

Elevated-temperature fatigue behavior and microstructure based cumulative damage evaluation of additive manufacturing superalloy under variable amplitude loading

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Abstract:

Fatigue properties under service conditions are a critical barrier to the reliable application of additive manufacturing (AM) metals. Yet, the associated damage mechanisms and life evaluation approaches, particularly at long term, elevated temperature and variable amplitude (VA) loading, are almost unclear. To address these, high and very-high cycle fatigue VA tests and meso-microscale analyses were performed to investigate damage mechanism of a laser powder bed fused superalloy with heat treatment at service temperature of 650 °C, and a microstructure based cumulative damage evaluation approach was proposed. Results show that interior failures characterized by defect-assisted faceted cracking are predominant. VA loading tends to sequentially activate multiple defects, resulting in competitive multi-site crack nucleation. Increased stress levels accelerate crack growth, leading to the formation of localized rough growth areas and crack deflection. Both primary and secondary cracks grow transgranularly, with crack paths showing negligible dependence on grain orientation. The interior crack nucleation and growth mechanisms under VA loading are elucidated. A cumulative damage evaluation model incorporating the remaining life factor, correlation function transformation, and a reconstructed stress-life relationship was developed, with the prediction results being in close accord with the experimental data under VA loading. These findings provide new insights into the interior crack nucleation and growth mechanisms in AM superalloys and offer a predictive framework for fatigue life estimation under realistic service conditions.

Keywords:

Laser powder bed fusion; Nickel-based superalloy; Very high cycle fatigue; Variable amplitude loading; Cumulative damage model.

Introduction

Additive manufacturing (AM) technology is increasingly employed in the production of load-bearing components in aerospace, energy, and other advanced engineering sectors, due to its ability to fabricate complex geometries with high design flexibility [1, 2, 3]. However, the complex thermal cycles and rapid cooling inherent to AM processes result in unique microstructures [4, 5], making it critical to specify mechanical properties in service environments, particularly fatigue properties [6], which are essential for ensuring long-term service safety [7]. Nickel-based superalloys are widely used in aero-engine and gas turbine blade components due to their excellent high-temperature performance [8]. Laser powder bed fusion (LPBF) offers a promising route for manufacturing such alloys with complex geometries [9]. However, the high thermal gradients and complex melt pool dynamics inherent to the LPBF process introduce defects such as pores, lack of fusion, and inclusions [10], that contribute to fatigue life scatter and early crack initiation [11, 12].

To ensure long-term service safety, the role of these risk factors in the fatigue failure mechanism needs to be further clarified, especially in the high-cycle fatigue (HCF) and very-high-cycle fatigue (VHCF) regimes. The fatigue behavior and failure mechanisms of as-built LPBF nickel-based superalloy were systematically investigated using multi-scale analytical techniques in previous study [13]. Yang et al. [14] conducted ultrasonic fatigue tests on LPBF Inconel 718 and found a competitive relationship between surface and interior failures at decreasing stress levels. In addition, the fatigue sensitivity was significantly affected by defect type, geometry, and location. Macoretta et al. [15] investigated the influence of surface roughness and size effects on the fatigue strength in the HCF regime. Based on the results, a fatigue strength prediction model accounting for surface failure effects was proposed. Shi et al. [16] performed a detailed analysis of interior failure cracks, revealing the influence of precipitation phases on fine grain formation and crack nucleation under VHCF state. Additionally, the high local residual stresses generated during the LPBF process further promote crack nucleation and growth [17].

Integrating experimental and semi-analytical methods to optimize LPBF process parameters is expected to enhance achieving a favorable balance among productivity, metallurgical quality, and mechanical properties [18]. However, solely relying on process parameter optimization to ensure component quality remains technically challenging [19]. Post-heat treatment can improve

1 mechanical properties by promoting microstructural homogenization, relieving residual stress, and
2 inducing the precipitation of strengthening phases [20]. After solution-aging, γ'' and γ' phases
3 precipitate, enhancing the fatigue strength of LPBF nickel-based superalloy through precipitation
4 strengthening and dislocation pinning mechanisms [21]. Solution-aged LPBF nickel-based
5 superalloy exhibits consistently higher fatigue life than as-built specimens under the same stress
6 level in HCF and VHCF regimes [22]. In addition, homogenization and hot isostatic pressing (HIP)
7 have been incorporated into the subsequent heat treatment process to promote the uniform
8 distribution of alloying elements and to close small defects in the material, thereby further
9 improving the fatigue performance [23].

10 Potential applications for LPBF nickel-based superalloy are hot end components of aero-
11 engines and gas turbines, such as turbine blades and compressor disks. These components are
12 typically subjected to complex alternating loads during actual service and are exposed to elevated
13 temperatures in the range of 600 °C to 700 °C for long term, with 650 °C being one of the most
14 representative and critical service temperatures. Recent findings suggest LPBF nickel-based
15 superalloys retain good fatigue performance at elevated temperatures [24, 25]. However, under high
16 temperature and long-life conditions, surface oxidation weakens the dominance of surface crack
17 nucleation, making interior failure the primary fatigue failure mode [26]. The interior failure
18 behaviors under CA loading at 650 °C were reported and a fatigue life prediction model that
19 integrates defect characteristics, stress ratio and temperature was developed [27]. Wang et al. [28]
20 showed that 78.6% of fatigue failures originated from defects, with fatigue life exhibiting significant
21 scatter at lower stress levels. The study also employed X-ray computed tomography to capture
22 defect aggregation and growth. Radhakrishnan et al. [29] investigated the effect of γ'' precipitate
23 size at 600 °C on the formation of persistent slip bands and the resulting fatigue resistance. Gruber
24 et al. [30] investigated the fatigue crack growth rate (FCGR) of LPBF Inconel 718 at 650 °C,
25 reporting that the two LPBF variants exhibited comparable FCGR to wrought Inconel 718 at $R=0.1$,
26 and lower FCGR at $R=0.5$.

27 Although considerable progress has been made under CA loading, fatigue prediction under VA
28 conditions remains a key challenge [31]. Traditional models like Palmgren-Miner linear damage
29 rule are limited by their neglect of load sequence and temperature effects [32]. In recent years,
30 research on the fatigue performance of structural components under varying loading spectra has
31 expanded with many cumulative damage models and life prediction methods being proposed. Chen
32

et al. [33] proposed a nonlinear fatigue damage accumulation model that accounts for load sequence effects. This model is based on the damage evolution equation under CA loading and incorporates the damage equivalence principle. Campagnolo et al. [34] extended the peak stress method to assess the fatigue strength of welded joints under VA uniaxial and multiaxial loading conditions. Bjørheim et al. [35] developed a nonlinear model grounded in the fatigue damage envelope theory to evaluate cumulative fatigue damage under multi-stage loading conditions. The model can better reflect the effect of complex loading history on fatigue life and helps to understand the fatigue behavior under VA loading. Shen et al [36] constructed a multifactor Inconel 718 variable fatigue life prediction model incorporating stress amplitude, load sequence, stress ratio and temperature based on CatBoost algorithm. Santos et al [37] used a machine learning approach to predict the fatigue crack growth path and fatigue life of compact tension specimens under VA loading conditions. Most existing research has focused on the fatigue behavior of AM materials under constant amplitude (CA) loading conditions [38, 39]. In contrast, real-world components often experience variable amplitude (VA) loading due to operational factors such as start-stop cycles, rotational speed fluctuations, and thermo-mechanical coupling. Despite its practical importance, VA fatigue behavior—especially at elevated temperatures—remains underexplored. Moreover, predictive models for fatigue life and damage accumulation modeling under VA loading are limited. Therefore, investigating fatigue behavior under VA loading is essential for advancing AM components toward reliable service applications.

In this study, VA fatigue tests are conducted on solution-aged LPBF nickel-based superalloy at 650 °C under HCF and VHCF regimes. A combination of scanning electron microscopy (SEM), three-dimensional ultra-depth of field (3D UDF), and electron backscatter diffraction (EBSD) is used to investigate fracture surfaces and crack evolution. A cumulative damage evaluation model is developed by integrating by integrating the remaining life factor, a transformed correlation function and a reconstructed stress-life relationship, aiming to predict fatigue life while accounting for defect–fracture surface interactions.

2 Experimental

2.1 Material and microstructure characterization

The nickel-based superalloy powder used for the LPBF process was produced via gas atomization. Its chemical composition is listed in Table 1. The SEM micrograph and particle size distribution of the powder are shown in Fig. 1(a). The particles are predominantly spherical with a smooth surface finish, characteristics that render the powder highly suitable for AM process. After measurement, the particle size ranges from 5 to 50 μm , with an average of 16.7 μm . The distribution is further characterized by d_{10} , d_{50} , and d_{90} values of 7.0 μm , 13.5 μm , and 31.0 μm , respectively. A schematic of the LPBF process and the main processing parameters are presented in Fig. 1(b). Cylindrical specimens with a diameter of 16 mm and a height of 145 mm were fabricated using GmbH EOS M280 machine.

Solution aging treatment was employed to enhance the mechanical properties of fabricated specimens. The heat treatment procedure involved solution annealing at 980 °C for 1 h followed by air cooling, and a two-step aging process: 720 °C for 8 h, furnace cooling to 620 °C at a rate of 55 °C/h, holding for another 8 h, and finally air cooling to ambient temperature. Some samples parallel to the building direction were cut from the center of the round bar for microstructure characterization. The sample surfaces were ground using 120 to 2500 grit SiC sandpaper, followed by mechanical polishing to a mirror finish using diamond abrasive paste with 1 μm particle size. The sample surfaces were etched with Nimonic etchant (HNO_3 , HCl , CuCl_2 , and H_2O) for 30 s, and then immediately rinsed with anhydrous ethanol. Microstructure and precipitated phases were examined using optical microscopy (OM, Leica DM6 M) and SEM (FEI Quanta 650). For EBSD analysis, samples were electro-polished in a solution of 80 vol% $\text{C}_2\text{H}_5\text{OH}$ and 20 vol% H_2SO_4 at 30 V for 30 s at 0 °C. EBSD was performed using the FEI Quanta 650 equipped with an Oxford Instruments detector, and the results were analyzed on AZtecCrystal software. Volumetric defects were characterized using an X-ray computed tomography system (XCT, diondo d2). Scans were performed on the thinnest section of the fatigue specimen in cone-beam CT mode. The tomography was conducted at 110 kV and 110 μA , and 1795 projections were reconstructed and visualized in VGStudio MAX 3.4 at a spatial resolution of 2 μm .

Table 1 The chemical compositions of nickel-based superalloy powder (wt%).

Element	Ni	Cr	Nb	Mo	Ti	Al	Co	C	Mn	Fe
wt%	52.16	19.34	5.07	2.96	0.89	0.51	0.049	0.0041	0.0021	balance

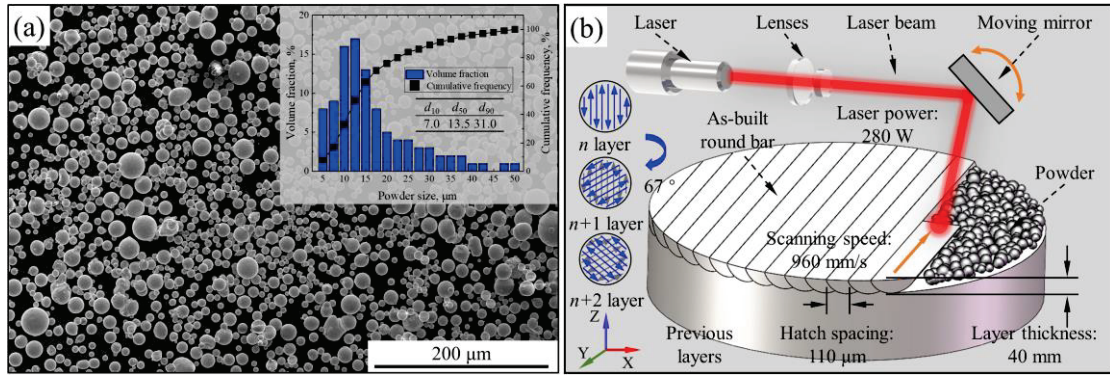


Fig. 1. Micrograph of powder under SEM and LPBF process: (a) nickel-based superalloy powder, and (b) schematic of LPBF process.

2.2 Tensile and fatigue tests

Solution-aged LPBF round bars were machined along the building direction into the tensile and fatigue specimens shown in Fig. 2 to strictly control geometry and surface condition. Prior to testing, specimens were progressively polished with sandpaper from 120 to 2500 grit to eliminate the effects of the machining-hardening layer and surface residual stresses. This process minimizes external variables, enabling a reliable evaluation of elevated-temperature mechanical behavior and failure mechanisms, and ensuring that the results reflect the intrinsic mechanical properties. LPBF nickel-based superalloy is a potential material for critical hot-end components in aero-engines and gas turbines, such as turbine blades and compressor disks, where 650 °C represents a typical and critical service temperature. Fatigue tests at this temperature can provide a realistic simulation of service conditions and help to reveal fatigue failure mechanisms at elevated temperature. Following ASTM E8M-22 and ASTM E21-20 metallic materials tensile standards [40, 41], monotonic tensile tests were conducted at 25 °C and 650 °C on an MTS 809 machine at a strain rate of 1 mm/min to evaluate the tensile properties and guide the design of subsequent fatigue tests. The fatigue specimen geometry and testing procedure followed the ASTM E466-21 metallic materials fatigue standard [42]. Axial-load CA fatigue tests were performed on a QB-100 electromagnetic resonance tester at 25 °C and 650 °C. All fatigue tests were conducted under ambient air conditions. For the 650 °C tests, the electromagnetic resonance tester was equipped with a resistance-heated high-temperature furnace, with three thermocouples positioned at the top, middle, and bottom to continuously monitor temperature. Stress-controlled fatigue loading was applied at a frequency of approximately 100 Hz, with maximum stress amplitudes set below 80% of the 0.2% yield strength. Stress-controlled mode was selected because, in the high-cycle and very-high-cycle fatigue regimes, material behavior is

predominantly elastic with minimal deformation per cycle. Controlling stress amplitude ensures consistent loading for accurate fatigue strength evaluation and better represents the service conditions of many engineering components such as rotating machinery or aerospace structures. The CA fatigue testing procedure at 25 °C was designed based on the corresponding tensile test results. The CA fatigue testing procedure at 650 °C was designed based on the tensile test results at 650 °C, with additional reference to the fatigue results obtained at 25 °C. The CA fatigue tests were conducted at five stress levels ranging from 600 to 800 MPa, ensuring that the fatigue life data span four orders of magnitude (10^4 , 10^5 , 10^6 , 10^7 cycles).

In actual service, turbine blades in aero-engines and gas turbines are subjected to rotational speed fluctuations, thermal shocks, and other forms of VA loading. To simulate these service conditions, VA fatigue tests with gradually increasing load amplitudes simulates service conditions involving progressive speed escalation. In this study, VA fatigue tests are performed using a step-loading mode with gradually increasing stress amplitudes at 650 °C and a stress ratio of $R=0.1$. This approach enables the investigation of cumulative fatigue damage behavior. The machine and testing process for the VA fatigue tests were consistent with those used in the elevated-temperature CA fatigue tests. The stress levels were determined with reference to the CA results, with the maximum stress of the VA loading set equal to that of the CA tests to ensure comparability. Additionally, in order to obtain more test data, seven stress levels were applied in the stress range. In this testing mode, specimens were initially subjected to a stress amplitude σ_1 , and loaded for n_1 cycles, where n_1 was less than the fatigue life at σ_1 . The stress amplitude was then increased to σ_2 , and n_2 cycles were applied. This procedure was repeated—typically with two to four stress levels—until specimen failure. The entire loading process was automatically controlled via computer. Following fatigue testing, all fracture surfaces were examined using SEM and 3D UDF microscopy to analyze crack nucleation and growth behavior, and to identify failure modes.

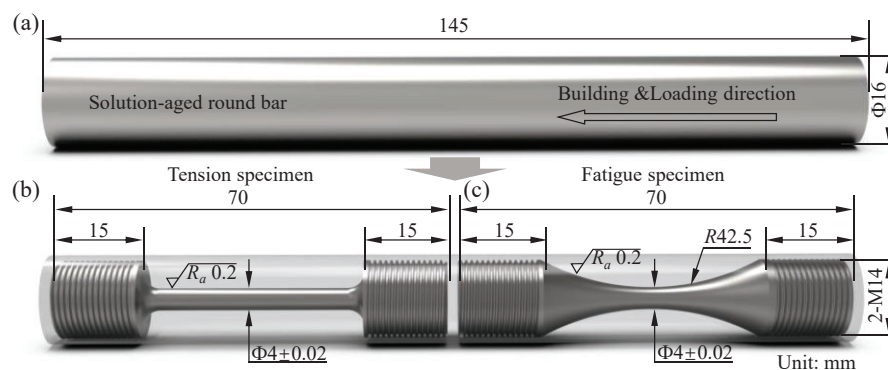


Fig. 2. Shape and size of tensile and fatigue specimens: (a) solution-aged round bar, (b) tensile specimen, and (c) fatigue specimen.

3 Results and discussions

3.1. Microstructure characterization

The microstructure comparison perpendicular to the building direction of the as-built and solution-aged LPBF nickel-based superalloys is shown in Fig. 3. The metallograph of the as-built sample shown in Fig. 3(a) clearly visualizes the streak laser scan traces, reflecting overlapping melt pools typical of LPBF. The steep thermal gradient during LPBF directs grain growth along the heat flow, producing predominantly equiaxed grains with a minor fraction of fine columnar grains, as shown in the inverse pole figure (IPF) of Fig. 3(b). Grain orientation is color-coded, with green representing $\langle 101 \rangle$ orientation. The $\{100\}$ pole figure (PF) exhibits a maximum multiple of uniform distribution (MUD) of 4.41, indicating a weak $\langle 100 \rangle$ texture. Rapid solidification under the thermal gradient leads to fine dendritic structures, with honeycomb intra-grain dendrites and cellular dendrites, as shown in Fig. 3(c). Nb and Mo segregate at cell walls, promoting formation of hard and brittle Laves phase, which diminishes ductility and fatigue crack growth resistance, thereby adversely affecting mechanical properties.

The metallograph of the solution-aged sample is shown in Fig. 3(d). Laser scan traces become blurred due to heat treatment, which promotes long-range elemental diffusion at melt pool boundaries, mitigates segregation from rapid solidification, and gradually eliminates melt pool and interlayer interfaces. Grain morphology concurrently becomes more equiaxed, promoting microstructural uniformity, as shown in Fig. 3(e). The maximum MUD value of the PF remains in the $\{100\}$ plane at 4.03, slightly lower than in the as-built state. The cellular-like structure of as-built samples disappears after solution aging, transforming into a needle-like morphology, as shown in Fig. 3(f). The microstructure of the solution-aged sample consists of γ matrix, γ'/γ'' strengthening phases, δ phase, and a minor fraction of undissolved Laves phase. The face-centered cubic (FCC) γ matrix provides ductility and toughness. Solution annealing dissolves Laves phase, releasing Nb into the γ matrix, while δ phase (Ni_3Nb) precipitates along grain boundaries and gradually into grains, moderately strengthening boundaries by restricting their migration. Subsequent two-step aging induces nanoscale γ' ($\text{Ni}_3(\text{Al}, \text{Ti})$) and γ'' (Ni_3Nb) precipitates, forming coherent lattices indistinguishable under SEM. The L_{12} γ' phase impedes dislocation motion, enhancing high-temperature strength, whereas the DO_{22} γ'' phase contributes more substantially to precipitation

strengthening, preserving excellent mechanical performance under elevated-temperature fatigue [43, 44].

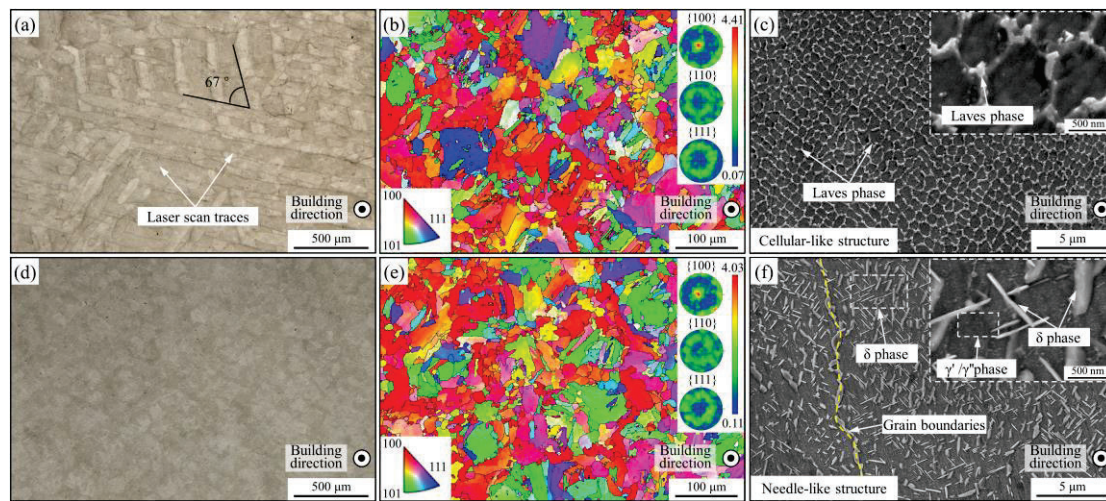


Fig. 3. Comparison of microstructure for as-built and solution-aged LPBF nickel-based superalloy: (a) metallograph of as-built sample, (b) IPF and PF of as-built sample, (c) cellular-like structure with Laves phase, (d) metallograph of solution-aged sample, (e) IPF of solution-aged sample, and (f) needle-like structure with δ , γ' and γ'' phase.

The microstructure parallel to the building direction of the solution-aged LPBF nickel-based superalloy is shown in Fig. 4. The metallograph in Fig. 4(a) reveals a distinct columnar grain morphology in the cross-section parallel to the building direction. These grains are markedly elongated in one direction, exhibiting columnar or lamellar forms. This morphology primarily results from the high temperature gradient and relatively low solidification front growth rate in the melt pool during LPBF, which together favor grain growth along the heat flow direction [45]. Compared with the metallograph, the EBSD results provide a clearer characterization of grain morphology, as shown in Fig. 4(b). Furthermore, layer-by-layer deposition in LPBF induces epitaxial grain growth, with new grains in the new melt pools growing on previously solidified ones, resulting in columnar grains elongating up to several hundred microns. The PF shows a MUD of about 3.45, indicating an absence of pronounced crystallographic texture. High-angle grain boundaries (HAGBs, $>15^\circ$) constitute approximately 76.7% of all grain boundaries. The prevalence of HAGBs contributes to enhanced strength and fatigue resistance at elevated temperature by hindering dislocation motion, promoting grain boundary strengthening, and suppressing crack growth [46]. SEM observations of the microstructure are shown in Fig. 4(c), resembling that in Fig. 3(f), with a needle-like morphology mainly composed of δ , γ' , and γ'' precipitates, all uniformly

distributed within the γ matrix.

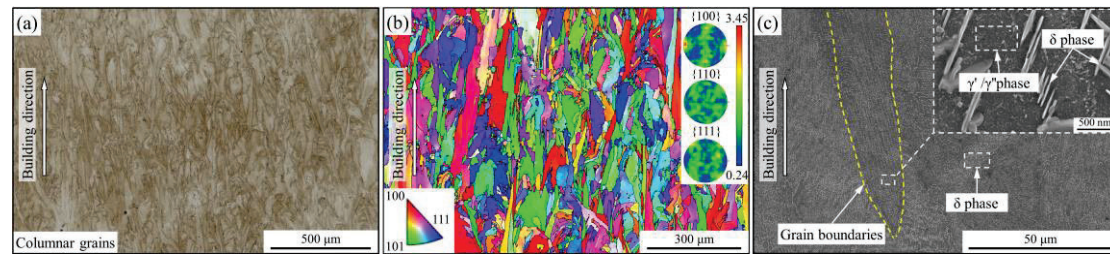


Fig. 4. Longitudinal microstructure of solution-aged LPBF nickel-based superalloy: (a) metallograph, (b) IPF and PF, and (c) microstructure with δ , γ' and γ'' phase.

3.2. Volumetric defects analysis

Volumetric defects are inherent to the LPBF process, where they promote fatigue crack nucleation under cyclic loading and thus degrade fatigue performance. XCT provides non-destructive, three-dimensional visualization and high resolution, which enables effective characterization of volumetric defect morphology and distribution. The XCT results of solution-aged LPBF nickel-based superalloy are shown in Fig. 5. Fig. 5(a) shows the XCT reconstructed cross-section image, revealing a curved contour in the test region. This results from scanning the thinnest part of the fatigue specimen, intentionally designed with a necked geometry to promote crack nucleation in this region during fatigue testing. The spatial distribution and typical morphology of defects are shown in Fig. 5(b), where the colors indicate volume changes. The main volumetric defects include lack of fusion, pores, and inclusions. Lack of fusion results from insufficient laser energy, pores form from trapped gas bubbles, and inclusions arise from oxides or impurities in the powder entrapped during rapid solidification of the melt pool. The equivalent diameter distribution of the defects is shown in Fig. 5(c), which is mainly concentrated in the range of 12.08 to 35.83 μm, with an average diameter of 19.91 μm. Small defects are widely existed inside the specimen, while larger defects may act as preferential locations for fatigue crack nucleation. The sphericity distribution of the defects is shown in Fig. 5(d). Sphericity value ranges from 0.4 to 0.68 with an average value of 0.56. Smaller defects generally exhibit higher sphericity and nearly spherical morphologies, corresponding mainly to pores. With increasing equivalent diameter, the sphericity decreases markedly, indicating that larger defects tend to display irregular or flat morphologies typically associated with lack of fusion.

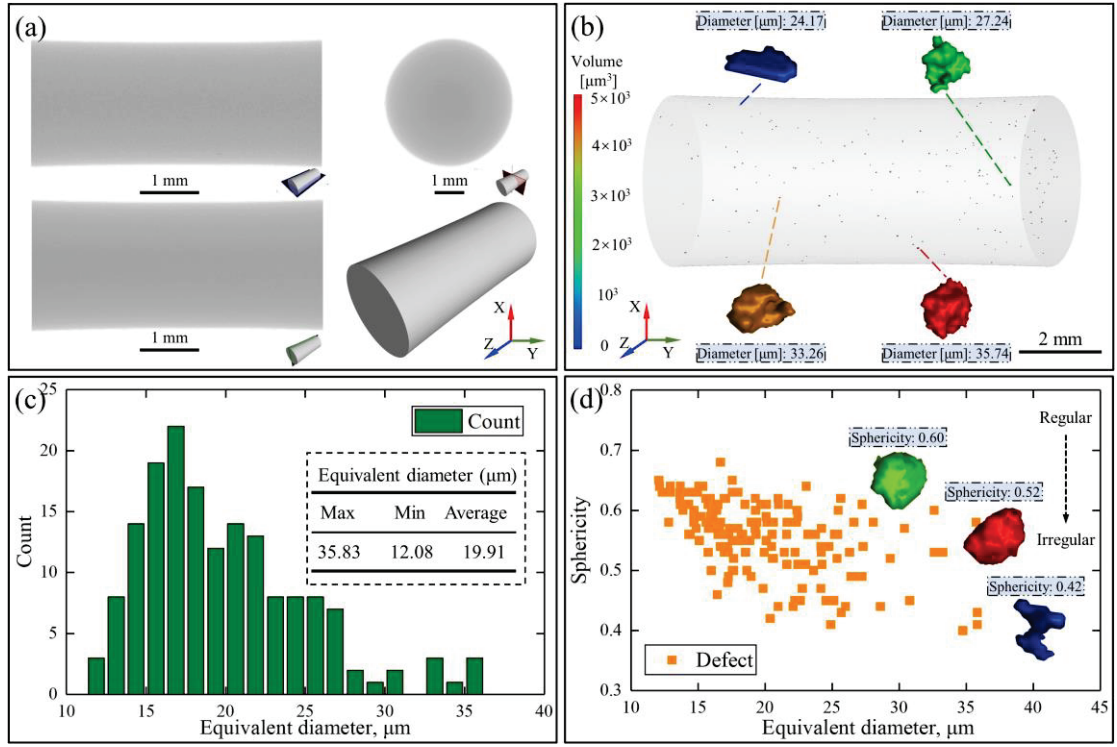


Fig. 5. XCT results of solution-aged LPBF nickel-based superalloy: (a) XCT reconstructed cross-sectional images, (b) spatial distribution of defects and typical defect morphology, (c) equivalent diameter distribution, and (d) sphericity distribution.

3.3 Tensile test results and fractures

The stress–strain curves and tensile properties of solution-aged LPBF nickel-based superalloy specimens at 25 °C and 650 °C are shown in Fig. 6. The yield strength and ultimate strength are approximately 1345 MPa, 1430 MPa at 25°C and 1065 MPa, 1160 MPa at 650 °C, respectively. The static strength decreases at 650 °C due to reduced lattice stability from thermal expansion, coupled with increased dislocation mobility and diffusion. However, the stress–strain curves remain essentially identical in the elastic stage, indicating that the LPBF nickel-based superalloy retains excellent elevated-temperature resistance within the elastic range. In addition, the measured reduction of area ($RA\%$) and elongation ($EL\%$) are approximately 37.0% and 16.0% at 25 °C, and 17.0% and 13.5% at 650 °C, respectively.

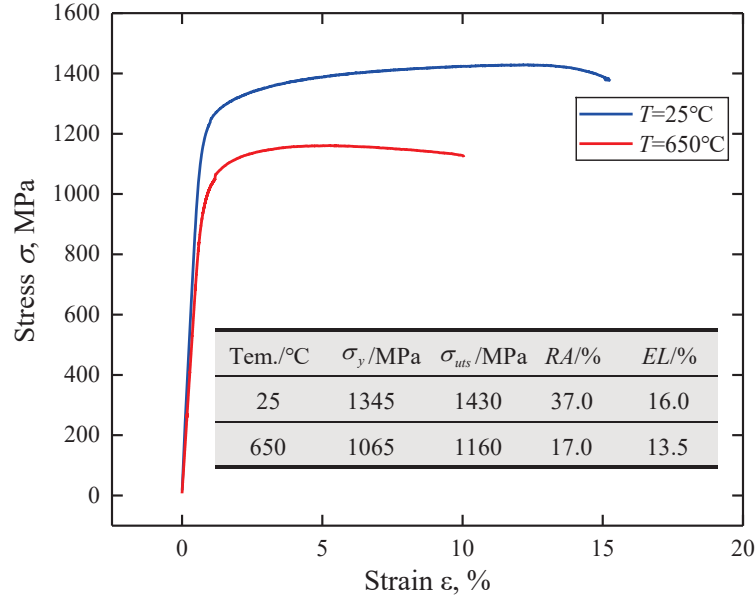


Fig. 6. Stress–strain curves and tensile properties of solution-aged LPBF nickel-based superalloy specimens at 25 °C and 650 °C.

The tensile fracture at 25 °C, shown in Fig. 7(a), can be divided into three typical regions: fibrous region, radial region, and shear lip. A locally magnified view of the fracture surface reveals numerous equiaxed dimples, characteristic of ductile fracture. The local magnified view of the fracture surface shows a large number of equiaxed dimples, which are typical features of ductile fracture. Second-phase particles are observed at the bottom of some dimples and are hypothesized to be precipitated strengthening phases. The tensile fracture at 650 °C is shown in Fig. 7(b). At elevated temperature, the fracture surface is oxidized, leading to the formation of a blue oxide film covering the area. Compared with the fracture at 25 °C, the dimples on the fracture surface at 650 °C are smaller and deeper. Elevated temperatures promote localized plastic deformation around precipitate-strengthened regions while inhibiting extensive void growth, thereby delaying void coalescence and resulting in the formation of finer and deeper dimples.

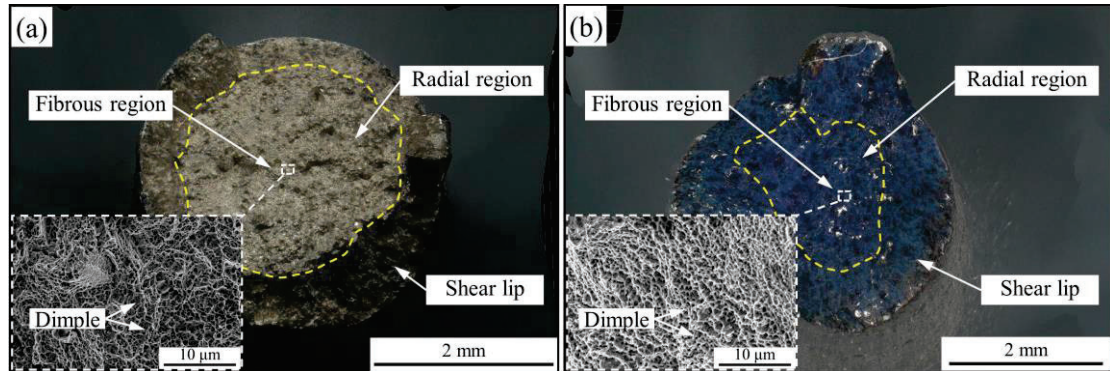


Fig. 7. Tensile fracture morphology of solution-aged LPBF nickel-based superalloy specimens at 25 °C and 650 °C: (a) Tensile fracture with dimple at 25 °C, and (b) Tensile fracture with dimple at 650 °C.

3.4 Fatigue test results

The load settings and fatigue test results of solution-aged LPBF nickel-based superalloy under CA loading with a stress ratio $R=0.1$ at 25 °C and 650 °C, and VA loading with a stress ratio $R=0.1$ at 650 °C, are listed in Tables 2, 3, and 4, respectively. The corresponding $S-N$ curves, describing the relationship between the maximum stress ($\sigma_{\max}$) and fatigue life (N_f), are shown in Fig. 8 and Fig. 9, respectively. The observed fatigue lives range from 10^4 to 10^9 cycles, showing a decreasing trend in fatigue strength, indicating the absence of a conventional fatigue limit at 10^7 cycles. From the CA fatigue data shown in Fig. 8, the overall dispersion of the data points at 25 °C and 650 °C is small, indicating that this LPBF nickel-based superalloy exhibits excellent fatigue resistance at elevated temperature. The CA fatigue data points at 25 °C and 650 °C are fitted with two separate $S-N$ curves, respectively. Fatigue failure modes can be classified into two categories depending on the crack initiation site: surface failure and interior failure. At 25 °C, surface failure predominates, whereas at 650 °C, interior failure is the dominant mode. The formation of a tight oxide film on the specimen surface at elevated temperature may be one of the most important factors contributing to the failure mode shift. This oxide film can passivate minute surface scratches to some extent, thus attenuating the local stress concentration effect. However, at high stress levels, the oxide film is prone to rupture, which in turn promotes rapid crack growth. On the contrary, as the stress level decreases, the oxide film can effectively inhibit the surface crack nucleation, so that the volumetric defects inside the specimen gradually become the main source of crack nucleation. As can be seen from Fig. 8, Tables 2 and 3, the fatigue life at 25 °C is usually slightly longer at the same high stress level, and the failures mostly originate from surface crack nucleation. As the stress level decreases, the fatigue life at 650 °C is slightly extended in individual cases, mainly related to the interior failure mechanism. Crack nucleation from volumetric defects within the material can extend fatigue life to some extent, as the crack must travel a longer path and pass through more grains or microstructural features before propagating to final fracture. Prof. Stinville et al. also provided supportive evidence for this trend in their research [47], and more fatigue tests to validate it in subsequent studies are necessary.

In Fig. 9, the CA and VA test results are plotted as dashed and solid lines, respectively. The

solution-aged LPBF superalloy exhibits higher fatigue life under VA loading compared to CA loading at equivalent maximum stress levels. This behavior can be attributed to the fundamental differences in damage accumulation between the two loading modes. Under CA loading, the total fatigue life is governed by a single applied stress amplitude. In contrast, VA loading exposes the material to multiple stress levels, where lower amplitudes dominate early stages of the loading history. Cumulative fatigue damage is thus distributed across several stress levels, effectively extending the overall fatigue life even when the highest stress level is the same as in CA loading. The transition in failure mechanisms is also evident as fatigue life increases. At 650 °C, as the life approaches or exceeds 10^6 cycles, a shift from surface-initiated failure to subsurface and interior failure is observed. This trend indicates that internal defects increasingly govern crack nucleation and growth in the HCF and VHCF regimes, emphasizing the importance of internal quality in additively manufactured components for high temperature fatigue-critical applications. Fractographic observations of representative fracture surfaces are discussed in the following section. In this study, fatigue tests were carried out at several stress levels, and the results obtained reveal the fatigue properties and failure behavior of the materials to a certain extent. However, more experiments need to be conducted to obtain a larger dataset in order to comprehensively and reliably assess the fatigue performance of materials. This will help to characterize the distribution and variability of fatigue life more accurately, thus enhancing the credibility and generalizability of the results.

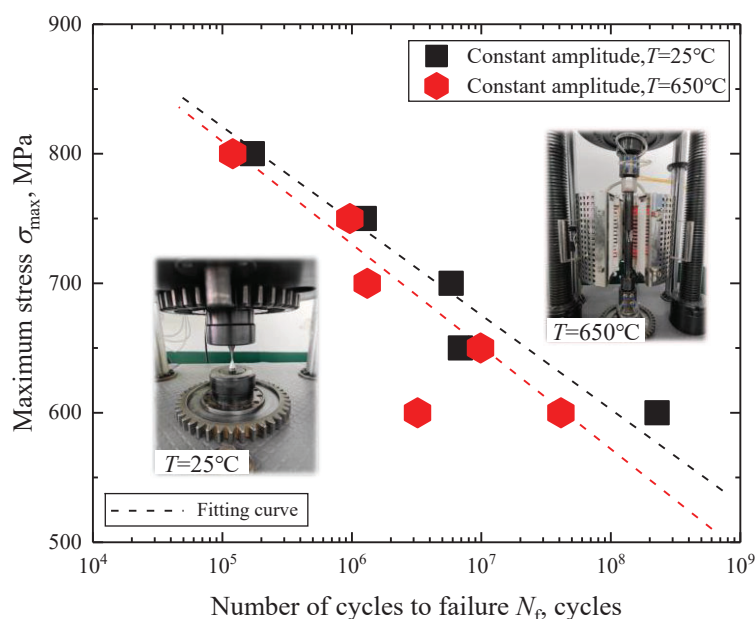


Fig. 8. *S-N* curve of solution-aged LPBF nickel-based superalloy specimens under CA loading at

25 °C and 650 °C.

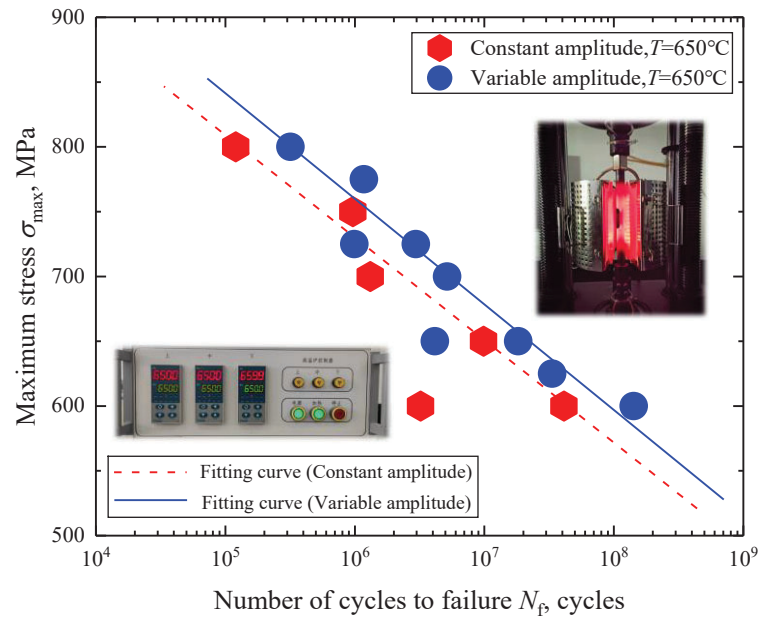


Fig. 9. *S-N* curves of solution-aged LPBF nickel-based superalloy specimens under CA and VA loading at 650 °C.

Table 2 Fatigue testing results under CA loading with stress ratio $R=0.1$ at 25 °C.

Spec. No.	$\sigma_{\max}/\text{MPa}$	N_f/cycles	Failure modes
CA 1	800	1.704E5	Surface failure
CA 2	750	1.247E6	Surface failure
CA 3	700	5.850E6	Surface failure
CA 4	650	6.895E6	Interior failure
CA 5	600	2.275E8	Interior failure

Table 3 Fatigue testing results under CA loading with stress ratio $R=0.1$ at 650 °C.

Spec. No.	$\sigma_{\max}/\text{MPa}$	N_f/cycles	Failure modes
CA 1	800	1.202E5	Surface failure
CA 2	750	9.642E5	Interior failure
CA 3	700	1.316E6	Interior failure
CA 4	650	9.868E6	Interior failure
CA 5	650	3.220E6	Interior failure
CA 6	600	4.120E7	Interior failure

Table 4 Fatigue testing results under VA loading with stress ratio $R=0.1$ at 650 °C.

Spec. No.	Load levels/MPa	Related cycles/cycles	N_f/cycles	Failure mode
VA 1	775→800	2.0E5→failure	3.185E5	Surface failure
VA 2	750→775	5.0E5→failure	1.176E6	Interior failure
VA 3	700→725	5.0E5→failure	9.912E5	Interior failure
VA 4	675→700→725	5.0E5→5.0E5→failure	2.955E6	Interior failure
VA 5	675→700	3.0E6→failure	5.154E6	Interior failure
VA 6	625→650	3.0E6→failure	4.149E6	Interior failure

VA 7	600→625→650	1.0E7→5.0E6→failure	1.824E7	Interior failure
VA 8	550→575→600→625	5.0E6→5.0E6→1.0E7 →failure	3.338E7	Interior failure
VA 9	550→575→600	5.0E7→5.0E7→failure	1.423E8	Interior failure

3.5 3D and 2D morphology of fracture surfaces

The typical fatigue fracture surfaces under CA loading are shown in Fig. 10. The fracture surface at 25 °C shows metallic luster, as shown in Fig. 10(a). According to the crack evolution characteristics at different stages, the fracture displays different colors under the OM, which can be classified into crack nucleation area, crack growth area and final fracture area. In the local enlargement view of Fig. 10(b), radial ridges extending outward from the crack nucleation site are visible in the crack growth area. A pore is identified in the crack growth path, but it is not the origin of the fatigue crack. An irregularly shaped defect is observed in the crack nucleation area, as shown in Fig. 10(c), with incompletely melted powder remaining on its surface. The sharp edges of this defect are susceptible to localized stress concentration under cyclic loading, thus inducing fatigue failure. The cleavage features formed during crack growth are also detected around the defect, indicating that the cracks are affected by brittle fracture during the early growth stage.

The fracture at 650 °C undergoes elevated temperature oxidation, with the surface covered by a purple oxide film, as shown in Fig. 10(d). In the local enlargement of Fig. 10(e), the complex stress state at the crack front under elevated temperature condition promotes the formation of localized secondary cracks in the surrounding material. Crack nucleation is further observed to originate from volumetric defects, which belongs to the interior failure mode. Further magnification of the crack nucleation area shows a nearly spherical pore at the center surrounded by some facets, as shown in Fig. 10(f). During cyclic loading, the stress concentration induced by the pore induces microcracks in the adjacent grains. These microcracks grow along the grains under shear stress, eventually leading to grain rupture and formation of facets. Obvious slip traces can be observed on the facets, which indicates that dislocation motion is activated several times under fatigue cyclic loading, and the material undergoes localized plastic deformation in the process. Further magnification of the crack nucleation area shows a pore at the center, surrounded by multiple facets, as shown in Fig. 10(f). Under cyclic loading, stress concentration around the pore initiates microcracks in the adjacent grains. These microcracks grow through the grains under shear stress, ultimately leading to grain rupture and the formation of facets. Slip traces are observed on the facets, indicating repeated dislocation activity during cyclic loading and localized plastic deformation of

the material.

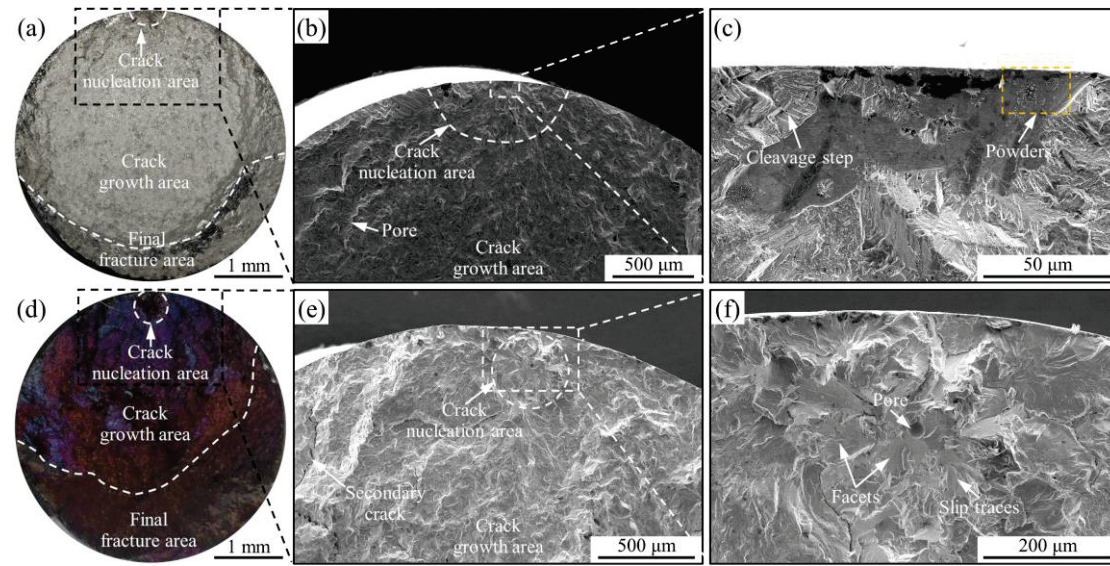


Fig. 10. Typical fatigue fracture surfaces under CA loading at 25 °C and 650 °C: (a) fracture surface at 25 °C ($\sigma_{\max}=700$ MPa, $N_f=5.850E6$), (b) local enlargement, (c) crack nucleation area, (d) fracture surface at 650 °C ($\sigma_{\max}=700$ MPa, $N_f=1.316E6$), (e) local enlargement, (f) crack nucleation area.

Typical fracture morphologies corresponding to subsurface and interior failures under VA loading are shown in Figs. 11-15, covering fatigue lives in both the HCF and VHCF regimes. The observed crack nucleation sources for subsurface and interior failures primarily include inhomogeneous microstructures, keyholes, lack of fusion (LoF) defects, gas pores, and crystallographic facets.

Representative of subsurface fatigue fracture under VA loading is shown in Fig. 11. The 3D morphology of the fracture is presented in Fig. 11(a), where a blue oxide film covers the surface due to oxidation at elevated temperature. The fracture surface can be divided into three distinct zones: crack nucleation, crack growth, and final fracture. The 3D pseudo-color height distribution map shown in Fig. 11(b) reveals the topographic inhomogeneity of the crack nucleation region. The profile of the measurement plane crossing the fracture surface can be nearly represented as a crack growth path, corresponding to a height difference of 287.62 μm .

The 2D SEM image of the crack nucleation and growth zone is shown in Fig. 11(c). A bright white area is evident near the crack origin, resembling the fine grain area (FGA) typically observed in high-strength steels subjected to VHCF [48]. In this study, we define this region as a faceted cracking area (FCA). Streak features indicative of long crack growth appear in the crack growth area, signaling the transition to an unstable fracture stage. An enlarged view of the FCA is provided

in Fig. 11(d), showing no apparent microstructural defects. Fracture analysis indicates that microcracks nucleate from subsurface inhomogeneous microstructure and grow under cyclic loading, forming crystallographic facets. Since the inhomogeneous microstructure is close to the specimen surface and subject to the surface stress gradient, the local stress concentration further exacerbates the crack growth rate when the crack extends to the surface. Therefore, subsurface-initiated failures generally correspond to shorter fatigue lives than interior-origin failures.

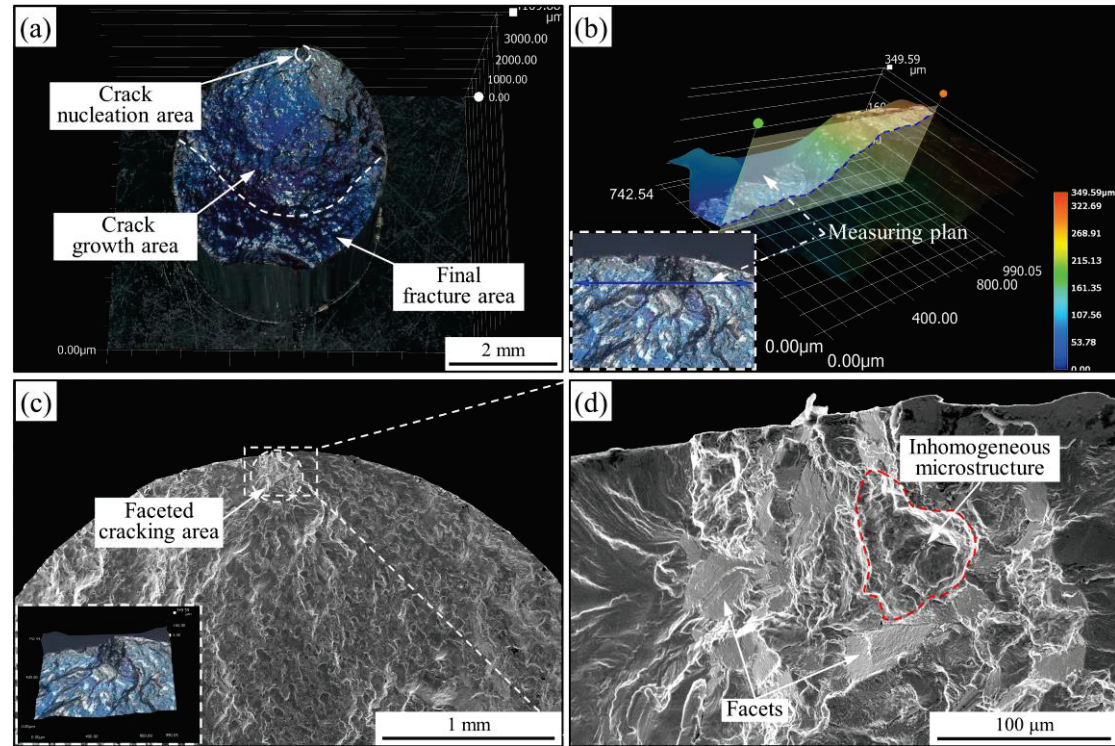


Fig. 11. Typical fatigue fracture surface under VA loading ($\sigma_{\max}=750 \rightarrow 775$ MPa, $N_f=1.176E6$): (a) 3D morphology of fracture surface, (b) pseudo-color map of crack nucleation area, (c) faceted cracking area, and (d) inhomogeneous microstructure.

The typical fatigue fracture caused by keyhole defects under VA loading is shown in Fig. 12. As shown in Fig. 12(a), the 3D fracture surface can also be divided into three characteristic regions: crack nucleation, stable crack growth, and final fracture. The surface is partially covered by yellow and purple oxide films, indicating localized oxidation during high-temperature testing. The crack nucleation and growth areas exhibit markedly different morphologies, which are clearly distinguishable in both 3D UDF and 2D SEM observations. The projected area of the crack nucleation area is quantified using ImageJ software to evaluate its size. Compared to the fracture in Fig. 11(a), the crack nucleation zone in Fig. 12(a) is noticeably larger, likely due to prolonged stable crack growth during the lower-stress stages of the VA loading sequence. The 3D pseudo-color map

of the crack nucleation area in Fig. 12(b) further highlights the height variation across the crack initiation zone, with a measured height difference of 549.51 μm across a cross-sectional profile intersecting the crack origin. OM reveals the crack origin as a bright white spot, indicating local microstructural or compositional heterogeneity compared to the surrounding matrix.

The 2D SEM image in Fig. 12(c) shows the crack nucleation region lacking clear beach marks in the FCA, which is attributed to the relatively narrow stress amplitude increments between loading blocks in the VA test. As a result, the fatigue crack growth rate exhibited only minor variations throughout different stages of the test. The complex stress state at the advancing crack front likely contributed to local stress concentration and the formation of secondary cracks. An enlarged SEM view of the crack initiation zone, shown in Fig. 12(d), reveals a sharp-edged internal defect at the center of the faceted cracking area. This feature closely resembles the keyhole-type defect described in the literature [49]. During LPBF process, high laser power densities cause rapid vaporization of metal powder, generating a vapor jet and associated recoil pressure that destabilize the melt pool. This instability leads to the formation of keyholes. High-magnification observation of the facets near the keyhole defects reveals many slip traces, indicative of slow crack growth during fatigue. These observations confirm that keyhole defects serve as stress concentrators that facilitate the nucleation and growth of fatigue cracks, especially under thermomechanical VA loading conditions.

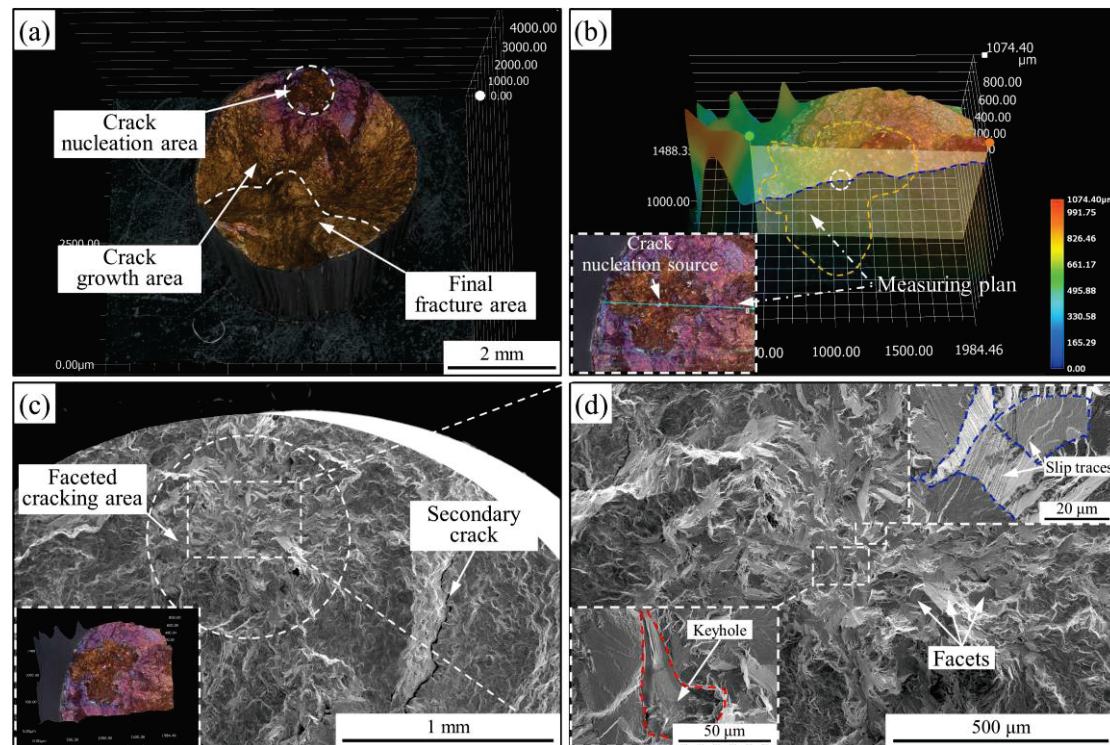


Fig. 12. Typical fatigue fracture surface under VA loading ($\sigma_{\max}=675 \rightarrow 700 \rightarrow 725$ MPa,

$N_f=2.955E6$): (a) 3D morphology of fracture surface, (b) pseudo-color map of crack nucleation area, (c) faceted cracking area, and (d) keyhole.

The typical fatigue fracture morphology caused by LoF under VA loading is shown in Fig. 13. The 3D morphology of the fracture surface, shown in Fig. 13(a), exhibits clear differentiation between the crack nucleation zone, stable crack growth zone, and final fracture zone. Distinct color variations correspond to different fracture stages, while significant plastic deformation in the crack nucleation zone increases surface roughness, resulting in darker contrast under 3D UDF imaging.

The 3D pseudo-color map in Fig. 13(b) reveals significant topographic complexity in the crack nucleation zone, with a height difference of 673.19 μm across the fracture surface along a cross-sectional measurement plane. This considerable elevation difference reflects the tortuous nature of crack propagation under VA loading. The 2D SEM image of the crack nucleation zone, shown in Fig. 13(c), reveals two crack nucleation sources are faintly visible in the center of the FCA. These multiple crack initiation sites likely result from local stress concentrations and microstructural inhomogeneities, both of which are exacerbated by variable loading amplitudes. Under VA loading, simultaneous or independent microcrack nucleation can occur at multiple defects, leading to crack coalescence or competitive propagation. Crack nucleation source I is indistinct under 3D UDF, while its possible inhomogeneous microstructure is observed in SEM. The magnified view of crack nucleation source II is shown in Fig. 13(d). An irregularly shaped defect is found, accompanied by partially unmelted powder particles at its base. Based on morphology and prior literature [50], this defect is attributed to LoF. LoF defects typically arise due to suboptimal LPBF processing parameters such as insufficient laser power, high scan speed, or nonuniform powder distribution, which hinder full melting and fusion of adjacent powder layers. A pore is also observed on the facet adjacent to the LoF, highlighting the spatial clustering of defects in some regions.

Due to their irregular geometry and sharp edges, LoF defects act as potent stress concentrators that significantly reduce fatigue life by accelerating crack nucleation. In comparison to smoother, more equiaxed pores, LoF defects are generally considered more detrimental to fatigue performance in AM components. These findings underscore the critical importance of defect control in LPBF processing for high-temperature fatigue-critical applications.

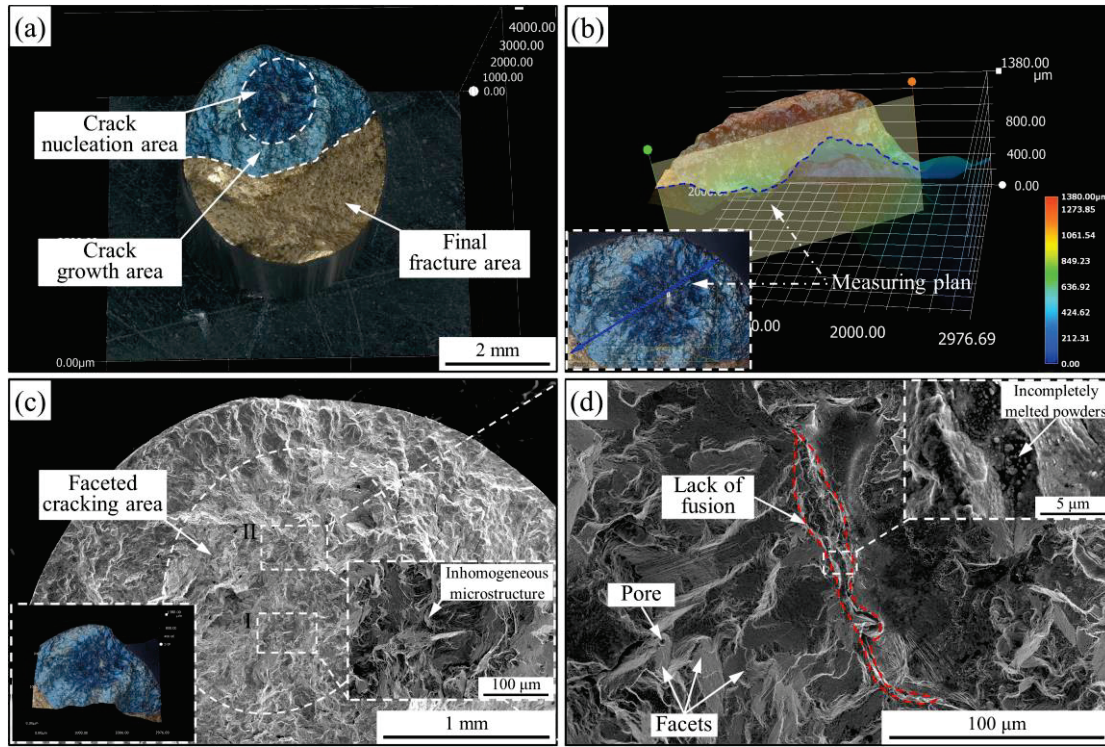


Fig. 13. Typical fatigue fracture surface under VA loading ($\sigma_{\max}=675 \rightarrow 700$ MPa, $N_f=5.154E6$): (a) 3D morphology of fracture surface, (b) pseudo-color map of crack nucleation area, (c) faceted cracking area, and (d) lack of fusion.

A representative fatigue fracture surface exhibiting competing failure behavior under VA loading is shown in Fig. 14. The 3D and 2D morphologies of the fracture surface are shown in Fig. 14(a), which reveal features associated with both interior and surface crack nucleation. Accordingly, the fracture surface can be divided into four distinct regions: interior crack nucleation area, surface crack nucleation area, stable crack growth area and final fracture area. When cracks originating from both the surface and interior regions propagate and eventually intersect, a competing failure mechanism emerges. This interaction alters the local stress field, intensifying the energy release rate and promoting crack deflection or bifurcation. Such interactions may further drive the nucleation of secondary cracks.

The 3D pseudo-color map of the interior crack nucleation area is shown in Fig. 14(b), revealing morphological inhomogeneity and localized elevation differences associated with crack evolution. Upon higher magnification, an FCA is identified, as shown in Fig. 14(c). A circular pore is observed at the center of the FCA, surrounded by multiple crystallographic facets (Fig. 14(d)). The pore likely formed due to gas entrapment during the LPBF process. During laser melting, protective gas absorption or localized metal vaporization may lead to gas bubble formation. When the cooling rate

is too high or the molten metal lacks sufficient fluidity, the trapped gas cannot escape before solidification, leading to the formation of isolated spherical pores within the solidified matrix [51].

Magnified SEM imaging of the pore edges reveal significant local plastic deformation, indicating the accumulation of cyclic damage. Stress concentration around the pore promotes dislocation pile-up, leading to the formation of slip bands that further evolve into microcracks. Under applied cyclic shear stresses, these microcracks grow transgranularly, fracturing individual grains and forming faceted features. High-magnification observations of these facets shows slip traces decorated with fine nanoscale precipitates. These particles are presumed to be γ' and γ'' phases, which are known to enhance matrix strength via precipitation hardening. γ' and γ'' phases contribute to the alloy's elevated-temperature fatigue resistance by restricting the dislocation motion.

These findings highlight that the interplay between microstructural features and surface or interior defects—particularly pores—governs crack nucleation and growth in the VHCF regime under VA loading.

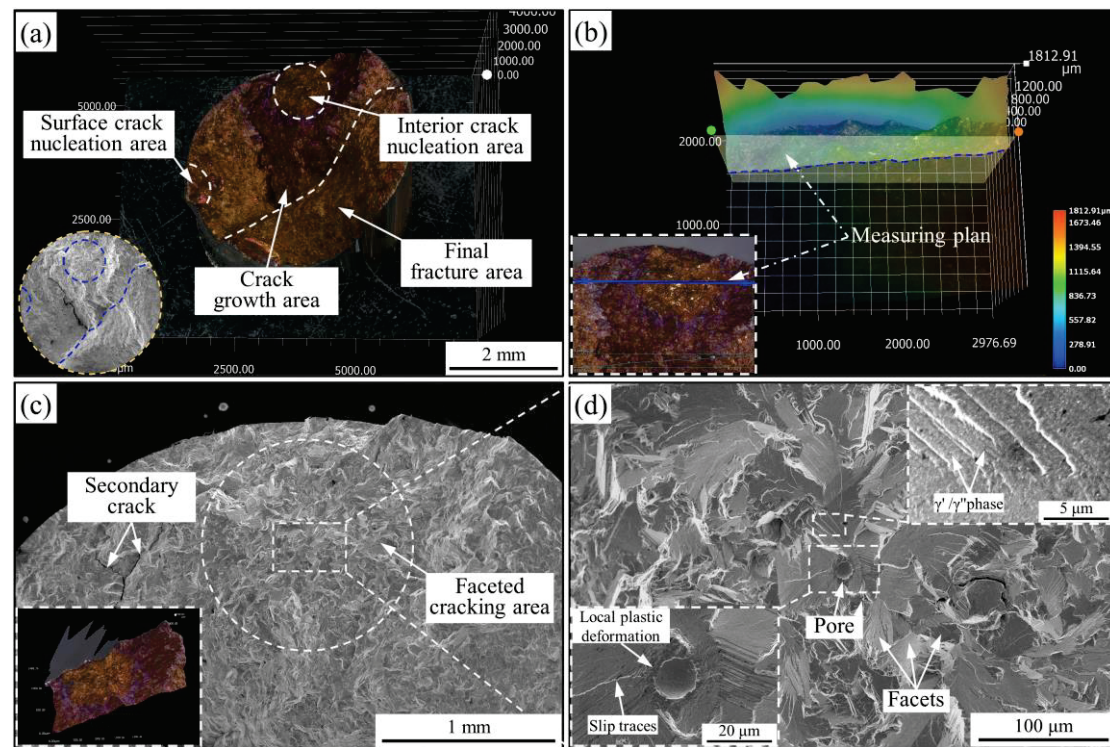


Fig. 14. Typical fatigue fracture surface under VA loading ($\sigma_{\max}=550 \rightarrow 575 \rightarrow 600 \rightarrow 625$ MPa, $N_F=3.338E7$): (a) 3D morphology of fracture surface, (b) pseudo-color map of interior crack nucleation area, (c) faceted cracking area, and (d) pore.

A representative VHCF fracture surface with fatigue life exceeding 10^8 cycles under VA loading is shown in Fig. 15. The overall 3D morphology of the fracture surface and the 3D pseudo-

1 color height distribution map of the crack nucleation area are presented in Fig. 15(a) and (b),
2 respectively. The fracture surface exhibits a substantial height variation of approximately 2262.13
3 μm , indicative of a complex crack propagation history. The 2D SEM morphology of the crack
4 nucleation area is shown in Fig. 15(c), where the nucleation site is near the specimen center,
5 consistent with interior crack initiation. The prolonged crack growth duration at lower stress levels
6 results in the development of a relatively large FCA, characteristic of VHCF behavior.
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11 Unlike the fatigue fractures caused by inhomogeneous microstructures or defects (Fig. 11–Fig.
12 14), no apparent defect is observed at the center the FCA in Fig. 15(d). Further magnification reveals
13 facets at the crack nucleation site. The absence of a discernible defect suggests that fatigue crack
14 initiation in this case is driven by microstructural factors rather than geometric or metallurgical
15 discontinuities. Specifically, variations in grain orientation influence the accommodation of local
16 elastic-plastic deformation, inducing local stress concentrations that activate specific slip systems
17 under cyclic loading. These stress concentration sites act as microcrack nucleation points even in
18 the absence of traditional defects.
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27 Although a pore is observed near the crack nucleation source, morphological and fractographic
28 evidence indicates that it did not directly trigger crack nucleation. This further supports the assertion
29 that microstructure-controlled crack nucleation dominates the VHCF regime under VA loading. The
30 observed fracture features highlight the role of local crystallographic features and microstructural
31 evolution as key mechanisms governing crack formation and early growth.
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37 The coexistence of defect-driven and microstructure-induced failure modes underscores the
38 complex fatigue behavior of LPBF nickel-based superalloys at elevated temperatures, emphasizing
39 the critical role of microstructural control and texture optimization in enhancing their VHCF
40 performance.
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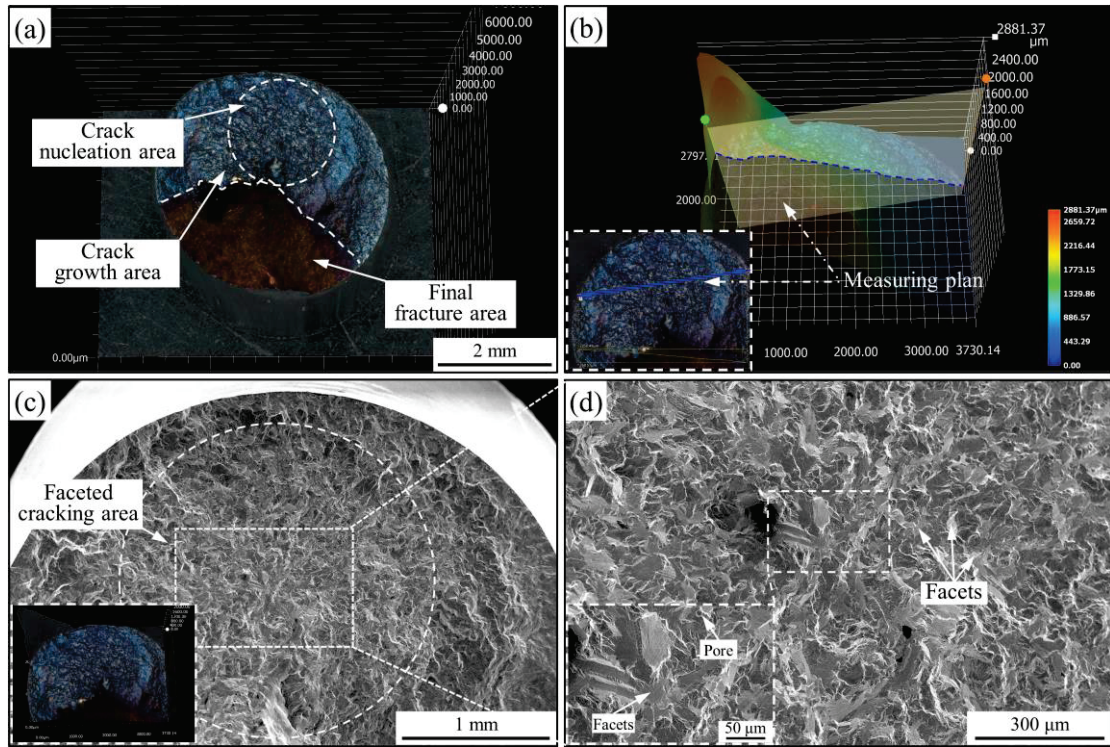


Fig. 15. Typical fatigue fracture surface under VA loading ($\sigma_{\max}=550 \rightarrow 575 \rightarrow 600$ MPa, $N_f=1.423E8$): (a) 3D morphology of fracture surface, (b) pseudo-color map of crack nucleation area, (c) faceted cracking area, and (d) facets.

3.6 EBSD testing of interior fatigue failure surface

To further elucidate the crack growth behavior under VA loading in the VHCF regime at 650 °C, the fracture surface shown in Fig. 14 is subjected to EBSD testing. The procedure and testing area are illustrated Fig. 16(a). The cutting plan for wire electrical discharge machining was selected on a plane perpendicular to the projected fracture surface, parallel to the building direction. Mechanical and electrolytic polishing were used to prepare high quality surface for EBSD analysis. The EBSD measurements were performed on a region beneath the fracture surface.

In addition to the primary crack on the fracture surface, secondary cracks oriented along the loading direction and propagating into the interior of columnar grains were also unexpectedly observed, as shown in Fig. 16(b). Such secondary crack growth behavior is rarely reported in AM materials. Under tensile-tensile cyclic loading, the continuous tensile stress maintains microcracks in an open state. Primary cracks usually propagate across multiple grains along inclined path relative to the loading direction. The secondary crack growth path also forms an angle with the loading direction, as the driving force arises from the component of the principal stress resolved along the crack growth direction. The low grain boundary density of the columnar grains aligned along the

1 building direction reduces the resistance to crack growth, promoting continuous crack growth along
2 the grain's longitudinal axis [52]. The secondary crack growth path exhibits significant deflection,
3 likely due to the variation in stress amplitude under VA loading, as well as the modulation of crack
4 growth by local microstructural features.
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8 The IPF map of the testing area, shown in Fig. 16(c), reveals no significant changes in grain
9 orientation distribution or overall texture compared to the initial state in Fig. 4(b), indicating that
10 fatigue loading did not induce notable texture evolution or preferential orientation reorganization.
11 Both primary and secondary cracks show no clear correlation with grain orientation, suggesting that
12 under the current loading conditions, crack growth is predominantly driven by external stresses,
13 with minimal influence from grain texture. Crack growth occurs predominantly via transgranular
14 fracture mode. Fine grains detached from the matrix after grain fracture are observed along the
15 secondary crack growth path, indicating that severe plastic deformation and localized stress
16 concentration occurred under cyclic loading, which reduced the intergranular bonding strength.
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19 The Kernel Average Misorientation (KAM) map in Fig. 16(d), which reflects local orientation
20 gradients related to dislocation density, residual stress, and localized plastic deformation [53],
21 reveals elevated KAM values along both primary and secondary crack paths. These elevated indicate
22 localized strain concentration and orientation gradients associated during crack growth process. This
23 behavior arises from the repeated exposure of the crack tip region to high stress and strain cycles
24 under fatigue loading, leading to dislocation accumulation and localized plastic deformation.
25 Furthermore, precipitation strengthening in the material impedes dislocation motion, thereby further
26 promoting strain localization in the vicinity of the crack path. Some columnar grains exhibit a
27 distinct alignment of high KAM values along their long-axis—an effect attributed to the nonuniform
28 plastic deformation induced by the rapid cooling process during LPBF process. As highlighted by
29 the dashed box in Fig. 16(d), these regions may indirectly influence the secondary crack growth
30 path.
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33 The Schmid factor (SF) map, shown in Fig. 16(e), characterizes the effectiveness of applied
34 loading in generating shear stress on the primary slip system $\{111\}\langle 110 \rangle$ of the nickel-based
35 superalloy. Some grains along the primary and secondary crack paths exhibit high SF values,
36 indicating easier slip systems activation under cyclic loading, thereby facilitating localized plastic
37 deformation and promoting crack growth. Crack growth tends to follow paths associated with
38 regions of high SF values. In addition, since there is an angle between the crack path and the loading
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direction, the distribution of SF influences the resolved into effective shear stress and hence preferred direction of crack growth.

Two notable deflections were observed along the secondary crack, with enlarged views presented in Fig. 16(f). These deflections may be associated with variations in loading amplitude and stress state, which modify the local stress intensity factor at the crack tip. IPF analysis indicates that the two crack deflections do not exhibit a strong correlation with the crystallographic texture, suggesting that grain orientation has a limited influence on the crack growth direction. The KAM values at crack deflection I are similar to those on the crack growth path and do not show a significant increase. However, SF mapping reveals that the crack deflection occurs due to the encountering hard-oriented grains with low SF values, which impede crack growth, redirecting it toward adjacent grains with higher SF values that are more prone to slip [54]. At deflection II region, the crack may preferentially grow along the favorable slip direction (dashed line), but is markedly impeded upon encountering a HAGB. This obstruction leads to dislocation accumulation near the boundary and intensified intragranular plastic deformation, reflected by higher KAM values in the region. Subsequently, the crack growth is deflected along the path of less resistance and further expanded along the HAGB.

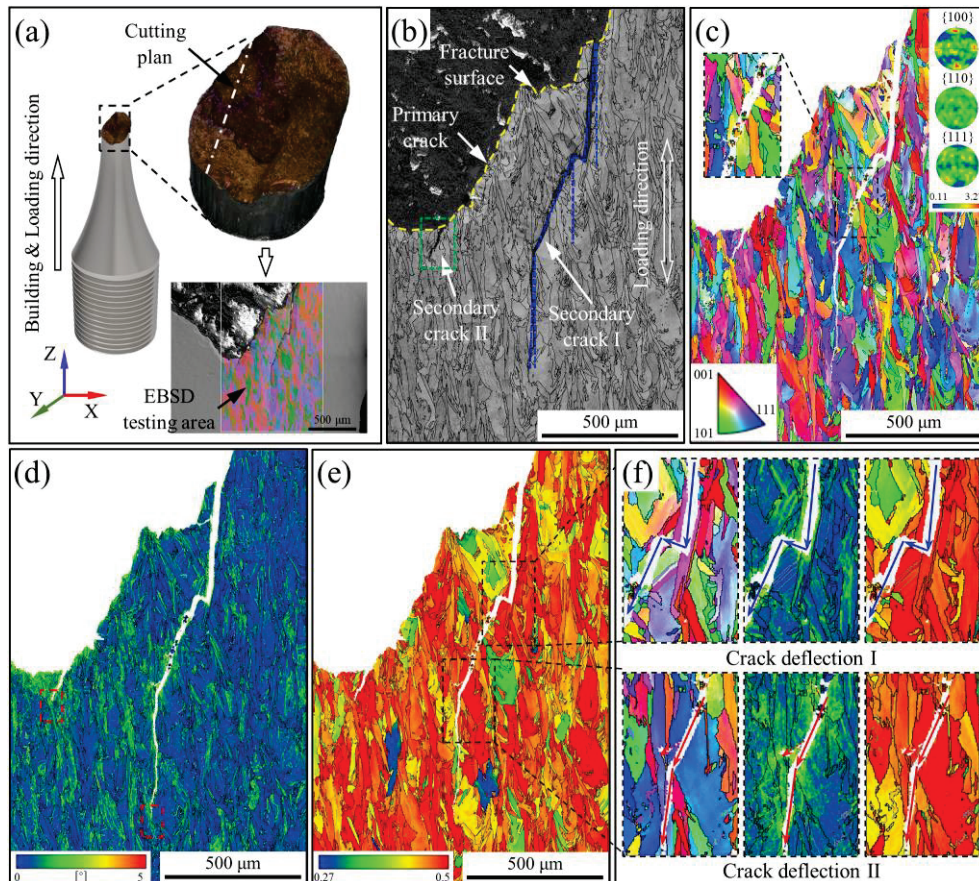


Fig. 16. EBSD analysis of local microtexture characteristics for cracks under VA loading in the VHCF regime at 650 °C ($\sigma_a=550\rightarrow575\rightarrow600\rightarrow625$ MPa, $N_f=3.338E7$): (a) schematic and test area of EBSD analysis, (b) primary and secondary crack morphology, (c) IPF, (d) KAM map, (e) Schmidt factor, and (f) crack deflection behavior.

3.7 Interior fatigue failure behavior

The preceding analysis reveals the fatigue failure mode and crack growth behavior under VA loading, and the interior fatigue failure process is summarized schematically in Fig. 17. The microstructure of solution-aged LPBF nickel-based superalloy primarily comprises δ and γ'/γ'' phases. The δ phase plays a beneficial role in suppressing grain boundary migration and hindering crack growth, while the γ'/γ'' phases enhance fatigue strength through precipitation strengthening and dislocation pinning mechanisms [55].

Before cyclic loading, XCT and SEM analyses reveal the inevitable presence of defects, including inhomogeneous microstructures, keyholes, LoF, and pores. Additionally, grains with different crystallographic orientations exhibit different slip system activation properties under applied loading, making them potential sources of crack nucleation in the VHCF regime. A limited number of dislocations introduced during the machining process are also observed, as shown in Fig. 17(a).

Under cyclic loading, the elastic-plastic incompatibility between the internal defects and surrounding matrix induces local stress concentration, promoting dislocation formation. These dislocations accumulate at the defect-matrix interface, gradually developing into slip lines and slip bands. As the slip bands evolve, they intensify local plastic deformation, ultimately leading to microcracks nucleation (Fig. 17(b)). The microcracks tend to grow along grains with high SF in a transgranular fracture mode under local shear force. Grain fracture results in the formation of facets and HAGBs with large orientation differences can either hinder microcrack growth or cause microcrack deflection. As the microcracks grows, fracture occurs in multiple adjacent grains, gradually forming a circular rough area centered around the defect, characterized by aggregated facets, which is called FCA in this study (Fig. 17(c)). This point marks the onset of Region II in Paris law, where the crack enters a stable propagation stage. Under VA loading, crack growth rate is significantly affected by the variation of the stress level. Theoretically, large stress level fluctuations can lead to concentric rings or bands of beach marks on the fracture surface due to the fluctuation of crack growth rate [56]. However, in the present study, no clear beach marks are observed on the

fracture surfaces. This absence may be attributed to the relatively small difference in stress levels between the applied loading blocks, which enables continuous fatigue crack growth across different stages without significant cyclic stagnation or restart. In addition, the stress ratio, loading sequence, and fatigue failure mode may contribute to the absence of typical fatigue striations. Under tension-tension cyclic loading with the stress ratio of $R=0.1$, no extrusion strips or stop marks are formed due to the lack of compressive stresses under the low-to-high loading sequence.

At elevated temperatures, facet aggregation is a typical feature of interior failure in LPBF nickel-based superalloys. The relatively uniform grain size leads to consistent facet dimensions resulting from grain fracture. This fracture mode can obscure the fatigue banding typically observed in fatigue fractures under VA loading, as continuous transgranular crack growth within similar-sized grains suppresses the formation of distinct cyclic fatigue bands. As cyclic loading continues, more facets are formed due to the progressive fracture of grains, resulting in the expansion of the FCA, which corresponds to Region II of the stable expansion phase of crack growth in the Paris law, as shown in Fig. 17(d). The stress intensity factor (SIF) is a critical parameter for characterizing the stress field at the crack front. During fatigue crack growth, the maximum SIF near the FCA approaches the threshold value required to initiate macrocrack growth [57]. Based on the projected area of the FCA and the applied stress range, the SIF for macrocrack at $R=0.1$ range from 16.17 to 21.74 $\text{MPa}\cdot\text{m}^{1/2}$, with an average of 19.59 $\text{MPa}\cdot\text{m}^{1/2}$. Once this threshold is exceeded, the interior crack transitions to a rapid and unstable growth. When the maximum SIF surpasses the fatigue fracture toughness of the alloy, the crack becomes fully destabilized, resulting in final fracture (Fig. 17(e)).

In summary, the defects induce fatigue crack nucleation under cyclic loading, followed by transgranular microcrack growth that produces an individual facet, and eventually the accumulation of these facets develop into the FCA. The defects-facets-FCA sequence constitutes the typical progression of interior fatigue failure in LPBF nickel-based superalloys at elevated temperature.

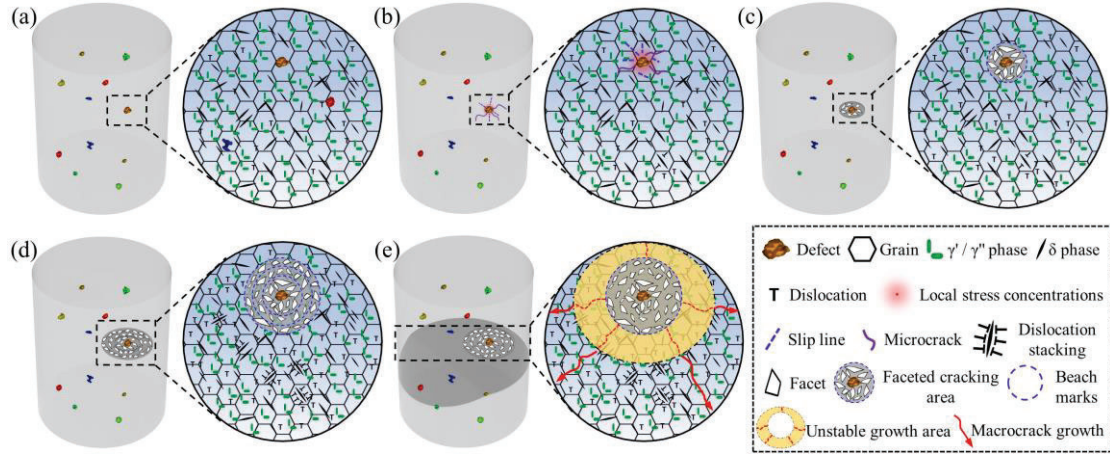


Fig. 17. Schematic of interior fatigue failure process: (a) grains with initial defects and orientation differences, (b) slip lines and microcracks, (c) facets and FCA, (d) formation of beach marks, and (e) final fracture with macrocrack.

3.8 Modeling of interior fatigue cumulative damage

3.8.1 Linear damage cumulative method

Miner's rule, also known as the linear cumulative damage hypothesis, is one of the most widely applied methods for evaluating fatigue life under VA loading. The rule is based on the assumption that fatigue damage accumulates linearly with the fraction of applied cycles at each stress level, and failure occurs when the total accumulated damage reaches unity. Mathematically, this can be expressed as:

$$\sum_{i=1}^k \frac{n_i}{N_i} = 1 \quad (1)$$

where n_i is the number of applied cycles at stress amplitude σ_i , and N_i is the corresponding fatigue life under CA loading at the same stress level.

For example, in a two-level loading sequence, if the material is subjected first to n_1 cycles at stress amplitude σ_1 (with corresponding fatigue life N_1), followed by n_2 cycles at stress amplitude σ_2 (with corresponding fatigue life N_2), the cumulative damage can be expressed as:

$$D = \frac{n_1}{N_1} + \frac{n_2}{N_2} \quad (2)$$

Failure is predicted to occur when $D=1$. More generally, for a multi-level loading history involving stress amplitudes $\sigma_1, \sigma_2, \dots, \sigma_i$, the rule extends to:

$$D = \frac{n_1}{N_1} + \frac{n_2}{N_2} + \dots + \frac{n_i}{N_i} = 1 \quad (3)$$

Miner's rule offers a simple and practical framework for engineering applications, it is intrinsically limited because it neglects load sequence effects and microstructural influences on fatigue damage evolution.

3.8.2 Definition of remaining life factor

The fatigue damage (D) accumulated in a material is influenced by multiple factors, including a material constant (p), the stress level (σ), the number of cycle loading (n), and the cyclic increment (dn) [58]. The damage increment (dD) is given by:

$$dD = f(D, \sigma, n, dn; p) \quad (4)$$

Furthermore, the relationship between the relative increase in fatigue damage and the relative increase in cycles can be expressed in normalized linear form as [59-61]:

$$\frac{dD}{D} = f(\sigma; p) \frac{dn}{n} \quad (5)$$

where $f(\sigma; p)$ is a function related to the fatigue process and material properties, reflecting the damage evolution characteristics of the material under specific loading conditions.

In engineering applications, the remaining fatigue life is often more relevant than cumulative damage itself. To better quantify this, a remaining life factor (RLF, α) is introduced as:

$$\alpha = \frac{N_R}{N_f} \quad (6)$$

where N_R denotes the remaining life after n cycles under the stress level σ . By incorporating this definition, Eq. (5) can be reformulated as:

$$D = 1 - \alpha^{f(\sigma, p)} \quad (7)$$

This formulation provides a convenient framework for describing the progression of fatigue damage using remaining life as a more directly applicable engineering metric.

Based on the conventional equivalent damage theory, if the equivalent damage curves, and the full cyclic loading history are known, then the remaining fatigue life at the final stress level can be predicted. For example, under a two-level loading condition, the damage evolution under each stress level can be plotted as equivalent damage curves, with stress level on the vertical axis and the number of cycles on the horizontal axis, as schematically shown in Fig. 18.

Since traditional fatigue limits do not exist for many materials, all equivalent damage curves in Fig. 18 are assumed to extend through the static yield point rather than the conventional fatigue endurance limit.

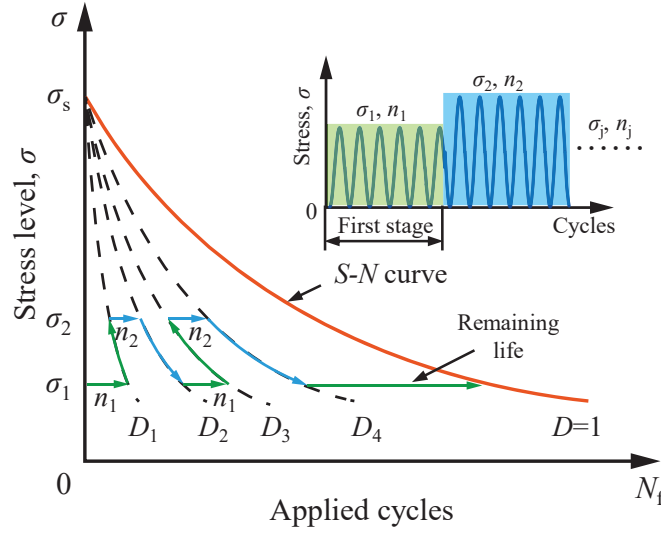


Fig. 18. Schematic illustration of the equivalent damage curve showing the relationship between cumulative damage and load cycles.

Several variables are defined in this study before proceeding with the discussion. Where σ_j represent the j -th stress level, n_j is the number of cycles applied at σ_j stress level. The n_{ij} denotes equivalent number of cycles at σ_j stress level with the fatigue life N_{fj} , which can cause the same damage as the number of cycles applied in the prior i -th stage. And α_{ij} is the RLF after i -th stage at σ_j stress level. The evolution of the RLF during cyclic loading is discussed as a case of two-level loading in Fig. 19. If n_1 cycles are applied at the first stress level σ_1 with the fatigue life N_{f1} , the corresponding α_{11} after the first loading level is given by:

$$\alpha_{11} = (N_{f1} - n_1) / N_{f1} \quad (8)$$

Next, the damage transition at the subsequent stress level evolves along a nonlinear damage curve, regardless of whether the loading sequence is low-to-high or high-to-low, i.e., the equivalent damage point migrates along the curve to (σ_2, n_{12}) , corresponding to the equivalent remaining life factor α_{12} can be expressed as:

$$\alpha_{12} = 1 - \frac{n_{12}}{N_{f2}} \quad (9)$$

where n_{12} represents the equivalent number of cycles at stress level σ_2 with a corresponding fatigue life N_{f2} , which equates to the damage accumulated during the first stage. Combining Eqs. (7), (8) and (9), this can be rewritten as:

$$\alpha_{12} = \alpha_{11}^{\frac{f(\sigma_1;p)}{f(\sigma_2;p)}} = \left(1 - \frac{n_1}{N_{f1}}\right)^{\frac{f(\sigma_1;p)}{f(\sigma_2;p)}} \quad (10)$$

where $f(\sigma_1; p)$ and $f(\sigma_2; p)$ are correlation functions corresponding to the stress levels σ_1 and σ_2 . Based on this relationship, the remaining life N_{R12} , representing the probable number of cycles at stress level σ_2 after the first loading stage, can be expressed as:

$$N_{R12} = \alpha_{12} N_{f2} \quad (11)$$

If n_{12} cycles are subsequently applied at σ_2 , and $n_2 < N_{R12}$, the corresponding RLF, α_{22} , can be expressed as:

$$\alpha_{22} = \alpha_{12} - \frac{n_2}{N_{f2}} \quad (12)$$

For multi-level loading, the generalized form of the remaining life factor after j stages α_{ij} can be expressed as:

$$\alpha_{ij} = \left\{ \left[\left(1 - \frac{n_1}{N_{f1}} \right)^{\frac{f(\sigma_1; p)}{f(\sigma_i; p)}} - \frac{n_2}{N_{f2}} \right]^{\frac{f(\sigma_2; p)}{f(\sigma_j; p)}} - \dots - \frac{n_{j-1}}{N_{f, j-1}} \right\}^{\frac{f(\sigma_{j-1}; p)}{f(\sigma_j; p)}} \quad (13)$$

And the corresponding remaining life at stage j N_{Rij} is expressed as:

$$N_{Rij} = \alpha_{ij} N_{fi} \quad (14)$$

It should be noted that the sequence of fatigue cyclic loading significantly affects damage accumulation under VA loading conditions. Specifically, the remaining life differs significantly between low-to-high and high-to-low loading sequences. This variation arises from the influence of different stress levels on microscopic damage evolution, especially in terms of fatigue crack nucleation and growth behavior.

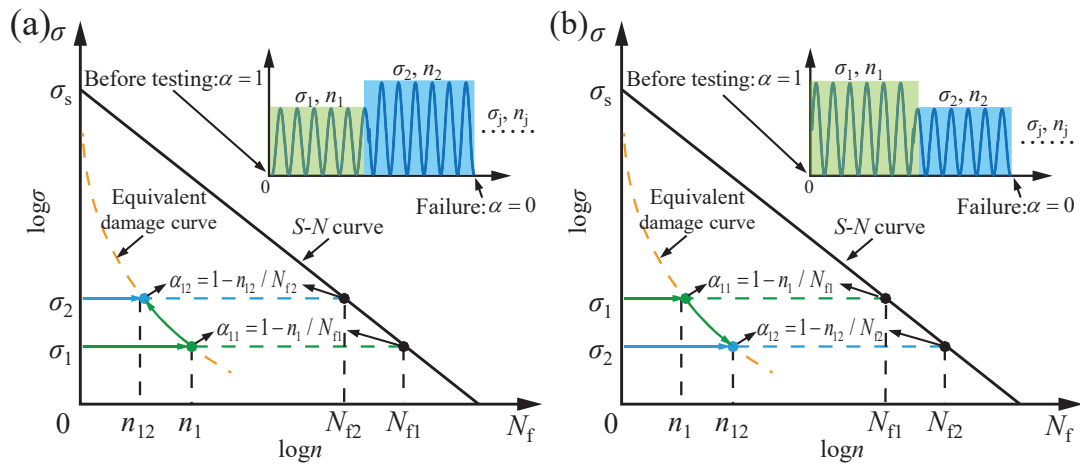


Fig. 19. Definition of the remaining life factor under two-level loading: (a) loading from low to high,

and (b) loading from high to low.

3.8.3 Determination of correlation function

A key challenge in remaining life assessment is to determine the correlation function $f(\sigma; p)$ applicable to different loading levels. To ensure meaningful damage representation, the equivalent damage curves need to satisfy two essential criteria: the curves must not intersect each other, and each curve must intersect the coordinate axes at a unique point, i.e., $\sigma=\sigma_s$.

If the effect of loading sequence is not considered, the equivalent damage curve can be described by a double logarithmic function:

$$\log \sigma = A \log n + B \quad (15)$$

where A is the intercept (coefficient), and B is the slope (exponent) of the log-log damage curve.

Since each damage curve passes through the point $(n=1, \sigma=\sigma_s)$, the exponent B can be determined as:

$$B = \log \sigma_s \quad (16)$$

Substituting Eq. 16 into Eq. 15 gives:

$$\log \sigma = A \log n + \log \sigma_s \quad (17)$$

For two-level loading, the corresponding equivalent damage points are (σ_1, n_1) and (σ_2, n_2) .

Applying log-log relation:

$$\log \sigma_1 = A \log n_1 + \log \sigma_s \quad (18)$$

$$\log \sigma_2 = A \log n_2 + \log \sigma_s \quad (19)$$

According to Eq. (10), replacing α with n gives:

$$\alpha_{12} = \alpha_{11}^{\frac{\log \sigma_s - \log \sigma_2}{\log \sigma_s - \log \sigma_1}} \quad (20)$$

To account for loading sequence effects, then, a load interaction factor is introduced, defined as [59]:

$$\mu_{j+1} = \left(\frac{\sigma_j}{\sigma_{j+1}} \right)^2 \quad (21)$$

where μ_{j+1} represents the load interaction factor between the j -th stress level and the $j+1$ -th stress level. This interaction factor is based on empirical parameters obtained from repetitive loading tests [62]. The value of μ_{j+1} is typically greater than 1 under high-to-low loading conditions and smaller than 1 under low-to-high loading conditions. Therefore, based on Eq. (21), Eq. (20) can be

introduced as:

$$\alpha_{12} = \frac{\log \sigma_s - \log \sigma_2}{\log \sigma_s - \log \sigma_1} \mu_2 = \alpha_{11} \frac{\log \sigma_s - \log \sigma_2}{\log \sigma_s - \log \sigma_1} \left(\frac{\sigma_1}{\sigma_2} \right)^2 \quad (22)$$

Accordingly, the ratio of correlation functions is given by:

$$\frac{f(\sigma_1; p)}{f(\sigma_2; p)} = \frac{\log \sigma_s - \log \sigma_2}{\log \sigma_s - \log \sigma_1} \left(\frac{\sigma_1}{\sigma_2} \right)^2 \quad (23)$$

Thus, for j -th level loading, the ratio of the correlation functions is:

$$\frac{f(\sigma_j; p)}{f(\sigma_{j+1}; p)} = \frac{\log \sigma_s - \log \sigma_{j+1}}{\log \sigma_s - \log \sigma_j} \left(\frac{\sigma_j}{\sigma_{j+1}} \right)^2 \quad (24)$$

This allows reconstruction of the correlation functions necessary for predicting fatigue life across VA sequences.

3.8.4 Reconstruction of stress-life curve

In the VHCF regime, the dispersion of fatigue life increases as the stress level decreases. Therefore, the fitting accuracy of the S - N curve is critical for accurate damage assessment. An S - N curve considering defect size and FCA is developed to replace the original S - N curve to improve the accuracy of fatigue life prediction. According to previous studies [63, 64], the LPBF defects facilitate crack nucleation under both CA and VA loading conditions, while the formation of FCA is instrumental in the stable crack growth process. The number of cycles consumed in forming the FCA accounts for almost the total fatigue life.

To quantify this relationship, a parameter δ is assumed to represent the ratio of FCA size to defect size:

$$\delta = \frac{\sqrt{area_{FCA}}}{\sqrt{area_{def}}} \quad (25)$$

Based on prior study [65], the fatigue strength of LPBF alloys with interior failure mechanisms can be modeled as:

$$\sigma_a = \exp[a + b \ln(N_f \delta^{-c})] \quad (26)$$

where a , b and c are material constants that can be determined by via nonlinear regression from experimental data.

The resulting reconstructed S - N curve, shown in Fig. 20, demonstrates good predictive accuracy, even in the VHCF domain. This supports its applicability for modeling damage accumulation in LPBF nickel-based superalloys.

According to Eq. (26), the predicted fatigue life N_{fj} at stress level σ_i can be expressed as:

$$N_{fj} = \exp\left(\frac{\ln \sigma_j - a}{b} + c \ln \delta\right) \quad (27)$$

This relationship enables more accurate integration of microstructural features and fracture behavior into fatigue life predictions under realistic service loading conditions. To comprehensively and reliably assess the fatigue properties of the materials, additional experiments should be conducted to obtain a larger dataset. This would improve the characterization of fatigue life distribution and variability, and provide a stronger experimental basis for developing fatigue life prediction models, thereby enhancing the credibility and generalizability of the findings.

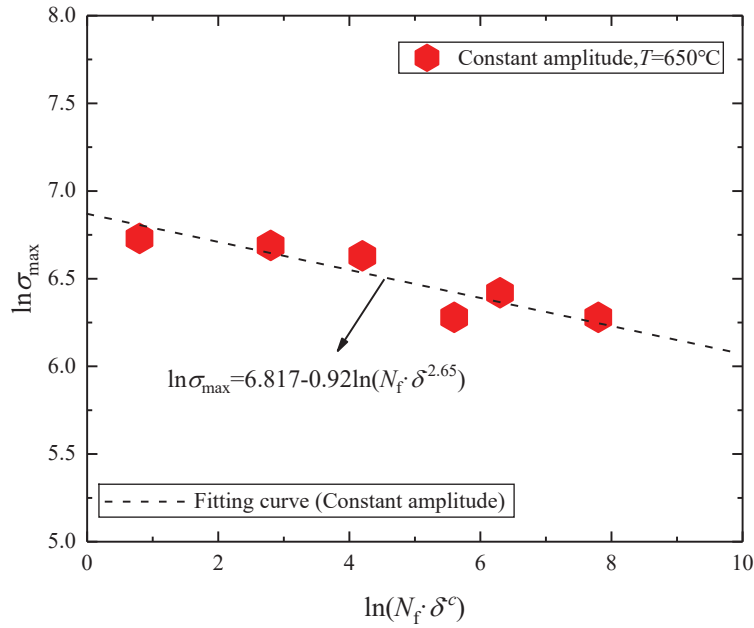


Fig. 20. Reconstructed S - N curve of LPBF nickel-based superalloy with defect-FCA induced failure.

3.8.5 Establishment of fatigue life model

By combining Eqs. (14), (24) and (27), the remaining fatigue life in relation to defect-FCA induced failure during fatigue damage can be expressed as:

$$N_{Rij} = N_{fj} \left\{ \left[\left(1 - \frac{n_1}{N_{f1}} \right)^{\frac{f(\sigma_1;p)}{f(\sigma_i;p)}} - \frac{n_2}{N_{f2}} \right]^{\frac{f(\sigma_2;p)}{f(\sigma_3;p)}} \cdots \frac{n_{j-1}}{N_{f,j-1}} \right\}^{\frac{f(\sigma_{j-1};p)}{f(\sigma_j;p)}} \quad (28)$$

Hence, incorporating the previous applied number of cycles, the total fatigue life model is expressed as:

$$N_{\text{pre}} = \sum n_j + N_{\text{fj}} \left\{ \left[\left(1 - \frac{n_1}{N_{\text{fi}}} \right)^{\frac{f(\sigma_1; p)}{f(\sigma_1; p)}} - \frac{n_2}{N_{\text{f2}}} \right]^{\frac{f(\sigma_2; p)}{f(\sigma_3; p)}} - \dots - \frac{n_{j-1}}{N_{\text{fj-1}}} \right\}^{\frac{f(\sigma_{j-1}; p)}{f(\sigma_j; p)}} \quad (29)$$

3.9 Perdition of fatigue life

According to Eqs. (3) and (29), the comparison between the predicted and the experimental results of the proposed method and Miner's rule is presented in Fig. 21. It can be noted that the cumulative life prediction method proposed in this study outperforms the traditional Miner's linear cumulative damage model in terms of overall accuracy. The predictions of Miner's rule are also accurate at high stress levels. This phenomenon can be explained by the difference in the predominance of crack nucleation and crack growth: the fatigue life of the long-life range is mainly consumed in the crack nucleation stage, whereas the short-life range is dominated by crack growth [35, 66]. In the "low–high" loading sequence, the damage associated with crack growth accumulates during the low-stress level. Once the stress level increases to the high-stress range, crack growth proceeds within only a few cycles and rapidly enters the growth stage, resulting in fatigue failure being primarily governed by the high-stress amplitude block. Compared with Miner's rule, the proposed method can reflect the nonlinear cumulative characteristics of fatigue damage more effectively, and its prediction results show better consistency with the experimental data. This indicates that the microstructure fatigue damage assessment method based on the defect-FCA induced failure proposed in this paper is effective for predicting the fatigue life of LPBF nickel-based superalloys under VA loading in both HCF and VHCF regimes. It is undeniable that Miner's rule still has its unique advantages in fatigue life prediction, such as the simplicity of the calculation process, wide applicability, and high economy, which makes it still widely adopted in engineering applications.

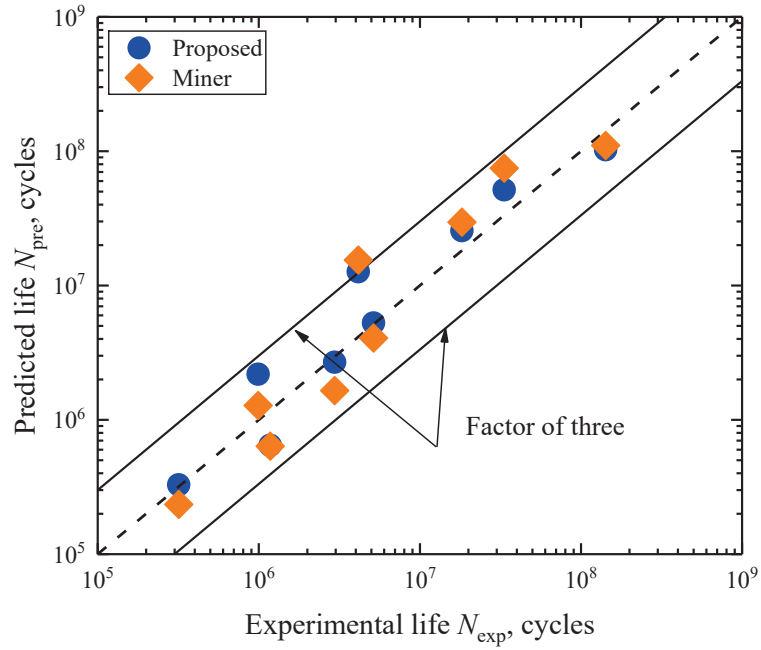


Fig. 21. Comparison between predicted and experimental results of the proposed method and Miner's rule.

To aid implementation, the overall fatigue damage evolution process is summarized in the flowchart shown in Fig. 22. The main processes are as follow: (1) Citation of equivalent damage curve theory, (2) Definition of remaining life factor, (3) Establishment of nonlinear damage evolution expression, (4) Evaluation of equivalent damage for two-level loading, (5) Extension to multi-level loading condition, (6) Prediction of remaining life under multi-level loading, (7) Determining the correlation function, (8) Recursive evaluation of multi-stage correlation function, (9) Size evaluation of defects and FCA, (10) S - N curve reconstruction incorporating defect-FCA interaction, (11) Formulation of a microstructure-sensitive fatigue life prediction model, (12) Validation via comparison of predicted and experimental fatigue lives.

Overall, the proposed framework provides a robust and versatile approach for modeling fatigue behavior in LPBF nickel-based superalloy. It provides a reliable basis for fatigue life prediction and reliability assessment under service-relevant uniaxial variable amplitude loading conditions. Nevertheless, the cumulative damage life prediction framework still presents opportunities for further refinement. At present, its construction primarily relies on experimental data obtained under specific loading conditions, limiting its validation across more complex loading spectra and diverse service environments. In addition, critical influencing factors—such as residual stress, thermal gradients, and post-processing treatments—have yet to be explicitly integrated into the modeling framework, despite their potential to markedly impact the fatigue performance of materials under

real-world operational scenarios. Moreover, it is necessary to conduct more experiments to accumulate data to further validate and refine the proposed framework. To enhance the model's predictive accuracy and generalizability across multiple scales and service conditions, future work may benefit from incorporating physically-based damage evolution mechanisms in conjunction with advanced statistical modeling and machine learning techniques.

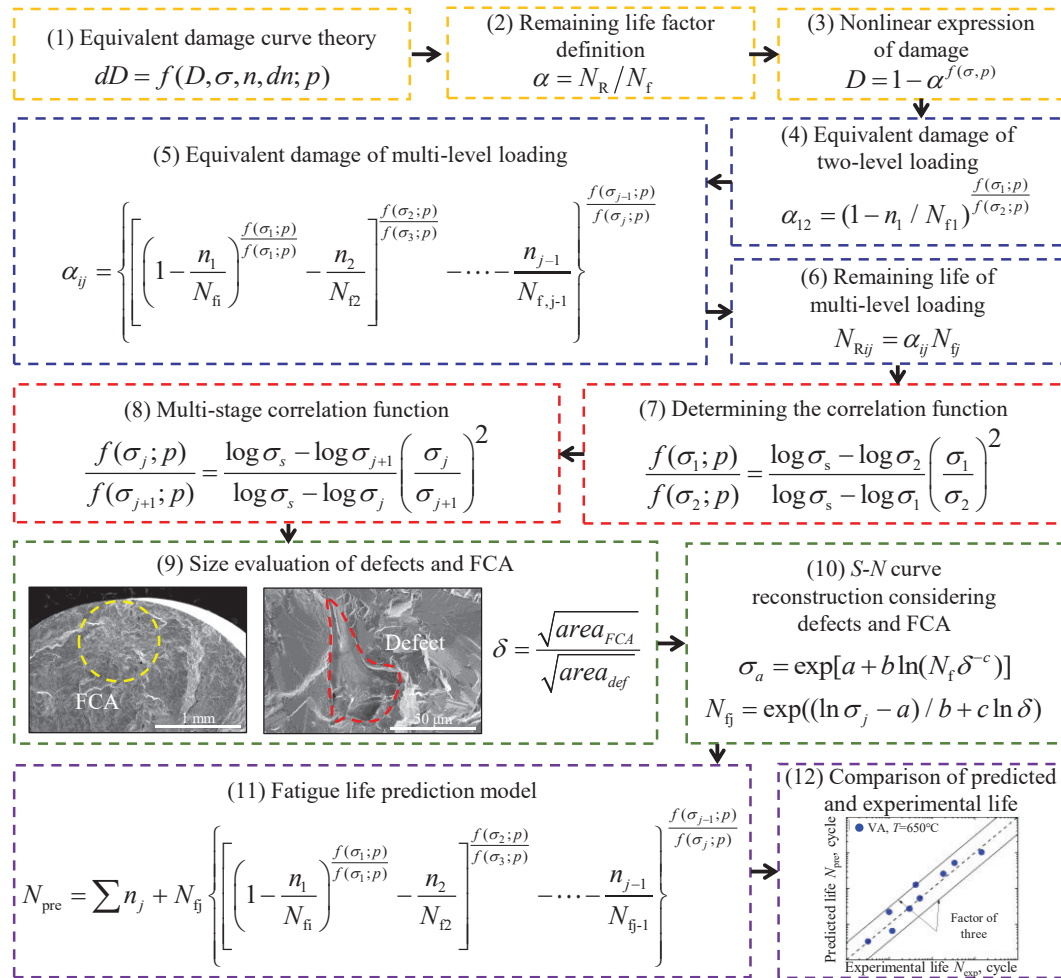


Fig. 22. The flowchart of fatigue damage evaluation for defect-FCA induced failure.

4 Conclusions

In this study, VA loading fatigue tests employing a step-loading approach with increasing stress amplitude were conducted on a LPBF nickel-based superalloy at 650 °C. The mechanisms of defect-induced interior failure and crack growth behavior were investigated. Based on these insights, a novel cumulative damage evaluation model accounting defect-FCA interaction was proposed. The main conclusions can be summarized as follows:

- (1) Interior fatigue failure predominates in LPBF nickel-based superalloy at 650°C under VA

loading. Crack nucleation sources primarily include inhomogeneous microstructures, keyholes, LoF, pores, and crystallographic facets.

(2) VA loading, characterized by stress amplitude fluctuations, can lead to variations in local stress distribution and plastic deformation. This promotes multi-source crack nucleation, including from multiple interior sites or from a combination of interior and surface sites.

(3) The primary and secondary cracks are characterized by transgranular fracture accompanied by grain fragmentation, and their crack growth paths are not obviously related to the crystallographic texture. The crack deflection behavior is affected by load variations, activation properties of the slip system, and grain boundary interactions.

(4) The interior fatigue failure mechanism of LPBF nickel-based superalloy under VA loading with increasing stress amplitude is elucidated. The fatigue response is shown to be governed by the interplay among the stress ratio, amplitude variations, loading sequence and the underlying grain microstructure.

(5) A novel cumulative damage evaluation model is established by incorporating the remaining life factor, correlation function transformation, and the reconstruction of the stress-life relationship accounting for defect-FCA. This approach incorporates microstructural features and load interaction effects. The prediction results are in close accord with the experimental data.

CRedit authorship contribution statement

Chuanwen Sun: Writing-Original Draft, Conceptualization, Methodology, Investigation, Data curation. **Wei Li:** Writing-Review & Editing, Methodology, Resources, Funding acquisition. **Ahmad Serjouei:** Writing-Review & Editing, Visualization, Investigation. **Cheng Li:** Formal analysis, Validation. **Rui Sun:** Supervision, Visualization. **Ibrahim Elbugdady:** Investigation. **Yuzhe Jin:** Formal analysis.

Declaration of Competing Interest

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

Data availability

The raw data required to reproduce these findings is available upon a reasonable request. Please directly contact the corresponding author of this paper.

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References

- [1] Chandra S, Radhakrishnan J, Huang S, Wei SY, Ramamurty U. Solidification in metal additive manufacturing: challenges, solutions, and opportunities. *Prog Mater Sci* 2024, 148: 101361. <https://doi.org/10.1016/j.pmatsci.2024.101361>.
- [2] Wu YZ, Fang JG, Wu C, Li CY, Sun GY, Li Q. Additively manufactured materials and structures: A state-of-the-art review on their mechanical characteristics and energy absorption. *Int J Mech Sci* 2023, 246: 108102. <https://doi.org/10.1016/j.ijmecsci.2023.108102>.
- [3] Khan N, Riccio A. A systematic review of design for additive manufacturing of aerospace lattice structures: Current trends and future directions. *Prog Aerosp Sci* 2024, 149: 101021. <https://doi.org/10.1016/j.paerosci.2024.101021>.
- [4] Pixner F, Warchomicka F, Lipińska M, Elmiger S, Jechtl C, Auer P, et al. Thermal cycling effects on the local microstructure and mechanical properties in wire-based directed energy deposition of nickel-based superalloy. *Addit Manuf* 2024, 83: 104066. <https://doi.org/10.1016/j.addma.2024.104066>.
- [5] Cai YH, Xiong J, Chen H, Zhang GJ. A review of in-situ monitoring and process control system in metal-based laser additive manufacturing. *J Manuf Syst* 2023, 70: 309-326. <https://doi.org/10.1016/j.jmsy.2023.07.018>.
- [6] Liu H, Yu HY, Guo C, Chen XL, Zhong SY, Zhou L, et al. Review on fatigue of additive manufactured metallic alloys: microstructure, performance, enhancement, and assessment methods. *Adv Mater* 2024, 36(17): 2306570. <https://doi.org/10.1002/adma.202306570>.
- [7] Salvati E. Evaluating fatigue onset in metallic materials: Problem, current focus and future perspectives. *Int J Fatigue* 2024, 188: 108487. <https://doi.org/10.1016/j.ijfatigue.2024.108487>.
- [8] Gudivada G, Pandey A K. Recent developments in nickel-based superalloys for gas turbine applications. *J Alloy Compd* 2023, 963: 171128. <https://doi.org/10.1016/j.jallcom.2023.171128>.
- [9] Nabavi SF, Farshidianfar A, Dalir H. Comprehensive review: Advancements in modeling geometrical and mechanical characteristics of laser powder bed fusion process. *Opt Laser Technol* 2025, 180: 111480. <https://doi.org/10.1016/j.optlastec.2024.111480>.
- [10] Gordon JV, Narra SP, Cunningham RW, Liu H, Chen HM, Suter RM, et al. Defect structure process maps for laser powder bed fusion additive manufacturing. *Addit Manuf* 2020, 36: 101552. <https://doi.org/10.1016/j.addma.2020.101552>.
- [11] Sun CW, Li W, Liu G, Sun Rui, Sun ZD. Multiple defect-driven high temperature fatigue internal cracking mechanisms of nickel-based superalloy fabricated by laser powder bed fusion at same stress level. *Mater Today Commun* 2024, 41: 111086. <https://doi.org/10.1016/j.mtcomm.2024.111086>.
- [12] Sanaei N, Fatemi A. Defects in additive manufactured metals and their effect on fatigue performance: A state-of-the-art review. *Prog Mater Sci* 2021, 117: 100724. <https://doi.org/10.1016/j.pmatsci.2020.100724>.
- [13] Sun CW, Li W, Sun R, Li XL, Liu G, Wang SW, et al. Multi-scale study of subsurface fatigue cracking behavior of laser-powder bed fused Inconel 718 at stress ratios and temperatures. *Mater Sci Eng A* 2024, 908: 146805. <https://doi.org/10.1016/j.msea.2024.146805>.
- [14] Yang K, Huang Q, Wang QY, Chen Q. Competing crack initiation behaviors of a laser additively manufactured nickel-based superalloy in high and very high cycle fatigue regimes. *Int J Fatigue* 2020, 136: 105580. <https://doi.org/10.1016/j.ijfatigue.2020.105580>.

- [15] Macoretta G, Romanelli L, Santus C, Romoli L, Lutey A, Uriati F, et al. Modelling of the surface morphology and size effects on fatigue strength of L-PBF Inconel 718 by comparing different testing specimens. *Int J Fatigue* 2024, 181: 108120. <https://doi.org/10.1016/j.ijfatigue.2023.108120>.
- [16] Shi T, Li J, Gao G, Sun JY, Yang ZG, Yan JY, et al. Microstructural evolution and formation of fine grains during fatigue crack initiation process of laser powder bed fusion Ni-based superalloy. *Addit Manuf* 2024, 85: 104151. <https://doi.org/10.1016/j.addma.2024.104151>.
- [17] Song X, Feih S, Zhai W, Sun CN, Li F, Maiti R, et al. Advances in additive manufacturing process simulation: Residual stresses and distortion predictions in complex metallic components. *Mater Des* 2020, 193: 108779. <https://doi.org/10.1016/j.matdes.2020.108779>.
- [18] Macoretta G, Bertini L, Monelli BD, Berto F. Productivity-oriented SLM process parameters effect on the fatigue strength of Inconel 718. *Int J Fatigue* 2023, 168: 107384. <https://doi.org/10.1016/j.ijfatigue.2022.107384>.
- [19] Niu XP, He C, Zhu SP, Foti P, Berto F, Wang LY, et al. Defect sensitivity and fatigue design: Deterministic and probabilistic aspects in AM metallic materials. *Prog Mater Sci* 2024, 144: 101290. <https://doi.org/10.1016/j.pmatsci.2024.101290>.
- [20] Laleh M, Sadeghi E, Revilla RI, Chao Q, Haghdadi N, Hughes AE, et al. Heat treatment for metal additive manufacturing. *Prog Mater Sci* 2023, 133: 101051. <https://doi.org/10.1016/j.pmatsci.2022.101051>.
- [21] Liu SY, Shao S, Guo H, Zong R, Qin CX, Fang XY. The microstructure and fatigue performance of Inconel 718 produced by laser-based powder bed fusion and post heat treatment. *Int J Fatigue* 2022, 156: 106700. <https://doi.org/10.1016/j.ijfatigue.2021.106700>.
- [22] Sun CW, Li W, Li C, Sun R, Liu G, Li XL. Study on microstructure and fatigue properties of laser powder bed fusion Nickel-based superalloy with heat treatment. *Addit Manuf Front* 2025: 200217. <https://doi.org/10.1016/j.amf.2025.200217>.
- [23] Yu CL, Huang ZY, Zhang Z, Shen JB, Wang J, Xu ZP. Influence of post-processing on very high cycle fatigue resistance of Inconel 718 obtained with laser powder bed fusion. *Int J Fatigue* 2021, 153: 106510. <https://doi.org/10.1016/j.ijfatigue.2021.106510>.
- [24] Nezhadfar PD, Johnson AS, Shamsaei N. Fatigue behavior and microstructural evolution of additively manufactured Inconel 718 under cyclic loading at elevated temperature. *Int J Fatigue* 2020, 136: 105598. <https://doi.org/10.1016/j.ijfatigue.2020.105598>.
- [25] Li WB, Pang JC, Zhang H, Li SX, Zhang ZF. The high-cycle fatigue properties of selective laser melted Inconel 718 at room and elevated temperatures. *Mater Sci Eng A* 2022, 836: 142716. <https://doi.org/10.1016/j.msea.2022.142716>.
- [26] Li W, Sun R, Hu TY, Li XL, Li C, Zhang YC, et al. Effect of elevated temperature on high-cycle and very-high-cycle fatigue properties of Ni-based superalloy manufactured by selective laser melting. *Int J Fatigue* 2021, 148: 106250. <https://doi.org/10.1016/j.ijfatigue.2021.106250>.
- [27] Sun CW, Li W, Liu G, Sun R, Wang CP, Li C, et al. Evolution behavior of competing fatigue failure and life prediction related to defect, stress and temperature for laser powder bed fused superalloy with solution aging treatment. *Int J Fatigue* 2025, 198: 109029. <https://doi.org/10.1016/j.ijfatigue.2025.109029>.
- [28] Wang ML, Yang XG, Li B, Shi DQ, Guo SQ, Fan YS. The dominant role of defects on fatigue behaviour of a SLM Ni-based superalloy at elevated temperature. *Int J Fatigue* 2023, 176: 107894. <https://doi.org/10.1016/j.ijfatigue.2023.107894>.

- [29] Radhakrishnan J, Li S H, Ramamurty U. Unnotched fatigue behavior of the laser powder bed fused and hot isostatic pressed Inconel 718 at 600°C. *Mater Sci Eng A* 2025, 920: 147527. <https://doi.org/10.1016/j.msea.2024.147527>.
- [30] Gruber K, Szymczyk-Ziółkowska P, Dziuba S, Duda S, Zielonka P, Seitzl S, et al. Fatigue crack growth characterization of Inconel 718 after additive manufacturing by laser powder bed fusion and heat treatment. *Int J Fatigue* 2023, 166: 107287. <https://doi.org/10.1016/j.ijfatigue.2022.107287>.
- [31] Foti P, Razavi N, Fatemi A, Berto F. Multiaxial fatigue of additively manufactured metallic components: A review of the failure mechanisms and fatigue life prediction methodologies. *Prog Mater Sci* 2023, 137: 101126. <https://doi.org/10.1016/j.pmatsci.2023.101126>.
- [32] Murakami Y, Takagi T, Wada K, Matsunaga H. Essential structure of SN curve: Prediction of fatigue life and fatigue limit of defective materials and nature of scatter. *Int J Fatigue* 2021, 146: 106138. <https://doi.org/10.1016/j.ijfatigue.2020.106138>.
- [33] Chen H, Yang F, Wu Z, Yang BW, Hou JZ. A nonlinear fatigue damage accumulation model under variable amplitude loading considering the loading sequence effect. *Int J Fatigue* 2023, 177: 107945. <https://doi.org/10.1016/j.ijfatigue.2023.107945>.
- [34] Campagnolo A, Vecchiato L, Meneghetti G. Multiaxial variable amplitude fatigue strength assessment of steel welded joints using the peak stress method. *Int J Fatigue* 2022, 163: 107089. <https://doi.org/10.1016/j.ijfatigue.2022.107089>.
- [35] Bjørheim F, Pavlou DG, Siriwardane SC. Nonlinear fatigue life prediction model based on the theory of the *S-N* fatigue damage envelope. *Fatigue Fract Eng M* 2022, 45(5): 1480-1493. <https://doi.org/10.1111/ffe.13680>.
- [36] Shen ZH, Huang ZY, Wang J, Qian HJ, Zheng LQ, Zhou QK, et al. Fracture mechanism and life prediction of IN718 under very high cycle variable amplitude loading at high temperature. *Eng Fail Anal* 2025: 109745. <https://doi.org/10.1016/j.engfailanal.2025.109745>.
- [37] Santos B, Infante V, Barros T, Baptista R. Study of fatigue crack propagation on modified CT specimens under variable amplitude loadings using machine learning. *Int J Fatigue* 2024, 184: 108332. <https://doi.org/10.1016/j.ijfatigue.2024.108332>.
- [38] Javidrad H, Koc B, Bayraktar H, Simsek U, Gunaydin K. Fatigue performance of metal additive manufacturing: A comprehensive overview. *Virtual Phys Prototyp* 2024, 19(1): e2302556. <https://doi.org/10.1080/17452759.2024.2302556>.
- [39] Yi M, Tang W, Zhu Y, Liang CG, Tang ZM, Yin Y, et al. A holistic review on fatigue properties of additively manufactured metals. *J Mater Process Tech* 2024, 329: 118425. <https://doi.org/10.1016/j.jmatprotec.2024.118425>.
- [40] ASTM E8/E8M-22, Standard Test Methods for Tension Testing of Metallic Materials, in Annual Book of ASTM Standards, Vol. 3.01, DOI: 10.1520/E0008_E0008M-22.
- [41] ASTM E21-20, Standard Test Methods for Elevated Temperature Tension Tests of Metallic Materials, in Annual Book of ASTM Standards, Vol. 3.01, DOI: 10.1520/E0021-20.
- [42] ASTM E466-21, Standard Practice for Conducting Force Controlled Constant Amplitude Axial Fatigue Tests of Metallic Materials, in Annual Book of ASTM Standards, Vol. 3.01, DOI: 10.1520/E0466-21.
- [43] Dang XF, Li Y, Chen K, Ramamurty U, Luo SH, Liang XQ, et al. Avoiding cracks in additively manufactured non-weldable directionally solidified Ni-based superalloys. *Addit Manuf* 2022,

- 59: 103095. <https://doi.org/10.1016/j.addma.2022.103095>.
- [44] Ma T, Zhang GP, Tan P, Zhang B. Effects of homogenization temperature on creep performance of laser powder bed fusion-fabricated Inconel 718 at 650°C. *Mater Sci Eng A* 2022, 853: 143794. <https://doi.org/10.1016/j.msea.2022.143794>.
- [45] Tajyar A, Brooks N, Holtham N, Rowe R, Newell DJ, Palazotto AN, et al. Effects of a modified heat-treatment on microstructure and mechanical properties of additively manufactured Inconel 718. *Mater Sci Eng A* 2022, 838: 142770. <https://doi.org/10.1016/j.msea.2022.142770>.
- [46] Zheng HQ, Zuo XW, Wan JF, Rong YH, Chen NL. Intrinsic mechanism of grain size effect and grain boundary misorientation angle effect on crack propagation in martensitic steels. *Eng Fail Anal* 2024, 163: 108497. <https://doi.org/10.1016/j.engfailanal.2024.108497>.
- [47] Stinville JC, Martin E, Karadge M, Ismonov S, Soare M, Hanlon T, et al. Fatigue deformation in a polycrystalline nickel base superalloy at intermediate and high temperature: Competing failure modes. *Acta Mater* 2018, 152: 16-33. <https://doi.org/10.1016/j.actamat.2018.03.035>.
- [48] Sakai T, Sato Y, Nagano Y, Takeda M, Oguma N. Effect of stress ratio on long life fatigue behavior of high carbon chromium bearing steel under axial loading. *Int J Fatigue* 2006, 28(11): 1547-1554. <https://doi.org/10.1016/j.ijfatigue.2005.04.018>.
- [49] Guo LP, Liu HJ, Wang HZ, Wei QL, Xiao YK, Tang ZJ, et al. Identifying the keyhole stability and pore formation mechanisms in laser powder bed fusion additive manufacturing. *J Mater Process Tech* 2023, 321: 118153. <https://doi.org/10.1016/j.jmatprotec.2023.118153>.
- [50] Mojumder S, Gan ZT, Li YF, Amin AA, Liu WK. Linking process parameters with lack-of-fusion porosity for laser powder bed fusion metal additive manufacturing. *Addit Manuf* 2023, 68: 103500. <https://doi.org/10.1016/j.addma.2023.103500>.
- [51] Meng G, Zhang JD, Li JC, Jiang ZZ, Gong YD, Zhao JB. Impact of pore defects on laser additive manufacturing of Inconel 718 alloy based on a novel finite element model: Thermal and stress evaluation. *Opt Laser Technol* 2023, 167: 109782. <https://doi.org/10.1016/j.optlastec.2023.109782>.
- [52] Lam MC, Lim SCV, Song HY, Zhu YM, Wu XH, Huang AJ. Scanning strategy induced cracking and anisotropic weakening in grain texture of additively manufactured superalloys. *Addit Manuf* 2022, 52: 102660. <https://doi.org/10.1016/j.addma.2022.102660>.
- [53] Harte A, Atkinson M, Preuss M, Fonseca J. A statistical study of the relationship between plastic strain and lattice misorientation on the surface of a deformed Ni-based superalloy. *Acta Mater* 2020, 195: 555-570. <https://doi.org/10.1016/j.actamat.2020.05.029>.
- [54] Ma MY, Wang B, Liu HQ, Yi DQ, Shen FH, Zhai TG. Investigation of fatigue crack propagation behavior of 5083 aluminum alloy under various stress ratios: Role of grain boundary and Schmid factor. *Mater Sci Eng A* 2020, 773: 138871. <https://doi.org/10.1016/j.msea.2019.138871>.
- [55] Zhang ZH, Liu YJ, Wang ZD, Huang JH, Chen SH, Ye Z, et al. Influence behavior and mechanism of γ' -or γ'' -precipitated phase types on damage resistance of heterogeneous interface in nickel-based superalloys. *J Alloy Compd* 2022, 924: 166559. <https://doi.org/10.1016/j.jallcom.2022.166559>.
- [56] Yue JX, Dong Y, Soares CG. An experimental-finite element method based on beach marks to determine fatigue crack growth rate in thick plates with varying stress states. *Eng Fract Mech* 2018, 196: 123-141. <https://doi.org/10.1016/j.engfracmech.2018.04.015>.
- [57] Li W, Sun R, Wang P, Li XL, Zhang YC, Hu TY, et al. Subsurface faceted cracking behavior

of selective laser melting Ni-based superalloy under very high cycle fatigue. Scripta Mater 2021, 194: 113613.

- [58] Fatemi A, Yang L. Cumulative fatigue damage and life prediction theories: a survey of the state of the art for homogeneous materials. Int J Fatigue 1998, 20(1): 9-34. [https://doi.org/10.1016/S0142-1123\(97\)00081-9](https://doi.org/10.1016/S0142-1123(97)00081-9).
- [59] Pavlou DG. The theory of the SN fatigue damage envelope: Generalization of linear, double-linear, and non-linear fatigue damage models. Int J Fatigue 2018, 110: 204-214. <https://doi.org/10.1016/j.ijfatigue.2018.01.023>.
- [60] Santecchia E, Hamouda AMS, Musharavati F, Zalnezhad E, Cabibbo M, Mehtedi, M El, Spigarelli S. A review on fatigue life prediction methods for metals. Adv Mater Sci Eng 2016, 2016(1): 9573524. <https://doi.org/10.1155/2016/9573524>.
- [61] Hectors K, De Waele W. Cumulative damage and life prediction models for high-cycle fatigue of metals: A review. Metals 2021, 11(2): 204. <https://doi.org/10.3390/met11020204>.
- [62] Aeran A, Siriwardane SC, Mikkelsen O, Langen I. A new nonlinear fatigue damage model based only on SN curve parameters. Int J Fatigue 2017, 103: 327-341. <https://doi.org/10.1016/j.ijfatigue.2017.06.017>.
- [63] Marines-Garcia I, Paris PC, Tada H, Bathias C. Fatigue crack growth from small to long cracks in very-high-cycle fatigue with surface and internal “fish-eye” failures for ferrite-perlitic low carbon steel SAE 8620. Mater Sci Eng A 2007, 468: 120-128. <https://doi.org/10.1016/j.msea.2006.08.131>.
- [64] Sun CQ, Lei ZQ, Xie JJ, Hong YS. Effects of inclusion size and stress ratio on fatigue strength for high-strength steels with fish-eye mode failure. Int J Fatigue 2013, 48: 19-27. <https://doi.org/10.1016/j.ijfatigue.2012.12.004>.
- [65] Sun ZD, Li W, Deng HL, Zhang ZY. Fish-eye failure analysis and life design approach for case-carburized gear steel based on statistical evaluation of defect size. Eng Fail Anal 2016, 59: 28-40. <https://doi.org/10.1016/j.engfailanal.2015.11.017>.
- [66] Chowdhury P, Sehitoglu H. Mechanisms of fatigue crack growth—a critical digest of theoretical developments. Fatigue Fract Eng M 2016, 39(6): 652-674. <https://doi.org/10.1111/ffe.12392>.

Conflict of interest

We declare that we have no financial and personal relationships with other people or organizations that can inappropriately influence our work. There is no professional or other personal interest of any nature or kind in any product, service and/or company that could be construed as influencing the position presented in, or the review of, the manuscript entitled.