



Ti6Al4V titanium alloy fatigue strength after AM- and machining-based process chains

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ABSTRACT

The paper proposes the prediction of the fatigue strength of additive manufactured Ti6Al4V titanium alloy machined under flood and cryogenic cooling conditions, making use of the approach based on linear elastic fracture mechanics that was used for the first time in the context of such process chains. The role of the machining-generated surface texture, hardening, microstructural characteristics and residual stresses states is discussed in detail, giving new insights about the correlation between surface integrity and fatigue strength when addressing defective metal alloys, as those fabricated through AM.

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1. Introduction

Amongst the advantages offered by additive manufacturing (AM), "complexity for free" is the most significant. This has been responsible for the expanded use of AM in a variety of industrial applications. However, the structural integrity of AM metal parts remains a concern. The fatigue strength is typically much lower than that of conventionally fabricated parts [1]. This is mainly due to the typical AM-inherent defects, such as lack of fusion, un-melted particles, rough surface finish, uneven microstructural features and residual stress. In addition to limiting the part fatigue life, these deficiencies increase the scatter in fatigue life data [1]. The fatigue performance of as-built AM metals has been widely investigated, showing that fatigue crack initiation is primarily a surface phenomenon. Further, surface topography and sub-surface defects play a major role in determining fatigue life. These conclusions are exemplified in [2] where the high-cycle fatigue of the as-built laser V-notched specimens generated in Ti6Al4V by powder bed fusion (LPBF). Similar results were found in [3] where the as-built LPBF Ti6Al4V was subjected to three-point bending fatigue testing.

Different approaches have been proposed to improve the fatigue performances of AM metal parts. They are based either on the optimization of the AM process itself or on post-AM treatments to properly condition the part surface finish and integrity. The former allows for the reduction of AM-induced defects, but they are often insufficient to ensure an acceptable part fatigue life. On the contrary, the latter can result in fatigue performances comparable to those of the conventionally fabricated metals. Amongst them, surface treatments, including machining operations, are the most investigated, aiming at reducing the surface texture and surface/sub-surface defects as well as inducing

a compressive residual state of stress. Roller burnishing was used in [4] to enhance the fatigue resistance of AM stainless steel by generating a beneficial surface layer and a deep compressive residual stress state. In [5] the fatigue strength improvement on LPBF Ti6Al4V parts was attributed to the surface texture reduction and the compressive residual stress state generated through shot and laser peening.

The scientific literature presents numerous approaches to design AM metal parts against fatigue. Apart from the traditional nominal stress approach that ignores the influence of any AM-inherent defect and, thus, limits its applicability to simple cases, fracture mechanics-based models have recently been shown to be more suitable in predicting the fatigue strength of AM metals. In particular, linear elastic fracture mechanics (LEFM)-based approaches have been adopted as they allow evaluating the severity of the stress field arising close to the defect [6]. Nevertheless, such models have been never applied to AM and machined workpieces, which are characterized by a strong variation of their surface integrity characteristics.

In this context, the objective of the work is twofold: (i) to give new insights on how the integrity of a machined surface affects the fatigue strength, and (ii) to provide theoretical modelling based on LEFM of fatigue strength of metals after AM- and machining-based process chains. LPBF Ti6Al4V machined under flood and cryogenic cooling conditions is the material object of the investigation.

2. Theoretical background on LEFM approach

The estimated fatigue limit of a defect-free metal $\Delta\sigma_{0,est}$ (MPa) can be evaluated as a function of the metal Vickers hardness HV in (kgf/mm²) (Eq. (1)) [6]:

$$\frac{\Delta\sigma_{0,est}}{2} = 1.6 HV \quad (1)$$

When dealing with defective metals, as the AM ones, the Murakami model [6] can be used to predict the fatigue limit $\Delta\sigma_{g,th}$ in (MPa)

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calculated at the gross section (Eq. (2)), and the threshold stress intensity factor of small defects $\Delta K_{th, SC}$ in (MPa $\sqrt{m}$) (Eq. (3)). They are both valid for surface defects and estimated as a function of HV and the $\sqrt{area}$ parameter in (μm) that is the square root of the area of the so-called volumetric “killer defect” projected on the plane perpendicular to the maximum tensile stress.

$$\frac{\Delta \sigma_{g,th}}{2} = \frac{1.43 \cdot (HV + 120)}{(\sqrt{area})^{\frac{1}{6}}} \quad (2)$$

$$\Delta K_{th, SC} = 3.3 \cdot 10^{-3} \cdot (HV + 120) \cdot (\sqrt{area})^{1/3} \quad (3)$$

The Murakami model can be applied provided that the $\sqrt{area}$ parameter of the tested sample is within a lower and an upper bound. The former is the critical defect size $\sqrt{area_c}$ (Eq. (4)), and the latter the transition defect size $\sqrt{area_t}$ (Eq. (5)). The lower bound is estimated on the basis of HV , and the upper bound is determined by using the threshold stress intensity factor of long cracks $\Delta K_{th, LC}$. AM metals typically show process-inherent defects whose size is in the order of magnitude of $\sqrt{area_t}$.

$$\sqrt{area_c} = \left(\frac{1.43 \cdot (HV + 120)}{1.6 HV} \right)^6 \quad (4)$$

$$\sqrt{area_t} = \left(\frac{\Delta K_{th, LC}}{3.3 \cdot 10^{-3} \cdot (HV + 120)} \right)^3 \quad (5)$$

$\Delta K_{th, LC}$ can be empirically estimated on the basis of HV and the microstructural parameter l in (μm) (Eq. (6)) [7].

$$\Delta K_{th, LC, est} \cong 4.5 \cdot l^{0.127} + 2.29 \cdot 10^2 \cdot HV^{-0.81} \quad (6)$$

In the case of the LPBF Ti6Al4V object of the present investigation, l is the width of the α phase lamellae, being the most relevant microstructural feature affecting crack growth [7].

Fig. 1 compares the Murakami model (Eq. (2) in blue line) with the traditional Kitagawa-Takahashi diagram (Eq. (7) in black line), which represents the pioneering model in presence of defective metals, and, therefore, conventionally used as reference.

$$\Delta \sigma_{g,th} = \begin{cases} \Delta \sigma_{0,est} & \text{if } a_{eff} \leq a_{0,est} \\ \frac{\Delta K_{th, LC, est}}{\alpha \cdot \sqrt{\pi \sqrt{area}}} & \text{if } a_{eff} > a_{0,est} \end{cases} \quad (7)$$

where the effective defect size a_{eff} is determined based on the dimensionless defect shape factor α , equal to 0.65 in the case of surface defects (Eq. (8)), while the material length parameter $a_{0,est}$ is given on the basis of $\Delta \sigma_{0,est}$ and $\Delta K_{th, LC, est}$ (Eq. (9)).

$$a_{eff} = \alpha^2 \cdot \sqrt{area} \quad (8)$$

$$a_{0,est} = \frac{1}{\pi} \left(\frac{\Delta K_{th, LC, est}}{\Delta \sigma_{0,est}} \right)^2 \quad (9)$$

The red arrows in Fig. 1 highlight that, contrarily to $\Delta \sigma_{0,est}$, $\Delta K_{th, LC}$ tends to decrease at increasing HV . Additionally, Fig. 1 graphically shows the range of applicability of the Murakami model, which can be significantly low for hard metals, as $\sqrt{area_c}$ and $\sqrt{area_t}$ reduce at HV increasing.

To account for the surface texture, the Murakami model can be modified by using the equivalent defect size $\sqrt{area_R}$, which is

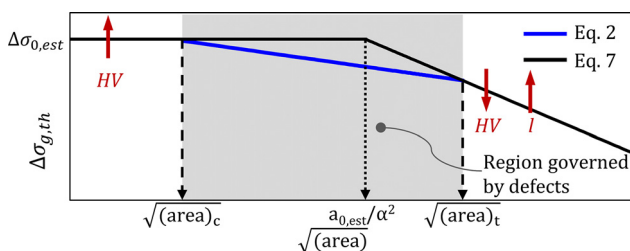


Fig. 1. Fatigue limit according to the Murakami model with emphasis on the region of its applicability.

evaluated on periodic roughness profiles using their pitch and maximum depth [6], equivalent to mean spacing RSm and surface roughness height Rz .

To account for the residual stress state [6], Eq. (2) can be modified in Eq. (10).

$$\frac{\Delta \sigma_{g,th}}{2} = \frac{1.43 \cdot (HV + 120)}{(\sqrt{area})^{1/6}} \cdot \left(\frac{1 - R^*}{2} \right)^\beta \quad (10)$$

where R^* stands for the load ratio ($R^* = \frac{\sigma_{g,min} + \sigma_{res}}{\sigma_{g,max} + \sigma_{res}}$), with σ_{res} representing the average value of the residual stresses, and β is expressed according to Eq. (11).

$$\beta = 0.226 + HV \cdot 10^{-4} \quad (11)$$

In this way the residual stresses are considered equivalent to a local mean stress capable to shift the load ratio $R \left(\frac{\sigma_{g,min}}{\sigma_{g,max}} \right)$.

3. Experimental procedures

The approach used in this study to evaluate the fatigue strength of Ti6Al4V samples after AM- and machining-based process chains is shown in Fig. 2. It consists of the following steps: a. finite element (FE)-based design of the fatigue sample geometry, b. near-net-shape fabrication of the sample through LPBF, c. finishing machining, and d. fatigue testing. The design of the fatigue sample is required to induce fatigue failure for a number of cycles within the high-cycle fatigue regime, namely in the range between 10^4 and $2 \cdot 10^6$ cycles. After having chosen the fatigue test typology (the rotating bending test after described), and the sample geometry, characterized by an hourglass gross section with minimum diameter 4.5 mm and total length 80 mm, the presence of undesired stress concentrations was checked using the commercial FE software AnsysTM. The maximum bending stress at the minimum cross section was calculated when applying a bending moment generating a nominal maximum stress of 1 MPa. Fig. 2a shows that the maximum principal stress was 1.01 MPa instead of 1 MPa, therefore a stress concentration of 1% exists, but negligible. The near-net-shape samples were then fabricated through LPBF by adding 1 mm on the gauge section diameter as metal allowance, and a 19 mm long cylinder to one side to allow the sample fastening on the lathe chuck (Fig. 2b). The LPBF process was carried out on a Sisma MYSINTTM 100 machine using argon gas atomized Ti6Al4V spherical powders with sizes from 15 to 45 μm (Fig. 2b) and shielded by high-purity argon gas. The samples were fabricated vertically (i.e. with the axis parallel to the building direction), using the island scanning strategy with laser power 125 W, scanning speed 925 mm/s, hatch spacing 80 μm , and layer thickness 20 μm . The AM parameters were previously optimized to obtain nearly full-dense samples. Before being removed from the Ti6Al4V build plate, the samples were heat treated at 950 $^{\circ}C$ under argon atmosphere for 30 min and cooled inside the oven until 400 $^{\circ}C$.

The final sample geometry was obtained by turning, firmly fastening the sample to the spindle on one side and to the tailstock on the other, as shown in Fig. 2c. A fresh VCET 110,301-UM 1115 insert was used to machine each sample. The finishing pass was carried out by

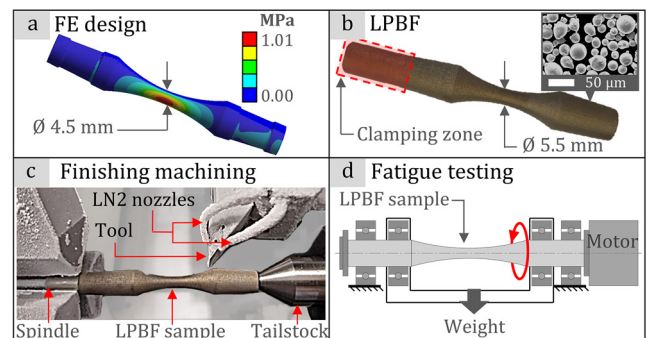


Fig. 2. Research approach: FE design (a), near-net-shape fabrication through LPBF (b), finishing machining (c), fatigue testing (d).

adopting depth of cut 0.25 mm, cutting speed 45 m/min, and feed 0.03 mm/rev. The tool geometry and cutting parameters were chosen according to the tool supplier's recommendations for difficult-to-cut metals. Both flood and cryogenic cooling conditions were used, being the latter well-known to improve the surface integrity and fatigue life of conventionally manufactured metals [8].

The microstructural characteristics of the machined LPBF Ti6Al4V samples were examined on three sections perpendicular to the building direction using the JSM™ 7900F ultra-high-resolution field emission gun scanning electron microscope (FEG-SEM) after grinding, polishing and etching with the Kroll's reagent. The Leitz™ Durimet microhardness tester was used to measure the Vickers microhardness with a load of 50 g for 30 s following the ASTM E92–17 standard. On each polished part, three measurement series with eight indentations at increasing distance from the machined surface were carried out. The axial residual stresses were measured using a Spider™ X GNR portable diffractometer working at 30 kV and 90 mA, at increasing distance from the machined surface by electrolytically removing layers of material.

The machined surface finish was evaluated by scanning an area of $0.67 \times 2.28 \text{ mm}^2$ through the Sensofar™ S-Neox optical profiler with a 20X magnification Nikon™ confocal objective. The arithmetic mean height S_a , reduced pit depth S_{vk} , and skewness S_{sk} were considered in accordance to ISO 25,178–2:2012-Part 2, as informative in relation to the fatigue behaviour of the metal alloys; specifically, S_{vk} and S_{sk} were identified as indicators of the presence of localized crack initiation sites [9].

A $2.5 \mu\text{m}$ S-filter and 0.25 mm L-filter were applied to evaluate the areal texture parameters on the S-L surface, following the ISO 16,610–61:2011 standard. The fatigue performances of the machined LPBF Ti6Al4V samples were assessed using rotating bending fatigue tests under fully reverse loading condition ($R=-1$) at a load frequency of 200 Hz on a METRO COM™ machine (sketch in Fig. 2d). The fatigue tests were considered aborted if one of the two following conditions was reached: (i) the fracture of a whole section, or (ii) the attainment of the $2 \cdot 10^6$ load cycles run-out criterion N_A . The rotational bending loading was chosen over the axial one because the former induces a stress gradient in the sample section so that the maximum stress occurs at its surface. In this way, the effect of the machining-induced surface integrity on the fatigue strength can be highlighted. At the end of each test, a fractography analysis was conducted on the fracture surface through the Quanta 450 FEI™ SEM at different magnifications. The killer defect area was then measured through the ImageJ™ software, following the guidelines given in [6].

4. Results and discussion

The subsurface microstructure after machining is shown in Fig. 3a, b, c and d, proving the formation of a deformed layer close to the machined surface, which is emphasized in the case of cryogenic machined samples. The magnified image of the subsurface microstructure in Fig. 3d shows, indeed, a pronounced bending of the LPBF Ti6Al4V α lamellae, generated by the significant strain hardening induced by the cryogenic cooling. Furthermore, in case of cryogenic cooling, it can be noted a strong microstructural refinement towards the nanometric range close to the surface, with the development of a harmonic structure, namely a combination of nano- and microstructure. Fig. 3e reports the axial residual stresses values at increasing distance from the machined surface. In general, the machining step ensured compressive residual stresses, but the cooling strategy had a significant impact on their intensity. In particular, cryogenic machined samples registered a 22% increase in the maximum compressive residual stress at around $30 \mu\text{m}$ below the surface. The more severely altered microstructure and greater compressed residual stress state result in harder surface and sub-surface of the cryogenic machined samples compared to the flood machined ones as shown in Fig. 3f. The results of the surface topography characterization are reported in Table 1 and based on four measurements for each condition. No significant differences induced by the two cooling strategies when considering the areal texture parameters are highlighted.

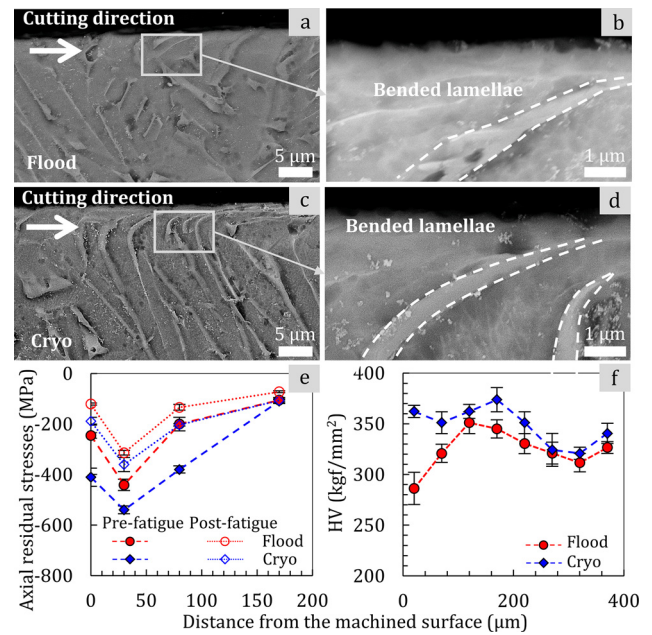


Fig. 3. Microstructures below the machined surface at 3000X (a, c) and 15000X magnifications (b, d), axial residual stresses (e), and micro-hardness (f) as a function of the distance from the machined surface and machining cooling strategy.

Table 1

Areal texture parameters average values and related standard deviations at varying machining cooling strategy.

Cooling strategy	S_a (μm)	S_{vk} (μm)	S_{sk}
Flood	0.39 ± 0.07	0.29 ± 0.02	0.42 ± 0.07
Cryo	0.42 ± 0.02	0.26 ± 0.02	0.35 ± 0.1

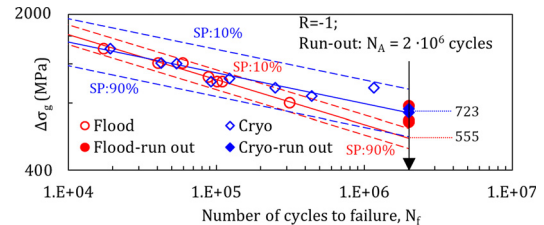


Fig. 4. Fatigue data in terms of nominal stress range at varying machining cooling strategy.

Fig. 4 shows the fatigue test results elaborated on the basis of the nominal stress approach, i.e. the Wöhler curve. The fatigue data are reported in terms of the nominal stress range $\Delta\sigma_g$ vs. the number of cycles to failure N_f , including the scatter bands referring to survival probabilities (SP) of 10% and 90% in dotted lines, with a confidence level of 95%.

Both flood and cryogenic machined samples show comparable fatigue strength below $1 \cdot 10^5$ cycles, above which different fatigue behaviours can be appreciated. The inverse slope k of the Wöhler curve of the cryogenic machined samples is, indeed, 1.5 times higher than flood ones, resulting in 30% higher fatigue strength (723 MPa and 555 MPa for cryogenic and flood samples, respectively). It is worth underlining that the data reported in Fig. 4 give the conventional representation of fatigue performances, neglecting the role that defects intrinsic to the AM process may have in affecting the Ti6Al4V fatigue strength.

To account for AM-inherent defects, the $\sqrt{\text{area}}$ parameter was evaluated by examining the samples fracture surfaces to identify the fatigue fracture initiation site, namely the killer defect. The killer defect was identified from the river marks departing from the crack initiation location (Fig. 5a), and its area calculated (see the magnified view with the defect red contour in Fig. 5b). The identified killer defects were of different nature, i.e. lacks of fusion, un-melted

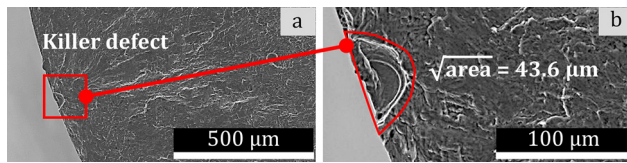


Fig. 5. Fatigue fracture surface (a), and detail of the killer defect of the cryogenic machined LPBF Ti6Al4V sample with $\Delta\sigma_g=1024$ MPa (b).

Table 2

Microstructural parameter, hardness and estimated fatigue properties from LEFM at varying machining cooling strategy.

Cooling Strategy	l (μm)	HV (kgf/mm ²)	$\Delta\sigma_{0,est}$ (MPa)	$\Delta K_{th,LC,est}$ (MPa*√m)
Flood	2.65	320	1024	7.234
Cryo	2.65	360	1152	7.039

particles and porosities, but they were found to be all located within the first 100 μm from the machined surface.

Their average size, expressed as $\sqrt{\text{area}}$, settles at 50 ± 8 μm for the flood, and at 55 ± 20 μm for the cryogenic machined samples.

Table 2 reports the values of the microstructural parameter l , hardness HV , $\Delta\sigma_{0,est}$ from Eq. (1), and $\Delta K_{th,LC,est}$ from Eq. (6). Both l and HV are given in the table as the average of the values recorded within the first 100 μm from the machined surface, since the killer defects always originate from this zone. Therefore, it is worth noting that the differences in the α lamellae thickness shown in Fig. 2b and d are not explicitly taken into account. Nevertheless, the strain hardening induced when deforming at low temperatures and the development of the harmonic structure can be considered responsible of the 12.5% average hardness increase when cryogenic machining compared to flooding. In this sense, the microstructural alterations are suitably caught by the hardness parameter of the model. $\Delta\sigma_{0,est}$ and $\Delta K_{th,LC,est}$ give insights on the fatigue limit in the absence of defects and the cracking propagation resistance, respectively. Specifically, the former increases of 12.5%, whereas the latter reduces of 3% passing from flood to cryogenic cooling. This means that the fatigue strength of hard materials, as is the cryogenic machined LPBF Ti6Al4V, is higher when they are defect-free, but, in case of defectiveness, it becomes more sensitive to defects of lower size, which tend to propagate more rapidly (from Eq. (7)).

The comparison between the experimental fatigue data presented on the basis of the LEFM approach and the ones predicted using the Murakami model Eq. (2) with the values in Table 2 is shown in Fig. 6, together with the critical and transition defect sizes, and the material length parameter $a_{0,est}$ (Eqs. (4), (5) and (9)). The fatigue data of all the samples fall within the short crack regime, included between $\sqrt{\text{area}}_c$ and $\sqrt{\text{area}}_t$. Given that the Murakami model differentiates the fatigue data from unbroken and broken specimens, there is a good agreement between the experimental data and the model outcomes since: (i) all the run-out samples are included within the error bands $\pm 20\%$, (ii) regardless of the cooling strategy, all the broken samples are placed well above the Murakami mean curve. The model can give accurate predictions even when surface texture and residual stresses-related parameters are not considered. The former was excluded since the killer defect analysis proved that the fatigue cracks never originated from the valleys. Regardless of the cooling strategy, finishing machining assures, indeed, the attainment of a sufficiently high level of surface finish (see Table 1), which appears not to influence the fatigue strength. The residual stress state was also excluded as the fatigue tests led to a significant release of that (see data in Fig. 3e), therefore making uncertain its actual effect on the fatigue performances [10].

From Fig. 6, if considering an average size of the killer defects equal to 50 μm and 55 μm for flood and cryogenic machined samples, respectively, the expected fatigue limits $\Delta\sigma_{g,th}$ settle at 655 MPa and 704 MPa, leading to an 8% difference, which is considerably lower than the one derived from the data in Fig. 4. Nevertheless, the LEFM

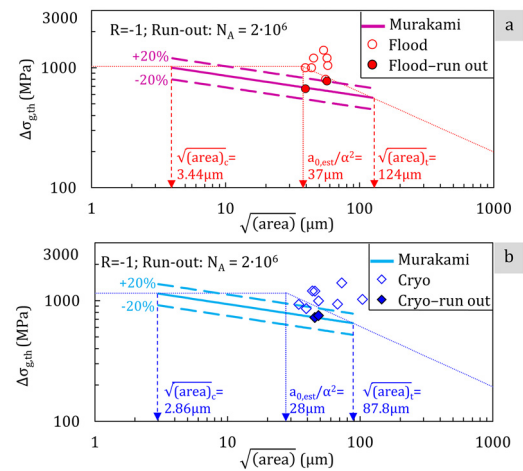


Fig. 6. Comparison between the experimental fatigue data and fatigue limits predicted according to the Murakami model: flood (a) and cryogenic machined (b) LPBF Ti6Al4V samples (error bands $\pm 20\%$).

modelling approach is noticeably more accurate in predicting the fatigue resistance of the LPBF Ti6Al4V as it captures the peculiarities of both AM and machining, that are the presence of AM-inherent defects and machining-induced surface integrity changes.

5. Conclusions

LEFM was successfully applied to predict the fatigue strength of LPBF Ti6Al4V samples machined under flood and cryogenic conditions. It was shown that the generation of hard and harmonic structures is the main responsible of the increase of the fatigue limit of the cryogenic machined samples over the flood machined ones. It was also shown that the traditional fatigue approach based on the nominal stress may lead to misleading insights on the fatigue behaviour, since it neglects the effects played by the AM-inherent defects and machining-induced surface integrity changes.

Declaration of Competing Interest

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

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