walter + Bai LFM-C 5 kN in-situ fatigue test machine. A Gazelle GZL-CL-41C6M camera, with a 2048×2048 pixel resolution, was installed to monitor crack propagation and facilitate digital measurement of the crack length. The optical setup provided a spatial resolution of $1.25 \mu\text{m}/\text{pixel}$. To account for potential crack-front tunneling the optical surface measurements were verified against the actual crack front profiles observed via post-mortem SEM fractography. This optical resolution yields a maximum estimated crack length measurement uncertainty of $\pm 5 \mu\text{m}$. For compact tension (CT) specimens used in this study ($W = 25 \text{ mm}$), this translates to a negligible uncertainty in the a/W ratio of ± 0.0002 . Furthermore, the ΔK was calculated in strict accordance with the ASTM E647 standard for CT specimens, utilising the standard geometric function, equation (1).

$$\Delta K = \frac{\Delta F}{B\sqrt{W}} \cdot \frac{(2 + \alpha)}{(1 - \alpha)^{\frac{3}{2}}} \cdot (0.886 + 4.64\alpha - 13.32\alpha^2 + 14.72\alpha^3 - 5.6\alpha^4) \quad (1)$$

Where $\alpha = a/W$, ΔF is the load range, B is the thickness, and W is the width. Due to the extensive duration of the near-threshold load-shedding tests, combined with the high manufacturing costs and material constraints associated with these specialised LPBF high-entropy alloy specimens, one specimen was evaluated for each of the three orientations, 0° , 45° , and 90° .

2.3. Characterisation techniques

Sections of the material were sliced and polished using vibrational polishers to investigate the microstructure. A JEOL JSM-7000F scanning electron microscope (SEM) equipped with energy-dispersive X-ray spectroscopy (EDS), was employed to analyse the distribution of atomic composition. Additionally, the SEM features an electron backscatter detector (EBSD, Oxford CMOS), which was utilised under optimal parameters to study grain, phase, and orientation distributions. Measurements are taken at the halfway point of the CT sample's thickness.

Residual stress characterisation of the samples was performed using a Bruker D8 Discover diffractometer equipped with a Cr Twist tube and a V filter ($\lambda = 2.2911 \text{ \AA}$). The system included a PolyCap™ for parallel beam generation with a 0.25° divergence and a LynxEye detector with a 2.7° active length in 2θ . Samples were mounted on an Eulerian Cradle with an automatic X-Y-Z stage. Data were collected from 125 to $131^\circ 2\theta$, using a step size of 0.05 and a 1-s acquisition time per step.

Strain values in side inclination mode were recorded for different sample tilt angles (Ψ) 8 steps, 0 – 0.7 range in $\sin^2 \Psi$ ($0, 18.4, 26.6, 33.2, 39.2, 45.0, 50.8$ and 56.8°) at constant azimuth angles (ϕ). Strain vs $\sin^2 \Psi$ was plotted to estimate the stress values. A minimum of six measurements is required on the Strain- $\sin^2 \Psi$ plot, using three values of ϕ ($0^\circ, 45^\circ$, and 90°) that include both negative and positive values for a comprehensive evaluation.

To examine the microstructure at elevated magnifications, a transmission electron microscopy (TEM) specimen was prepared in the form of a lamella, measuring 4 by $5 \mu\text{m}$ with a thickness ranging between 50 and 60 nm . This was achieved using a standard lift-out protocol with an Omniprobe micromanipulator, which is mounted on a focused ion beam Helios NanoLab 650 dual beam system. Room temperature imaging was performed using an image-corrected FEI Titan microscope. TEM images

were acquired utilising a coupled CCD camera (Gatan) for EDS at high resolution. A thorough examination was conducted on multiple lamellas sourced from both tested and non-tested samples. Additionally, high-resolution TEM (HRTEM) images and selected area electron diffraction (SAED) patterns were obtained using a Tecnai T20 from Thermo Fisher.

To evaluate the fracture surface topography and estimate the roughness-induced crack closure (RICC) effects, 2D surface profiles were extracted from the fracture surfaces in the near-threshold region. The topographical data were acquired utilising a Leica DM6 M optical microscope via systematic surface scanning. Quantitative fractographic metrics, including root-mean-square roughness (R_q), maximum profile height (R_z), root-mean-square slope (R_{dq}), linear tortuosity index (R_L), and mean step heights, were calculated to distinguish the crack propagation behaviour.

3. Results and discussion

3.1. ΔK_{th} values for different CD-BD condition samples

The FCG data are presented in Fig. 3 for the 0° , 45° and 90° orientations. The ΔK_{th} values obtained are $4.83 \text{ MPa}\sqrt{\text{m}}$ for the 0° orientation, $6.82 \text{ MPa}\sqrt{\text{m}}$ for the 45° , and $5.17 \text{ MPa}\sqrt{\text{m}}$ for the 90° orientation. While the 0° and 90° orientations show comparable threshold values, the 45° orientation exhibits a distinctly higher resistance, as summarised in Fig. 3. It should be noted that the data was obtained under constant ΔF control at each step. As the crack extends within a single step, the ΔK and crack growth rate increase before the subsequent load reduction. Furthermore, the local non-monotonic fluctuations (e.g., temporary plateaus) observed in the near-threshold regime for the 0° and 90° orientations are attributed to a combination of these discrete load-shedding transients and local microstructural interactions. As the crack propagates through grains with varying crystallographic orientations or encounters distinct cellular boundaries, the local resistance to crack growth fluctuates.

In the higher ΔK regime, corresponding to the initial Paris Law region (Region II), the crack growth rates for all three orientations converge to a nearly identical behaviour. This aligns with the established understanding that the growth of the Paris regime is quite insensitive to microstructural variations, suggesting a convergence of the data that supports experimental reliability. In contrast, the curves diverge as they approach the near-threshold regime (Region I). This divergence is also physically expected, as the fatigue threshold is known to be sensitive to local crack-tip interactions with microstructural

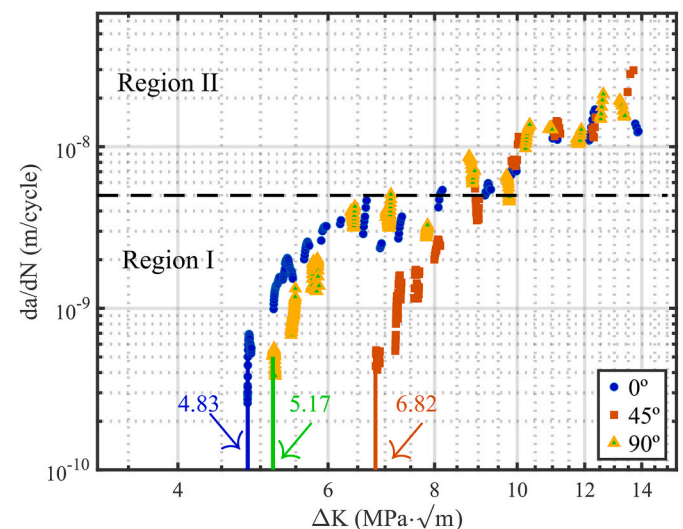


Fig. 3. FCGR data for 0° , 45° and 90° orientations and ΔK_{th} values. Data reflects discrete step-wise load shedding under constant ΔF control.

features. Therefore, the significant differences measured in the ΔK_{th} values are considered a genuine material response to the anisotropic microstructure.

In order to translate the observed anisotropy in fatigue thresholds into a practical measure of engineering defect tolerance [20], K-T diagrams are constructed for each orientation, Fig. 4. The K-T diagram bridges the gap between two failure regimes: (1) a stress-based fatigue limit, $\Delta\sigma_0$, which governs defect-free material, equation 2, and (2) linear elastic fracture mechanics (LEFM) based threshold, ΔK_{th} , for materials containing long cracks, equation (3) [21].

$$\text{Regimen 1. } \Delta\sigma_a = \Delta\sigma_0 \quad (2)$$

$$\text{Regimen 2. } \Delta\sigma_a = \frac{1}{Y} \frac{\Delta K_{th}}{\sqrt{\pi \cdot a}} \quad (3)$$

Where Y is the geometrical correction factor and a is the crack length. The intersection of these two regimes yields a characteristic material length, l_0 , associated with the El Haddad intrinsic crack length [22]. This parameter provides an estimation of the material tolerance to small defects, equation (4).

$$l_0 = \frac{1}{\pi} \left(\frac{\Delta K_{th}}{Y \Delta\sigma_0} \right)^2 \quad (4)$$

For this analysis, the experimentally determined ΔK_{th} values for each orientation were used. It is important to note that the calculation of the l_0 relies on the assumption of a constant defect-free material fatigue limit ($\Delta\sigma_0 \approx 700$ MPa) [12]. We acknowledge that the intrinsic fatigue limit (S-N behaviour) of AM materials is anisotropic [23]. However, acquiring comprehensive high-cycle fatigue data to determine statistically robust, orientation-specific $\Delta\sigma_0$ values for this specific LPBF condition requires extensive testing that falls outside the scope of the present near-threshold fracture mechanics study. Consequently, the absence of such data in the literature necessitated the use of a single, representative bulk value [24]. Therefore, while the K-T diagram remains a valuable comparative tool to assess the relative defect tolerance between orientations, the absolute values of l_0 should be interpreted with this limitation in mind. Furthermore, for anisotropic AM materials, the geometric nature of process-induced irregularities, such as LOF defects, dictates that their effective projected area is direction-dependent [25]. Consequently, a comprehensive evaluation of defect tolerance in operational components must consider both the intrinsic ΔK_{th} anisotropy reported here and the orientation dependence of the effective defect projection size.

Based on this, the analysis of the experimental ΔK_{th} values yields

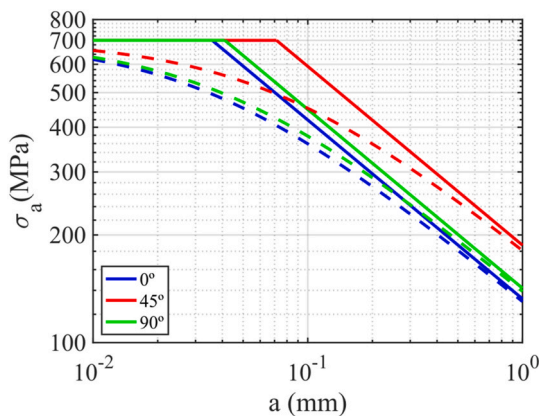


Fig. 4. K-T diagram (continuous line) of the HEA 8677 additively manufactured in 0°, 45° and 90° orientations as a function of the critical defect size. Model parameters derived from ΔK_{th} values and literature [14] are used: $Y = 0.65$ and defect-free material fatigue limit $\Delta\sigma_0 = 700$ MPa. The dotted lines correspond with El Haddad empirical model.

characteristic length of $l_0 = 36 \mu\text{m}$ for 0°, $l_0 = 72 \mu\text{m}$ for 45° and $l_0 = 41 \mu\text{m}$ for 90°. The K-T analysis shows that a 100 μm manufacturing defect could reduce the material safe operating stress from 700 MPa to nearly 300 MPa, with l_0 values ranging from 36 to 72 μm . This theoretical prediction aligns well with the experimental observations from previous work [12], where samples tested at 700 MPa showed reduced fatigue lives and some specimens at 600 MPa failed to achieve run-out, despite micro-CT analysis revealing no detectable defects in the tens of microns range. For non-ideal flaws, such as LOF type, this critical length can be compared to an equivalent crack size estimated using models such as Murakami's approach [26], based on the projected area perpendicular to the loading direction (LD).

This distinction between orientations is best understood by considering the two regimes of the K-T diagram. In the microstructurally small crack regime, the behaviour is similar for all orientations because the analysis assumes a constant fatigue limit $\Delta\sigma_0$. In this regime, failure is governed by crack initiation, where factors like the dimensions and proximity of manufacturing defects play a crucial role in determining the local stress concentration [27]. However, once a defect exceeds l_0 , its subsequent propagation is no longer governed by the features of the initial defect. Instead, growth is controlled by the material intrinsic toughness and its interaction with the bulk microstructure [28]. Therefore, the K-T analysis demonstrates that while all orientations may have similar resistance to crack initiation from very small flaws, the 45° orientation is more resistant to the continued growth of larger cracks, a direct result of its superior microstructure-driven fatigue threshold.

The residual stress state was evaluated to assess its potential influence on the fatigue behaviour. In the as-built condition, tensile residual stresses of +188 MPa were measured, which are characteristic of the steep thermal gradients inherent to the LPBF process. Following the ageing treatment at 780 °C ($T \approx 0.65T_m$), surface measurements on the polished specimens indicated a shift to compressive stresses: −116 MPa for the 0° orientation, −161 MPa for the 90° orientation, and −56 MPa for the 45° orientation (with an error margin of ± 15 MPa). While these compressive values are suggested to be primarily an artefact introduced by the superficial plastic deformation during mechanical polishing rather than a true reflection of the bulk stress state [21], they provide a critical mechanistic insight. Compressive residual stresses act to increase the fatigue threshold by promoting crack closure. However, the 45° orientation exhibits the lowest magnitude of compressive stress (−56 MPa) yet possesses a ΔK_{th} 40% higher than the 0° and 90° orientations. This inverse relationship demonstrates that the superior fatigue resistance of the 45° orientation cannot be attributed to residual stress-induced closure. Instead, it is governed by the intrinsic microstructural barriers, as detailed in the subsequent sections.

The anisotropy observed in the fatigue threshold is of critical importance. Specifically, the increase in ΔK_{th} of approximately 2 MPa $\sqrt{\text{m}}$ for the 45° orientation enhances defect tolerance by expanding the crack arrest region beneath the El Haddad curve. This enhanced resistance to defect-driven fatigue failures represents a central finding of this investigation. Such variation in mechanical behaviour originate from the underlying microstructure developed during the LPBF process and subsequent heat treatment. The following section will therefore detail the specific hierarchical and oriented microstructural features—namely, the FCC/L1₂ cellular/grain structure and the distribution of L2₁ particles along the cell and grain boundaries.

3.2. Mesostructure and microstructure of LPBF 8677 alloy

To elucidate the influence of different structural length scales on the FCG behaviour, the role of the mesostructure, the melt pool boundaries (MPBs), is investigated. Mesostructural features are known to influence the crack growth behaviour of additively manufactured alloys, as has been demonstrated for materials like AlSi10Mg [29]. A comparative analysis was therefore performed on the 8677 alloy. The as-built microstructure in a 90° orientation sample, shown in Fig. 5a, exhibits

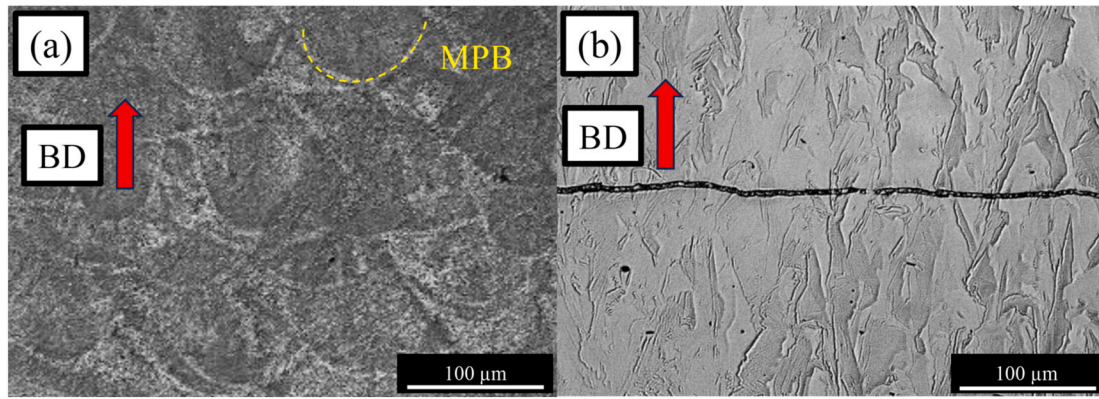


Fig. 5. Effect of heat treatment on the mesostructure and its interaction with a fatigue crack in the 90° orientation. (a) In the as-built condition, distinct MPBs are visible. (b) Following heat treatment, the MPBs are homogenised. A propagating fatigue crack is observed to follow a straight path, demonstrating no interaction with the prior MPB locations and indicating that the mesostructure does not govern the fatigue behavior in the heat-treated state.

a distinct mesostructure characterised by well-defined, semi-elliptical melt pool boundaries. However, following the post-process heat treatment applied to all fatigue samples in this study, these features are homogenised and are no longer discernible, Fig. 5b.

The critical evidence is presented in Fig. 5b, which shows a fatigue crack propagating through the heat-treated material. The crack path is

straight showing no deflection due to MPB locations. If the MPBs were acting as weak paths, a wavy crack path mirroring the semi-elliptical boundary shape would be expected. The absence of such behaviour provides strong evidence that the mesostructure does not govern the fatigue crack propagation in the heat-treated condition. Consequently, the origin of the observed fatigue anisotropy must be sought at the

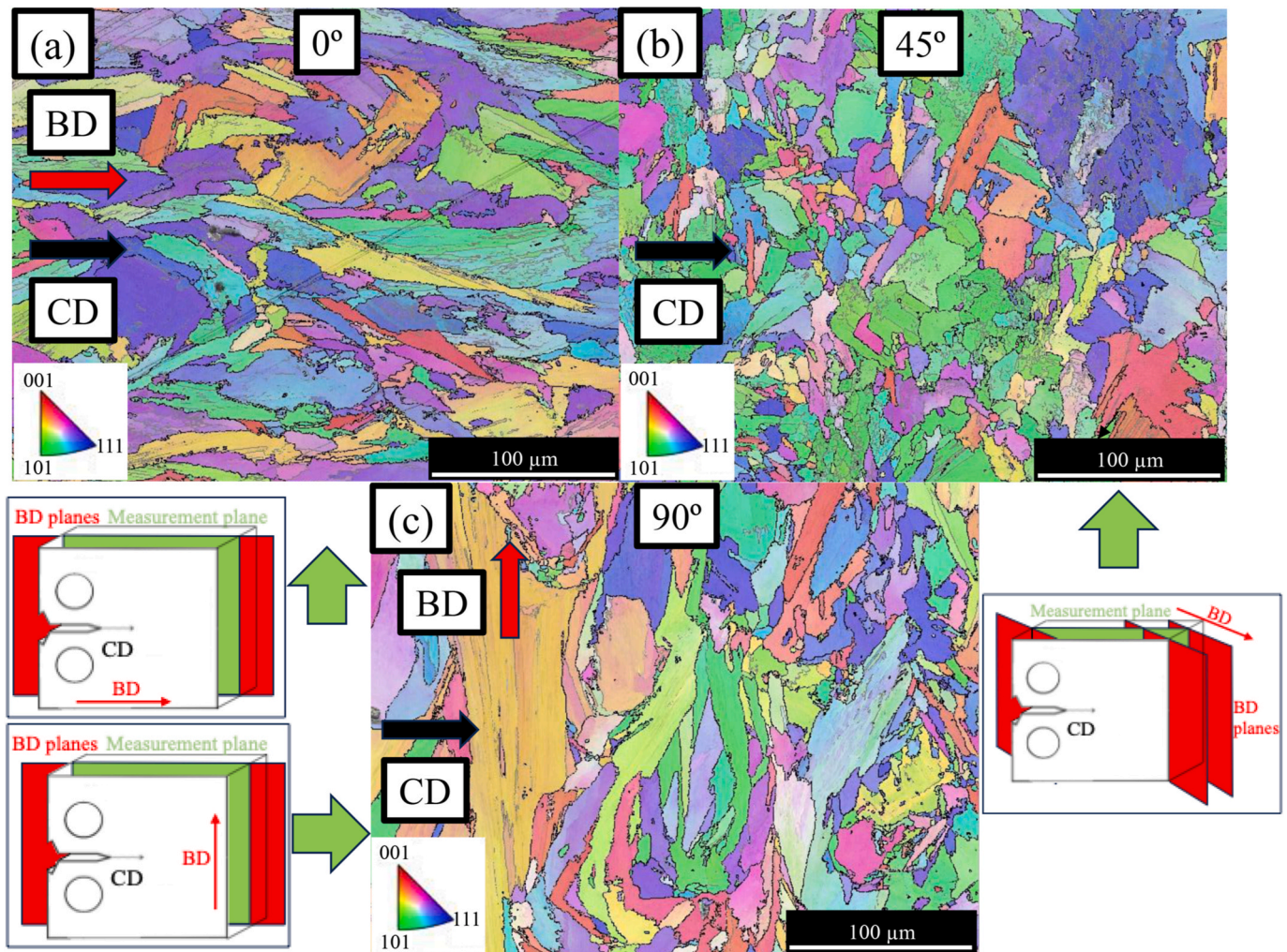


Fig. 6. Microstructure of a LPBF sample. EBSD images of (a) 0° orientation, columnar structure, (b) 45° and (c) 90°. Measurements are taken at the halfway point of the CT sample's thickness.

microstructural level, in the crystallographic texture and the cellular sub-structure, which are the focus of the subsequent analysis.

The EBSD images depicting the 0° , 45° and 90° orientations are shown in Fig. 6a, b and 6c. The 0° and 90° orientation maps illustrate an anisotropic structure defined by elongated horizontal grains, Fig. 6a, and elongated vertical columnar grains, Fig. 6c. This structural characteristic is associated with the manufacturing process, as the grains exhibit preferential alignment along the BD. The 45° orientation, Fig. 6b, does not exhibit a preferred orientation. The grain distribution consists of columnar grains that are sectioned such that oval-shaped is showed, as the measurement plane does not align with the BD planes. The observed columnar grains, at 0° and 90° orientation, measure approximately $5\ \mu\text{m}$ in width and exceed $100\ \mu\text{m}$ in length, extending across the MPs along the BD and displaying an high length-to-width ratio of more than 10. For clarity, the BD planes are defined as those containing the BD within each reference system, as illustrated in the insets of Fig. 6.

A TEM image, accompanied by complementary EDX data, shows

nearly spherical $L1_2$ phases with diameters ranging from 15 to 20 nm, as illustrated in Fig. 7a.

While direct volumetric quantification was not performed in this study, literature on identical LPBF 8677 alloys subjected to the same ageing treatment (780°C for 2 h) reports that these coherent $L1_2$ precipitates constitute the primary intra-cellular strengthening phase with a substantial volume fraction of approximately 40–45% [10,12]. Inside the cells, the analysis identifies two distinct microstructures: one type is enriched in titanium and nickel ($L1_2$), as illustrated in Fig. 7c and e, while the other is enriched in iron and cobalt (matrix), depicted in Fig. 7b and f. The distribution of aluminium exhibits a slight enrichment in the $L1_2$ phase; however, this enrichment is local, as aluminium is situated within the $L2_1$ phase [14]. The SAED patterns confirm the coherent $L1_2$ superlattice, indicating the presence of superlattice diffractions (100) and (-100) , and as the zone axis is different (031) and (0-3-1) weak, can be spotted [30,31], Fig. 7g.

A comprehensive analysis of the sub-grain microstructure of the 8677 alloy samples was performed utilising TEM and SAED. The matrix

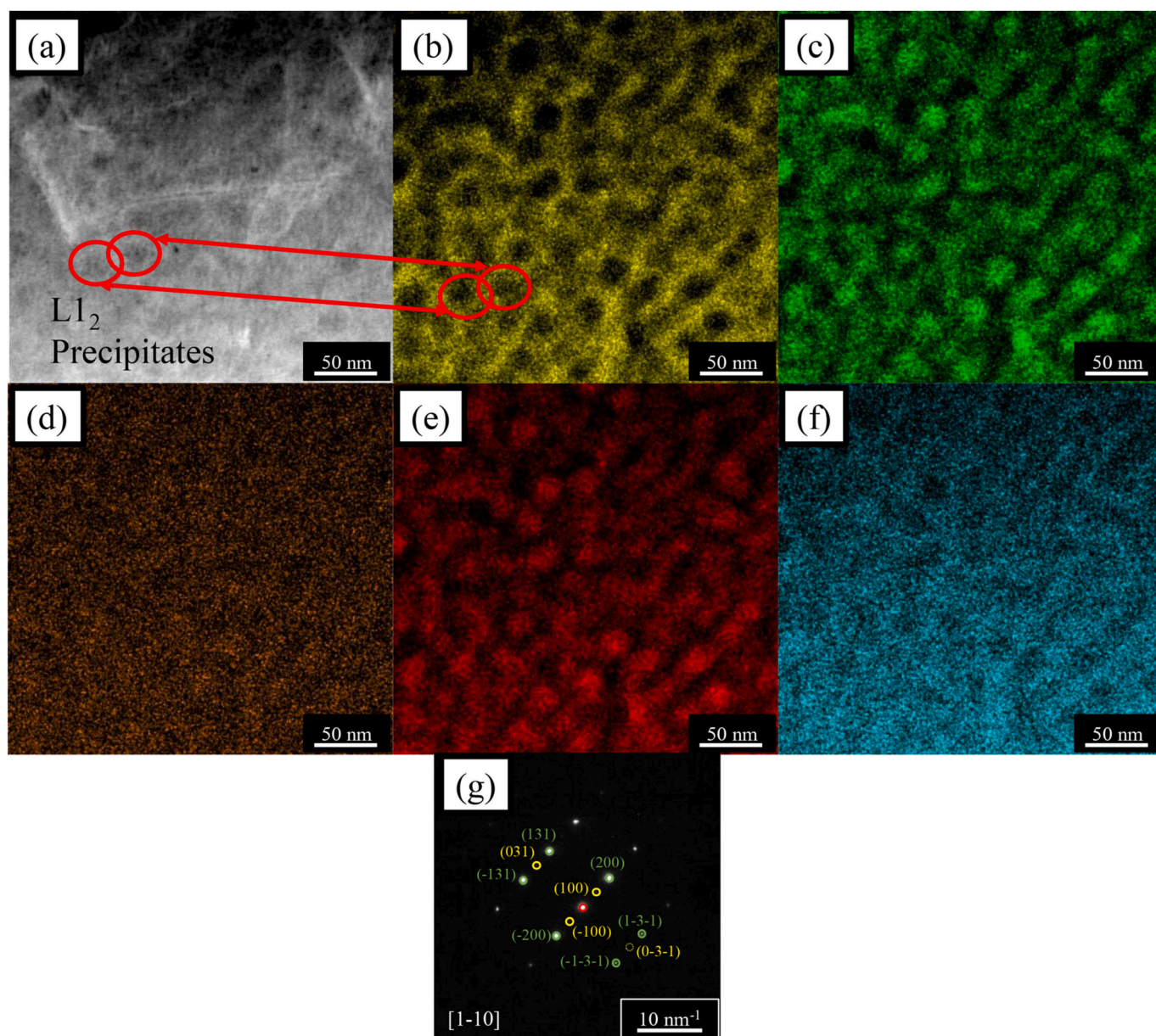


Fig. 7. TEM characterisation of the matrix and $L1_2$ phases. (a) TEM image reveals the matrix structure (FCC), accompanied $L1_2$ phase through black circles. (b), (c), (d), (e) and (f) shows the Fe, Ti, Al, Ni and Co distribution. (g) TEM SAED pattern of $L1_2$.

was confirmed as an FCC structure through HRTEM and SAED patterns, Fig. 8a. Along the cell boundaries, nanoprecipitates (100–200 nm) were observed along the cell boundaries, identified as the L_{21} phase based on SAED patterns, Fig. 8b. Based on characterisations in analogous studies, this L_{21} phase represents a minor volume fraction of approximately 6–13% [10,12]. Crucially for the subsequent discussion on deformation mechanisms, the average edge-to-edge interparticle spacing between these L_{21} boundary precipitates is restricted, measuring approximately 300 nm [10,12]. The cell structures exhibited a columnar morphology with a size of 0.5–1 μm and boundaries approximately 100 nm thick where the L_{21} phase precipitates, Fig. 8c. EDS revealed that these nanoprecipitates were enriched in Al and Ti while being depleted in Fe, Fig. 8c and d, with elemental distribution showing homogeneity within the cell sub-structure and enrichment of Al and Ti at the cell boundaries [10,32]. These L_{21} particles exhibited a random distribution along the boundaries, and their morphology suggests an incoherent interface with the FCC matrix [12]. The precipitation of the particles is understood to occur at these solute-rich locations during the post-LPBF aging at 780 °C [10,33].

TEM analysis, Fig. 9a, confirms the presence of the incoherent L_{21} particles, with dimensions in the range of 100–200 nm, situated along both the grain boundaries and the finer cellular boundaries. A detailed analysis reveals a complex, hierarchical microstructure. At the largest scale, the material consists of primary grains, which are large domains of

similar crystallographic orientation separated by high-angle grain boundaries. Within these larger grains exists a much finer, sub-grain cellular structure. This cellular network is a direct consequence of the redistribution of alloying elements during the rapid solidification inherent to the LPBF process, forming internal walls enriched with segregated elements. As summarised in Fig. 9b, the dominant structure throughout the specimen is a dual-phase FCC/ L_{12} matrix after post-LPBF ageing at 780 °C. It is important to note that in the as-built state, the material consists of a single-phase FCC structure [12]. This matrix is organised into a fine cellular substructure, with an average cell size of 0.6–0.7 μm . These cells shape into larger columnar grains that can extend for many tens of micrometers in length. The coherent L_{12} precipitates within the FCC matrix are nanoscale, with sizes 15–20 nm present all around the matrix.

3.3. Evidence of the anisotropic mechanisms

3.3.1. Crystallographic orientation and crack path analysis

The complex, hierarchical, and anisotropic nature of the microstructure established in the previous sections is correlated here with its effects on fatigue resistance. This section presents experimental evidence obtained in the near-threshold region from crack path analysis, fractography, and TEM to elucidate the origins of the orientation-dependent fatigue behaviour.

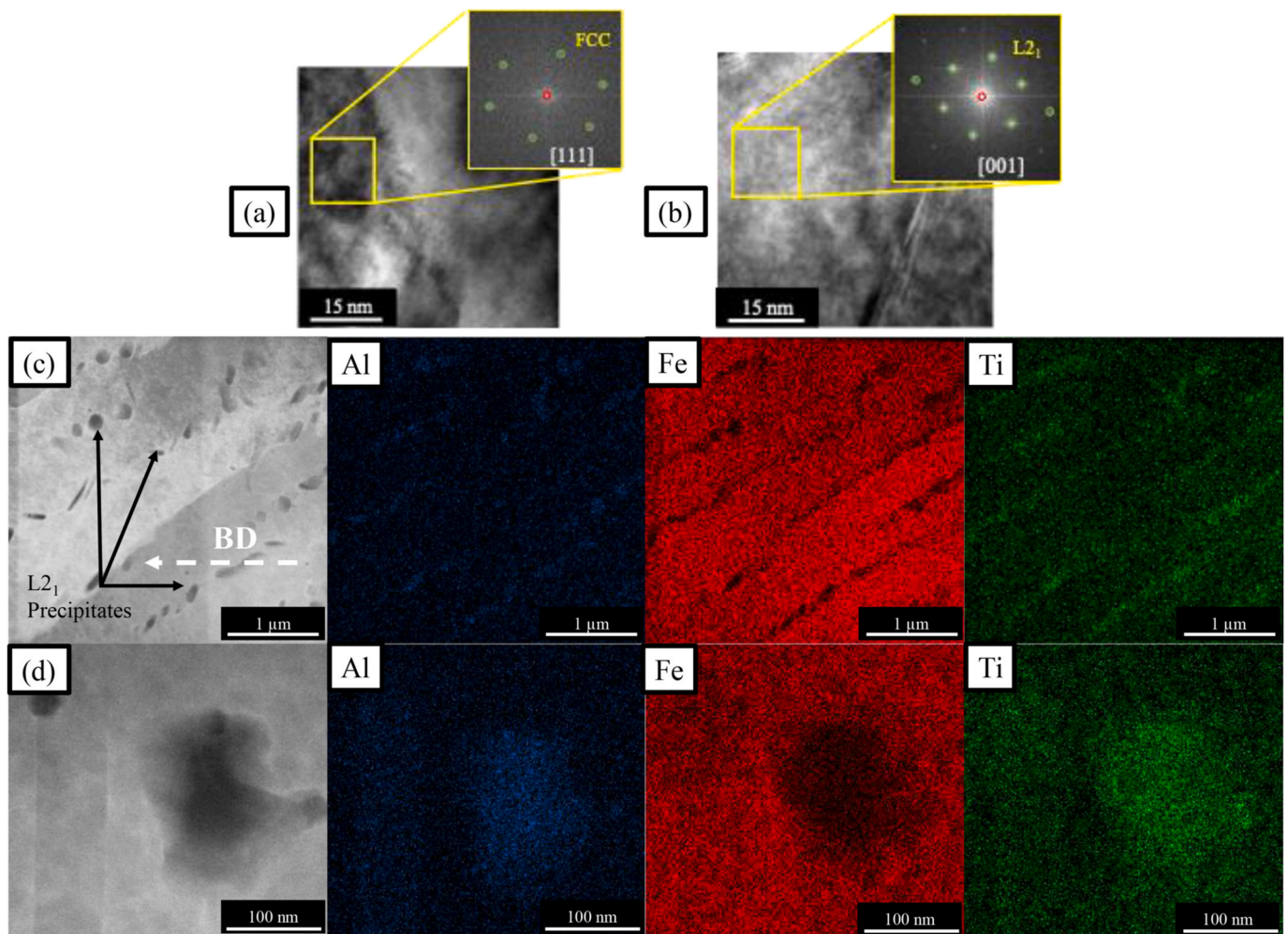


Fig. 8. Dark-field TEM images of the matrix, L_{12} and L_{21} precipitates. (a) The HRTEM image reveals the matrix structure (FCC), accompanied by its corresponding SAED pattern in the inset. (b) HRTEM of the L_{21} precipitates and its corresponding SAED pattern. (c) TEM image showing the columnar structure and the L_{21} precipitations in the cellular boundaries of the composition distribution of Al, Fe and Ti. (d) HRTEM image of an L_{21} precipitation and the distribution of the corresponding elements.

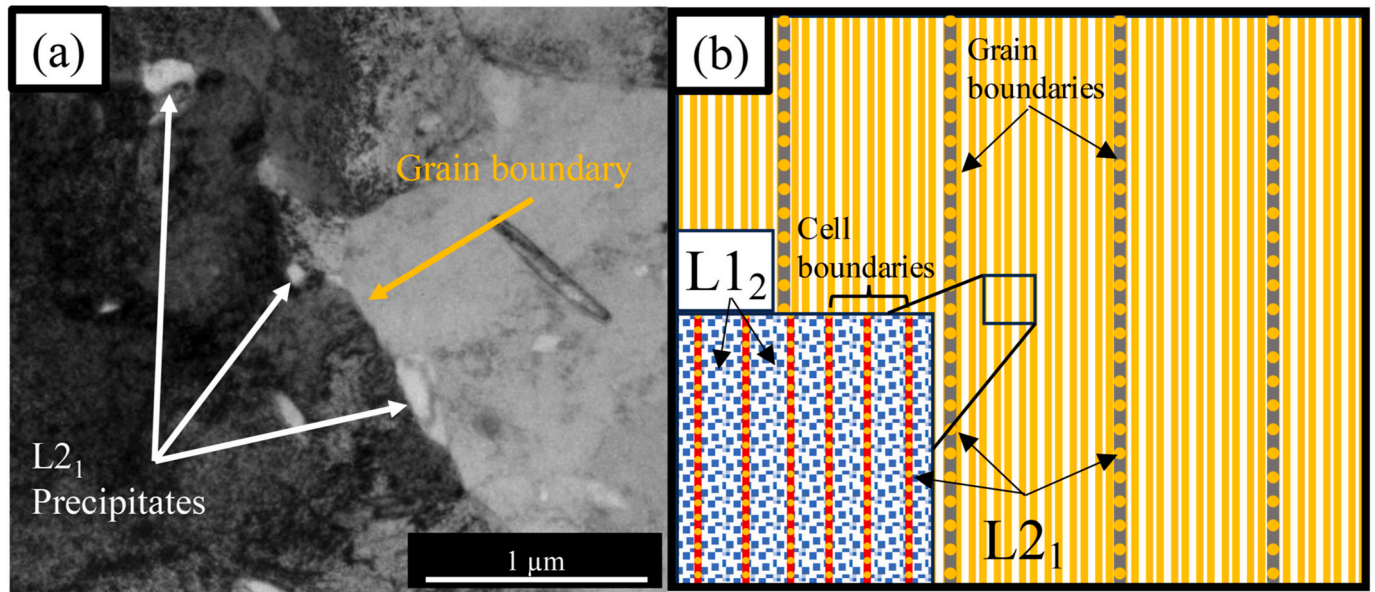


Fig. 9. Multi-scale microstructural features of the HEA alloy. (a) TEM micrograph confirming $L2_1$ particle size (100–200 nm), and (b) a schematic overview of the hierarchical structure, including grains, sub-grain cells, and the distribution of $L1_2$ and $L2_1$ precipitates.

The consequence of the anisotropic crystallographic textures is observed in the resulting crack path. For the 0° and 90° orientations, the crack follows a transcellular path, cutting straight through the cellular substructure as shown in Fig. 10a and g. Furthermore, the Kernel average misorientation (KAM) map for the 0° orientation, Fig. 10c, reveals the subtle misorientations of the pre-existing cellular structure. When correlated with the crack trajectory, it confirms that the crack propagates transcellularly, independent of this cellular morphology. An identical behaviour, where the cellular structure does not affect the crack path, is observed for the 90° orientation, Fig. 10h and i.

The behaviour observed at a 45° orientation differs. In this case, the crack path follows the boundaries of the cells rather than exhibiting transcellular behaviour, propagating instead along an intercellular pathway, as shown in Fig. 10d and e. This provides clear evidence that the crack navigates the cell boundaries, adhering to the path determined by the microstructural morphology.

Although extrinsic crack-shielding mechanisms may affect near-threshold behaviour, their influence on the large-scale anisotropy differences observed between build directions is considered secondary to the intrinsic microstructural response. As all orientations were tested under identical atmospheric conditions, the contribution from Oxide-Induced Crack Closure (OICC) is expected to be consistent. Furthermore, while RICC is present and may contribute to the absolute ΔK_{th} values, the analysis of the crack paths reveals that the 45° orientation possesses a smoother trajectory compared to the more branched 90° orientation, Fig. 10. This observation suggests that RICC alone cannot account for the 40% increase in the fatigue threshold of the 45° specimen. Therefore, while extrinsic factors influence the near-threshold regime, the significant performance gap is attributed to the intrinsic resistance offered by the hierarchical cellular structure.

3.3.2. Fractographic evidence

Fracture surface topography is used to evaluate the energy absorbed during propagation for a given orientation. In previous research, the morphological features of the fracture surface have been correlated with energy dissipation mechanisms. Specifically, it is well established that greater energy is required to propagate a tortuous crack along the nominal fracture plane [29,34].

In the current research, such tortuosity is attributed to the deflection of the crack tip along specific microstructural features, which

induces a crystallographically guided fatigue crack path. This phenomenon is illustrated by the 0° orientation. As depicted in Fig. 11a, a slight misalignment between the solidification cells and the BD plane can be observed. Consequently, a faceted or “sawtooth” morphology is adopted by the crack rather than a linear trajectory. This path represents a compromise: the crack maintains an overall propagation direction perpendicular to the applied load (Mode I) but locally deflects to follow the $\{111\}$ compact planes of the FCC structure that correspond to the FCC matrix and $L1_2$ phase. These slip planes are inclined at an angle of 54° with respect to the cell boundaries. The boundaries are decorated with $L2_1$ phase. The crack path progresses through the same compact planes until mode I is no longer preferred or until it encounters an $L2_1$ particle that deflects and continues through another compact plane, observing a transcellular behaviour.

This two-dimensional path has a direct three-dimensional expression on the fracture surface. The fractography in Fig. 11b reveals that the path is marked by fine, parallel tear ridges or steps. These features correspond to the junctions between the small, angled facets of the sawtooth path observed on the surface.

An analysis of the crack profiles at 45° and 90° orientations indicates a greater degree of branching in the 90° orientation, as illustrated in Fig. 12a and b, which contributes to energy dissipation through increased fracture surface area. Conversely, the 45° orientation demonstrates a comparatively smoother crack path. However, rather than relying on geometric tortuosity, the 45° orientation dissipates energy through extensive plastic deformation. The fracture surface of the 45° sample, shown in Fig. 12c, is the definitive mark of this high-energy, cyclic plastic tearing mode. The fractograph reveals two key features in an intercellular mechanism. Firstly, there are regions where the topographical outlines of entire cellular structures are visible, appearing as if they have been detached during fatigue growth. This is thought to provide evidence of crack propagation along the cell boundaries.

While direct observations of dislocation dynamics were not performed in this study, it is hypothesised that this cyclic plastic tearing process is associated with localised strain concentrations. Given the hierarchical microstructure, these concentrations are likely exacerbated by the presence of the $L2_1$ precipitates decorating the cell boundaries, which can act as preferential sites for void nucleation. The fatigue behaviour of AM alloys is intimately linked to this interaction between the crack tip's plastic zone and the solidification cells, which act as

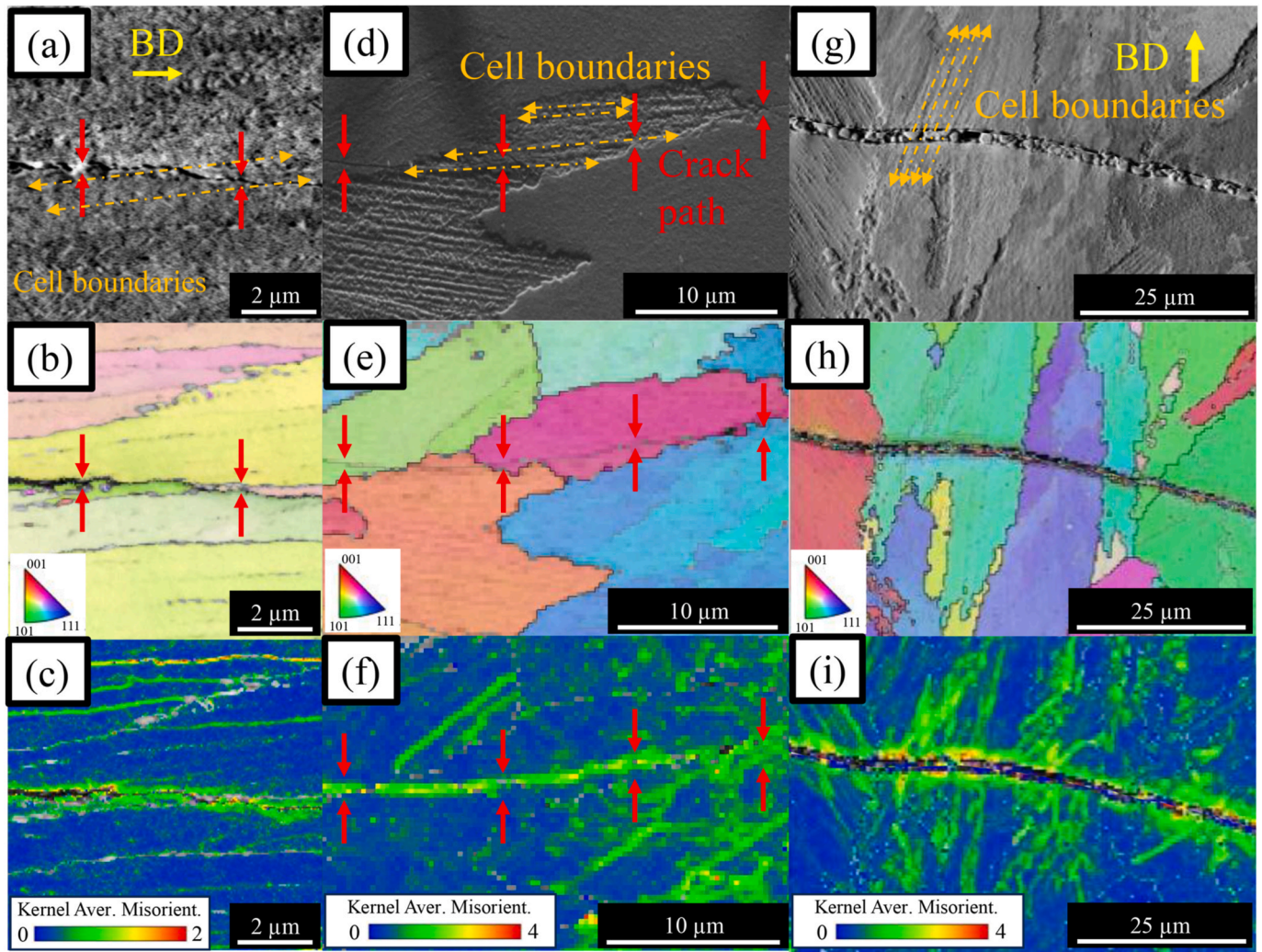


Fig. 10. Microstructure analysis of fatigue crack paths in the near-threshold fatigue region. Surface crack path morphology. (a) SEM, (b) EBSD and (c) KAM map for the 0° orientation, (d) SEM, (e) EBSD and (f) KAM map for the 45° orientation and (g) SEM, (h) EBSD, (i) KAM map for the 90° orientation. Note the different color scales used for the KAM maps. (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

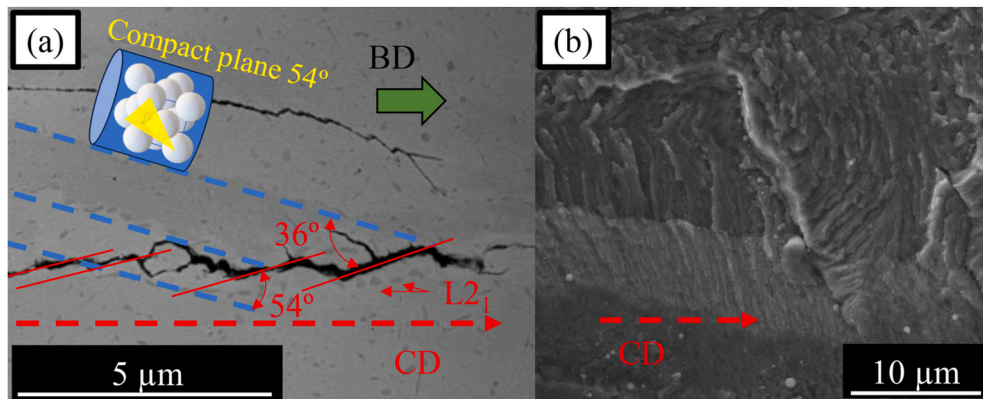


Fig. 11. Evidence of crystallographically guided fracture in the 0° orientation within the near-threshold region. (a) Schematic illustrating the misaligned cellular structure, which causes the crack path to adopt a faceted, sawtooth morphology as it locally follows slip planes. (b) The corresponding fractograph reveals fine tear ridges, which are the topographical manifestation of this faceted path.

primary barriers to crack propagation in the near-threshold regime [35]. Therefore, the combination of intercellular separation and extensive dimple rupture powerfully confirms that overcoming these cellular barriers required significant plastic deformation and energy absorption,

which is consistent with a high-energy, cyclic plastic tearing. This provides the direct microstructural explanation for the superior performance of the 45° orientation, validating its higher measured ΔK_{th} and greater defect tolerance, l_0 .

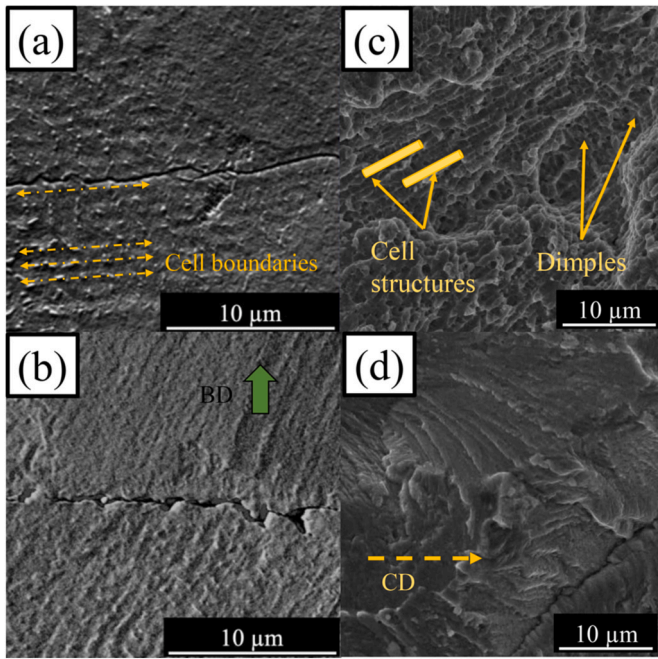


Fig. 12. Crack path morphology on the near-threshold for (a) 45° orientation and (b) 90° orientation. And fractographs of (c) 45° orientation showing cell structures detached and cyclic plastic tearing and (d) 90° orientation with no cyclic plastic tearing.

In contrast, the 90° orientation exhibits a low-energy fatigue propagation mode, consistent with the observations in the 0° orientation. The low-magnification overview in Fig. 12d shows a predominantly faceted surface. The complete absence of the cyclic plastic tearing and cellular detachment seen in the 45° sample confirms that the crack followed a low-energy crystallographic path with lower plastic deformation. A subtle but consistent difference is observed between the 0° and 90° orientations, with the 90° sample exhibiting a higher fatigue threshold. This mechanical difference corresponds directly to the fracture surface topography. Comparative analysis of the fractography, Figs. 11b and 12d, reveals that the 90° orientation displays a greater degree of topographical roughness. A more tortuous crack path requires greater energy dissipation for crack propagation, which accounts for the higher measured fatigue threshold.

This increased roughness is proposed to arise from the crack's interaction with the microstructural features. In the 90° configuration, the crack propagates perpendicular to the elongated axes of the columnar cells. This forces the crack to traverse a greater number of cell and grain boundaries, with each interaction inducing a minor deflection. The accumulation of these deflections results in a more tortuous overall path and the higher energy expenditure required for crack growth, leading by a bigger plastic wake, Fig. 10c and i.

The surface topography parameters analysed of the crack profiles near the threshold area across the thickness evaluate the morphological differences between the build orientations, as summarised in Table 2.

The analysis shows a distinct local facet orientations and microstructural interactions. The 0° specimens exhibited flatter crack surface, characterised by a R_{dq} of 0.24 and a R_L of 1.02. Meanwhile, the 45° specimens shows a tortuous crack path, with the R_{dq} parameter

increasing to 0.54 and mean step heights growing from 18.7 μm to 25.9 μm .

These local deflections and increase of the asperities in the 45° orientation (R_q and R_z) suggest an enhanced RICC effect. This quantitative evidence supports the fact that the elevated fatigue threshold observed in the 45° condition is not only governed by intrinsic microstructural variations, but is augmented by geometric interlocking and closure mechanisms. This intercellular separation, combined with localised cyclic plasticity, supports that overcoming these cellular barriers required significant plastic deformation and energy absorption.

3.4. The role of L2₁ particles on the near-threshold FCG

The incoherent L2₁ phase is reported in the literature as a microstructural feature influencing the mechanical properties of the 8677 alloy [36,37]. Studies have associated an increased volume fraction of L2₁ precipitates with a lower ductility, attributing this to a weak L2₁/matrix interface [10,11].

In a low-stress near-threshold fatigue regime, a more nuanced role for the L2₁ phase is revealed. Notably, a deflection of the crack was observed upon encountering the L2₁ particles at a 0° orientation, as illustrated in Fig. 11a. However, no additional interactions have been recorded. To investigate the particle-dislocation interactions in detail, a TEM lamella was extracted from a single grain adjacent to the crack tip, containing the crack-tip plastic zone, Fig. 13a. HRTEM analysis of this region confirms that under these low-stress conditions, the L2₁ particles act as potent obstacles to dislocation motion, Fig. 13b–e. However, in contrast to observations from higher-stress tests, no evidence of interfacial decohesion nor particle cracking is found. This suggests that near the fatigue threshold, the crack-tip driving force is insufficient to cause interfacial failure. The L2₁ particles, therefore, do not act as damage initiation sites nor weak interfaces that would assist nor accelerate crack propagation, as often observed under monotonic high-stress conditions. Instead, in this near-threshold regime, they resist crack advance by functioning as potent barriers to dislocation motion. The resulting dislocation pile-ups at the L2₁ interfaces accommodate significant plastic strain without decohesion, dissipating energy and contributing to the toughening of the material.

The intensity of this dislocation-obstacle interaction is anisotropic, providing direct microstructural evidence for the observed differences in fatigue resistance. A stark contrast in dislocation activity between the orientations is observed.

- In the 45° orientation, where the crack is forced to follow an intercellular path, the interaction with L2₁-decorated boundaries is maximised. This leads to the formation of dense dislocation tangles and pile-ups around the L2₁ particles located in the cell boundaries, as evidenced in Fig. 13b and c. This intense dislocation activity is a primary mechanism of energy dissipation, contributing to the localised plastic deformation that manifests as a cyclic plastic tearing.
- In the 0°/90° orientation, the crack bypasses the cell boundaries. This results in minimal interaction with the L2₁ obstacles, leading to a lower density of accumulated dislocations at these interfaces, as seen in Fig. 13d and e.

Therefore, the results demonstrate that in the near-threshold regime, L2₁ phase functions as a source of toughening through obstacle-induced plasticity and crack deflection. This mechanism is fully activated in the 45° orientation, contributing to its high toughness, but is avoided in the 0° and 90° orientations, which is consistent with their low-energy fatigue crack growth behaviour and lower fatigue thresholds.

Geometrically Necessary Dislocation (GND) density maps were extracted from high-resolution EBSD data around the crack propagation paths, Fig. 14. In the 0° orientation, Fig. 14a, where the crack propagates parallel to the elongated cell boundaries, the maximum GND density

Table 2
Fractographic metrics for the near-threshold fracture surfaces.

Orientation	R_q (μm)	R_z (μm)	R_{dq}	R_L	Mean Step Height (μm)
0°	9.75	42.90	0.24	1.02	18.7
45°	13.42	68.51	0.54	1.12	25.9

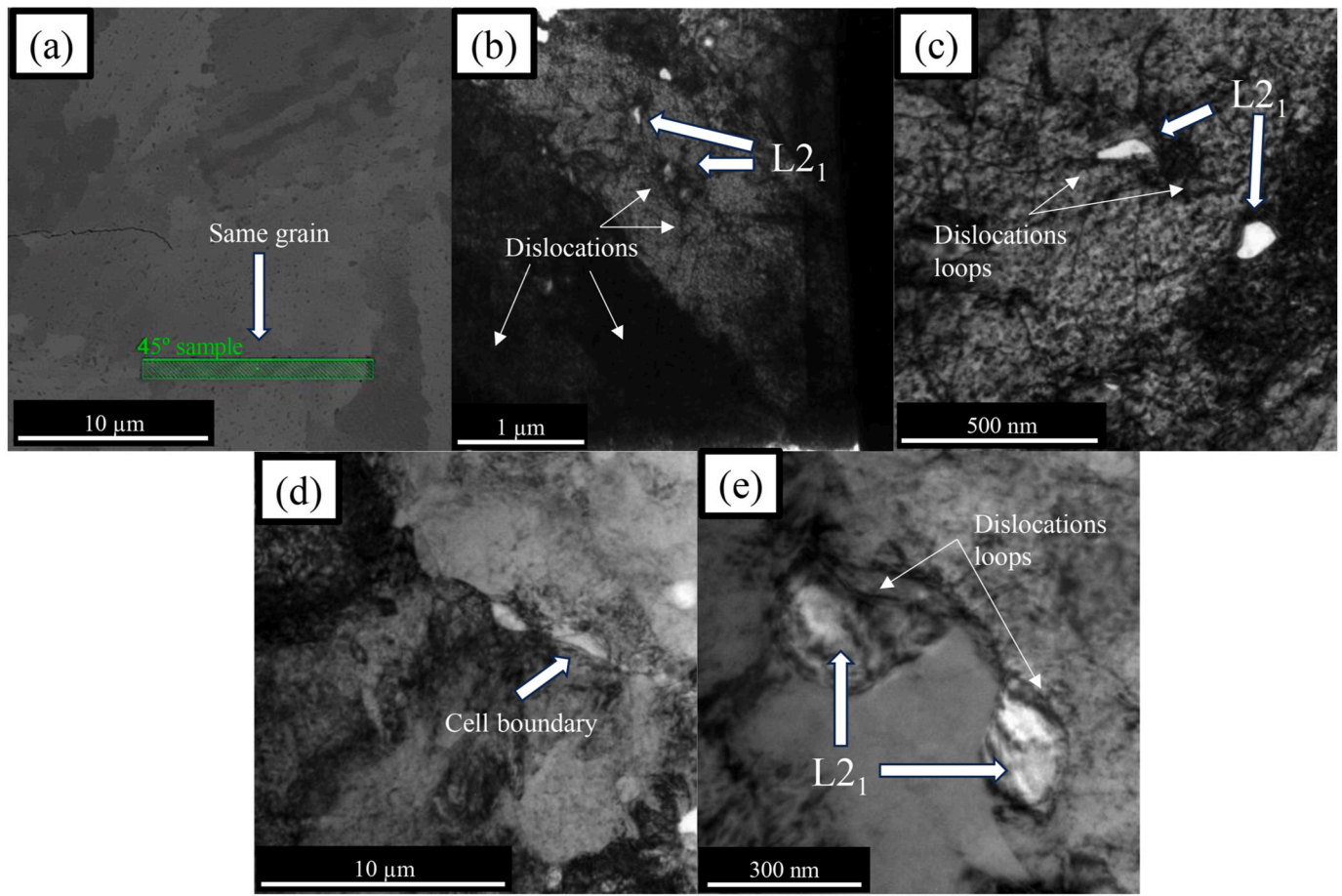


Fig. 13. Dark-field TEM images of dislocation-particle interactions in the near-threshold regime. For the 45° orientation: (a) location of the extracted lamella near the crack tip, (b) high dislocation density within the cellular structure, and (c) magnified view showing dense dislocation loops accumulating around $L2_1$ particles. For the 90° orientation: (d) corresponding overview of dislocation structures and (e) magnified view revealing a lower density of dislocations interacting with the $L2_1$ particles.

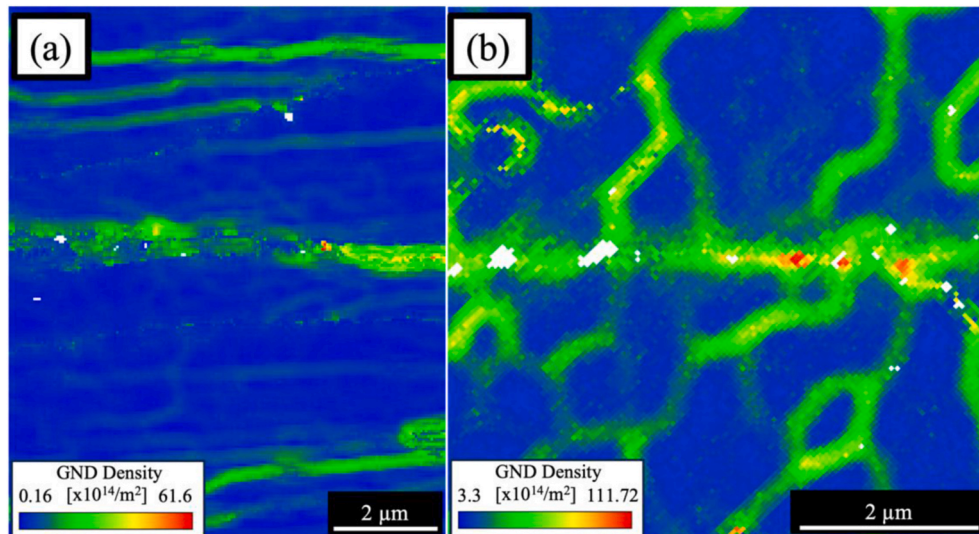


Fig. 14. GND density maps of the crack propagation path in the near-threshold regime for (a) the 0° orientation, showing localised deformation parallel to the cellular structure, and (b) the 45° orientation, demonstrating dislocation pile-up and intense energy dissipation at the cell boundaries.

remains linked to these cell boundaries. The plastic deformation is localised along the transcellular crack wake, with minimal strain accumulation at the surrounding boundaries. Conversely, the 45°

orientation, Fig. 14b, exhibits an increase in GND density. In this condition, the crack is forced to navigate the circular cross-sections of the cellular structure. The GND map illustrates how dislocations, generated

within the cells, pile up against the L2₁-decorated boundaries.

3.5. Postulated governing mechanisms of near-threshold anisotropy

The following mechanisms are proposed as a mechanistic interpretation of the preceding fractographic and microstructural data. The anisotropic fatigue behaviour observed in this alloy is suggested to be governed by a competition between different crack propagation paths.

To evaluate slip activation across these paths, the maximum Schmid factor (m) for the primary $\{111\}\langle 110 \rangle$ slip systems was extracted from the EBSD data. The analysis demonstrates high and uniform values across all conditions relative to the loading axis: $m \approx 0.46$ for the 0° orientation, and $m \approx 0.45$ for both the 45° and 90° specimens. This uniformity indicates that transcellular slip activation is favourable regardless of the build orientation. The governing mechanism in the near-threshold region is determined by the geometric alignment of the macroscopic maximum shear stress with the microstructural paths (cellular and grain boundaries). Under Mode I loading, the maximum shear stress occurs on a plane oriented at 45° to the tensile axis, creating two possible scenarios.

3.5.1. Crack propagation mechanism in 0° and 90°

In the 0° and 90° orientations, the governing mechanism is a low-energy transcellular fatigue crack growth. The primary $\{111\}$ compact planes of the FCC matrix are aligned close to the 45° plane of maximum shear stress, 45° to the LD, as depicted in Fig. 15a-d. This crystallographic condition provides an energetically convenient path, allowing the applied Mode I stress to be resolved into a high shear component that activates dislocation slip. Instead of requiring high-energy crack blunting, this favourable alignment promotes localised planar slip at the crack tip. The resulting localised plastic deformation accommodates the forward growth of the crack, leading to a plasticity-controlled transcellular path. This flatter crack advance is energetically more favourable than navigating the tortuous cell boundaries, Fig. 15g and h. This proposed mechanism is powerfully supported by direct experimental evidence. First, the crack path analysis in Fig. 10a-c and 10h-i confirms that the crack propagates in a transcellular path, ignoring the cellular substructure. Second, this low-energy crack path results in limited plastic deformation, which is validated by the low density of dislocation pile-ups observed at cell boundaries in the TEM analysis, Fig. 13d-e. Finally, the resulting fatigue failure is dominated by low-energy mechanisms, as evidenced by the crystallographic facets that dominate the fracture surfaces in Figs. 11, 12b and 12d.

3.5.2. Crack propagation mechanism in 45°

In contrast, the 45° orientation derives its superior fatigue resistance from an intercellular crack propagation mechanism. Since the EBSD data demonstrates that the internal slip planes are favourably oriented for slip ($m \approx 0.45$), dislocations are generated. However, the grain and cell orientations established by the LPBF process align the L2₁-decorated cell boundaries as physical barriers against the advancing crack path, Fig. 16a and b. The crack follows the path of the next lowest resistance route, which, in this orientation, is along the cell boundaries, as they are aligned with the plane of maximum shear stress, $\tau \approx \tau_{max}$, Fig. 16c. The crack path analysis in Fig. 10d-f, shows the crack following a tortuous, intercellular path along the cell boundaries. This forced interaction with the L2₁-decorated boundaries necessitates extensive plastic deformation, as confirmed by the intense dislocation pile-ups observed via TEM, Fig. 13b-c. This high-energy, deformation-rich process is validated by the cyclic plastic tearing morphology, characterised by cell detachment and extensive dimple rupture on the crack surface, Fig. 12a and c.

3.6. A unified model for anisotropic crack propagation in near-threshold fatigue

A fundamental distinction of the near-threshold fatigue findings is

the complete absence of decohesion or fracture of the L2₁ phase. This stands in contrast to the general characterisation in the literature where the incoherent L2₁ phase is associated with reduced ductility due to its weak interface with the FCC matrix [10,12]. However, this apparent discrepancy is resolved when considering that the failure mechanism of the L2₁ phase is dependent on the applied stress regime.

High-Strain and Tensile Regimes in Literature: In studies involving monotonic tensile loading or high plastic strain, the L2₁/matrix interface is a well-documented site for damage initiation [10,14]. The mechanical incompatibility between the hard L2₁ particle and the softer matrix leads to severe stress concentrations from dislocation pile-ups, causing interfacial decohesion and the nucleation of microvoids. This behaviour is a primary contributor to ductility loss in aged samples under tension [10–12,14].

Near-Threshold Fatigue Regime: The results demonstrate that under the low driving force of the near-threshold condition ($\Delta K_{th} \approx 4.8\text{--}6.8 \text{ MPa}\sqrt{\text{m}}$), the mechanism changes, Fig. 3. The stress intensity at the crack tip is insufficient to accumulate enough dislocation around the L2₁ phase to increase the stress intensity factor locally and make the particle crack or make it able to be failure initiation site. Instead, the L2₁ phase functions as a potent obstacle to dislocation motion. This is confirmed by the TEM analysis, Fig. 13, which shows dense dislocation pile-ups at the L2₁ interface without any evidence of decohesion. In this regime, the particles contribute to toughness by forcing crack deflection and dissipating energy via localised, obstacle-induced plasticity.

Another critical distinction is the description of anisotropic fracture mechanisms. For instance, Liu et al. [11] reported a mixed-mode fracture in their 0° oriented tensile samples was reported, with cracks propagating both transgranularly and intergranularly. However, for their 90° samples, they reported that failure occurred primarily via decohesion [11]. Other studies have proposed a unique “intercellular fracture” mode, where cracks follow the boundaries of dislocation cells [10].

Transitioning from direct experimental observations to a mechanistic interpretation, the current findings provide the empirical basis for a unified model that encompasses these seemingly disparate observations. In this postulated model, the crack follows the path of least resistance, governed by the local resolved shear stress. When a low-energy crystallographic path is available, as in the 0° and 90° orientations, the favourable alignment of $\{111\}$ slip planes promotes planar slip localisation. This localisation lowers the energy required for cyclic crack advance, creating a transgranular path despite the alloy's intrinsic macroscopic ductility, resulting in a lower fatigue threshold.

When the cell boundaries are aligned with the maximum shear stress plane, as in the 45° orientation, this low-energy transcellular advance is geometrically obstructed by the cellular walls, and the crack is forced to follow the microstructural path along the boundaries, Fig. 16. In this regime, the L2₁ precipitates located at the cell walls act as effective barriers to dislocation motion. The high dislocation density observed in Fig. 13 is process-induced; it arises from the intense cyclic stress field ahead of the crack tip where dislocations pile up against these precipitates. This dynamic interaction promotes significant plastic dissipation, resulting in a cyclic plastic tearing intercellular mode and superior fatigue resistance.

The mechanistic origin of this regime-dependent behaviour can be further clarified by comparing the cyclic plastic zone size, r_{pc} , with the microstructural scales of the alloy. Employing Irwin's classical fracture mechanics model for cyclic plasticity under plane strain conditions ($r_{pc} \approx \frac{1}{6\pi} \left(\frac{\Delta K}{\sigma_y} \right)^2$), the estimated r_{pc} at the ΔK_{th} is in the range of 1.3–2.5 μm . This dimension is comparable to the individual cell size, 0.6–0.7 μm , and the width of the grains, $\approx 5 \mu\text{m}$. In this regime, the crack tip interacts with microstructural barriers, such as L2₁-decorated cell boundaries, making the propagation sensitive to the build orientation.

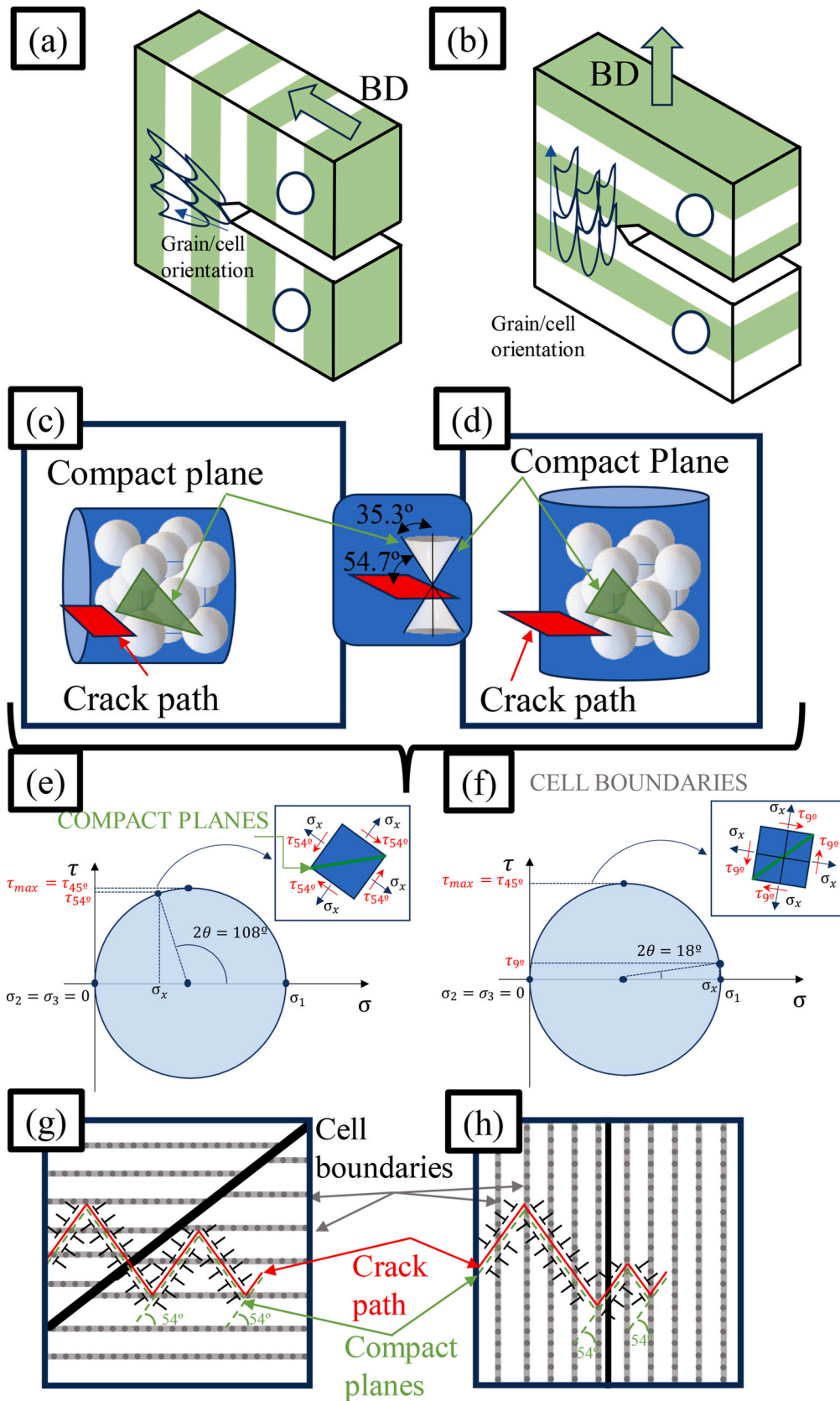


Fig. 15. Schematic representation of the crystallographic orientation of the layers pattern and crack path for (a) 0° orientation, (b) 90° orientation, Crack path interaction with the cellular structure and the compact planes for (c) 0° orientation, (d) 90° orientation, (g) Mohr circle for 54° compact plane stress states in the 0° and 90° orientation, and representation of the crack path interaction with the cell boundaries and the compact planes for (f) 0° orientation and (g) 90° orientation.

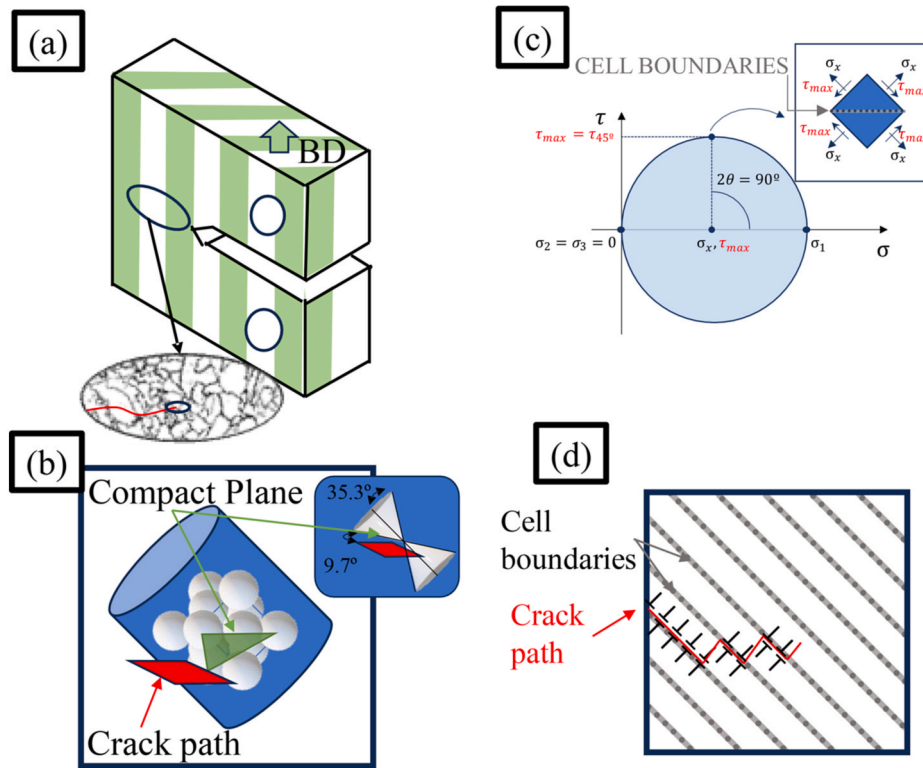


Fig. 16. Schematic representation of the crystallographic orientation of the layers pattern and crack path for (a) 45° orientation, (b) crack path interaction with the cellular structure and the compact planes, (c) Mohr circle for 9° compact plane stress states, (d) Mohr circle for 45° cell boundaries, and (e) representation of the crack path interaction with the cell boundaries and the compact planes.

As ΔK increases into the Paris regime, the r_{pc} expands significantly, encompassing multiple cells and grains. This leads to a volumetric averaging of the microstructural resistance, explaining the convergence of the FCGR curves and the diminishing influence of orientation at higher driving forces.

Therefore, it is proposed that the fatigue propagation mode is not fixed but is a direct consequence of the underlying texture. The mixed-mode behaviours reported in the literature can thus be understood as a dynamic competition between these paths, where the crack may switch between transgranular and intergranular modes depending on the crystallographic orientation of the individual grains it encounters.

4. Conclusions

This research presents a systematic investigation into the influence of build orientation on the fatigue crack growth resistance in the near-threshold regimen of the HEA 8677. The key findings are summarised as follows.

- Dual-phase microstructure and texture. TEM patterns confirm a dual-phase structure in all samples, consisting of an FCC matrix and a coherent L_{12} phase. The build direction influences the crystallographic texture. An L_{21} phase precipitates at cellular and grain boundaries during one step aging.
- Pronounced fatigue anisotropy. The alloy exhibit significant anisotropy in its near threshold fatigue resistance. The 45° orientation possesses a superior fatigue threshold $\Delta K_{th} \approx 6.82 \text{ MPa}\sqrt{\text{m}}$, approximately 40% higher than the 0° $\Delta K_{th} \approx 4.83 \text{ MPa}\sqrt{\text{m}}$, and 90° orientations $\Delta K_{th} \approx 5.17 \text{ MPa}\sqrt{\text{m}}$.
- Defect tolerance and microstructural links. The enhanced defect tolerance of the 45° orientation, quantified by its larger characteristic length, $l_0 = 72 \mu\text{m}$, in the K-T diagram, compared with the 0°, $l_0 = 36 \mu\text{m}$, and 90°, $l_0 = 41 \mu\text{m}$, orientations.

- Direct fractographic and microstructural evidence. The fatigue crack surfaces exhibit a transition from low-energy crystallographic facets in the 0°/90° orientations to cyclic plastic tearing and cell detachment in the 45° orientation. Furthermore, TEM analysis confirms that the cyclic plastic tearing, inter-cellular path in the 45° sample is associated with intense dislocation pile-ups at L_{21} -decorated cell boundaries, an effect absent in the other orientations.
- Refined role of the L_{21} Phase. In the near-threshold condition, the crack-tip driving force is insufficient to cause the decohesion reported in high-strain tensile studies. Instead, L_{21} particles act as obstacles that contribute to toughness by inducing dislocation pile-ups and forcing crack deflection, a mechanism activated only in the 45° orientation.
- Proposed governing mechanism identified. The fatigue anisotropic mechanisms are proposed to be governed by two distinct crack propagation paths, determined by the macroscopic alignment of the microstructural barriers, given that transcellular slip is active in all orientations $m \sim 0.45\text{--}0.46$.
 - o In the 0° and 90° orientations, a favourable alignment of $\{111\}$ slip planes creates a transcellular path, leading to low-energy faceted fatigue propagation and lower fatigue resistance.
 - o In the 45° orientation, enhanced fatigue resistance arises from the geometric alignment of the L_{21} -decorated cell boundaries, which act as barriers. This forces the crack to propagate intercellularly, promoting intense induced plasticity that leads to cyclic plastic tearing.

CRediT authorship contribution statement

Pablo M. Cerezo: Conceptualization, Formal analysis, Investigation, Methodology, Software, Validation, Visualization, Writing – original draft, Writing – review & editing. **Teresa Guraya:** Data curation, Formal analysis, Funding acquisition, Investigation, Project administration,

Supervision, Writing – review & editing. **Jose A. Aguilera**: Investigation, Software, Writing – review & editing. **Alejandro S. Cruces**: Conceptualization, Methodology, Project administration, Supervision, Validation, Writing – review & editing. **Zhangwei Chen**: Methodology, Resources, Validation, Writing – review & editing. **Zhiyuan Liu**: Conceptualization, Methodology, Resources, Writing – review & editing. **Zhan Wen Chen**: Conceptualization, Formal analysis, Funding acquisition, Methodology, Project administration, Supervision, Validation, Writing – review & editing. **Pablo Lopez-Crespo**: Conceptualization, Formal analysis, Funding acquisition, Investigation, Methodology, Resources, Supervision, Validation, Writing – review & editing.

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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Data availability

Data will be made available on request.

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