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# Very high cycle fatigue (VHCF) response of additively manufactured materials: a Review

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## Abstract

*Recently, the diffusion of components produced through Additive Manufacturing (AM) processes has significantly increased. Concurrently, to ensure the safe employment of AM components, the research has focused on the assessment of their fatigue response. In the present paper, a comprehensive literature review on the Very High Cycle Fatigue (VHCF) response of materials produced through AM processes is provided. The experimental results on the VHCF response of Ti6Al4V, AlSi10Mg, AlSi7Mg, AlSi12, IN718 and 316L specimens produced through different AM processes are summarised. The main factors affecting the VHCF response of AM materials are finally compared and critically analysed.*

*Keywords: Very High Cycle Fatigue (VHCF); Additive Manufacturing (AM); Aluminium Alloy; Ti6Al4V; In718*

## Data Availability Statement

The authors confirm that the data supporting the findings of this study are available within the article and its supplementary materials.

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## Terminology legend

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**VHCF:** *Very High Cycle Fatigue*

**HCF:** *High Cycle Fatigue*

**LCF:** *Low Cycle Fatigue*

**AM:** *Additive Manufacturing*

**BPH:** *Base Plate Heating*

**SR:** *Stress Relief*

**SLM:** *Selective Laser Melting*

**EBM:** *Electron Beam Melting*

**SEM:** *Scanning Electron Microscope*

**LOF:** *Lack of fusion*

**bal.:** *Balance*

**vertically built:** *the angle between the building plate and specimen axis is 90°*

**horizontally built:** *the angle between the building plate and specimen axis is 0°*

**DAS:** *dendrite arm spacing*

**S-N:** *stress-life (number of cycles)*

**wt-%:** *weight percentage*

**HIP:** *hot isostatically pressure*

**μ-CT:** *micro computed tomography*

**SIF:** *Stress Intensity Factor*

**ILB:** *improper layer bonding*

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## Introduction

In the last years, among the emerging processes to produce components employed in engineering applications, the Additive Manufacturing (AM) technology is one of the most employed, since it permits to produce parts with complex geometry, minimizing the manufacturing constraints. AM processes are widely employed for the production of components used in many different applications, like aerospace<sup>1-5</sup>, motor race<sup>6</sup>, healthcare<sup>7,8</sup> and are substituting traditional manufacturing processes in the production of heat exchangers<sup>9,10</sup>, biomedical devices<sup>11</sup> and structural brackets<sup>12,13</sup>. For the production of metallic components, Selective Laser Melting (SLM) and Electron Beam Melting (EBM) are largely used. The research currently focuses on the optimization of these manufacturing processes in order to make them more efficient and effective in the production of components that can be safely used. For what concerns the use of AM components in structural applications, the focus is to guarantee the structural integrity of AM components in applications in which both static and fatigue loads are applied<sup>14-17</sup>. In the literature, the quasi-static response of AM parts has been largely experimentally investigated<sup>18-21</sup>. In addition, the response of AM specimens when fatigue loads are in the Low Cycle Fatigue (LCF) and High Cycle Fatigue (HCF) regions has been also extensively investigated<sup>22-25</sup>, focusing on the crack initiation and propagation mechanisms from manufacturing defects. Design methodologies have been also developed to prevent premature failures due to process-induced defects<sup>16</sup>. On the other hand, the fatigue response of AM parts beyond 10<sup>7</sup> cycles, in the Very High Cycle Fatigue (VHCF) regime<sup>26-28</sup>, has been less extensively investigated in the literature, even if it is well known that the required fatigue of components is significantly larger than in the past and, for several applications, is even above 10<sup>9</sup> cycles. In<sup>23</sup> ("*Regarding the fatigue limit of Ni-based superalloy 718, a supplementary description may be necessary in terms of VHCF.*") and in<sup>29</sup> ("*Due to the lack of VHCF studies for AM materials, there is a critical need to obtain the fatigue behaviour at gigacycles and the influences of design parameters, size/geometry, surface roughness, etc.*") the need for the experimental assessment of the VHCF response of AM parts is clearly pointed out, in order to ensure their structural integrity and prevent failures due to low amplitude loads applied for a very high number of cycles. This would extend the

knowledge on the fatigue response of AM components and would ensure their safe employment even in structural applications.

In the present paper, a comprehensive literature review on the VHCF response of parts produced through AM processes is provided. All the papers, published at present and focusing on the VHCF response of AM specimens, have been investigated. The main factors affecting the VHCF response of AM materials are compared and critically analysed, highlighting the critical aspects that should be considered in the future research in this field.

For the sake of clarity, the paper has been organized by analysing in each Section the VHCF response of a singular material. In **Section 1** and **Section 2** the VHCF response of the AM aluminium alloys is reported. In **Section 3** the VHCF response of Ti6Al4V, in **Section 4** the VHCF response of Inconel 718 and finally in **Section 5** the VHCF response of 316L stainless steel are specifically described.

### VHCF response of AM material

In the following Sections, the fatigue response in the VHCF region of specimens produced through AM processes is analysed in detail. Each investigated material has been considered separately. For each material, the experimental results obtained in the literature are at first analysed and commented in a specific Subsection (Subsection Experimental results) and then compared in the subsequent Subsection (Subsection Discussion).

The experimental data have been retrieved from Tables, when available, or from the Figures, by using the software *Engauge digitizer*. In the following, vertical specimens are those built with their axis perpendicular to the building platform (and parallel to the building direction), whereas horizontal specimens are the specimens produced with their axis parallel to the building platform (and perpendicular to the building direction). The defect at the origin of the fatigue failure will be called “critical defect”.

As will be highlighted in the following Sections, the main fatigue failure cause of AM parts is the presence of defects. Indeed, Additive Manufacturing (AM) materials are characterised by a non-negligible defect population. Minor fluctuations of the process parameters irreversibly lead to local imperfections, defects, and microstructural changes. These defects represent the preferred site for the nucleation and subsequent propagation of cracks, leading to the AM part collapse. The effect of size, location and distribution of the defects is provided by the Murakami model<sup>30,31</sup>, reported in Equation 1.

$$s = \frac{C_1 \cdot (HV + 120)}{(\sqrt{a_c})^{\frac{1}{6}}} \cdot \left(\frac{1 - R}{2}\right)^{0.226 + HV \cdot 10^{-4}} \quad (1)$$

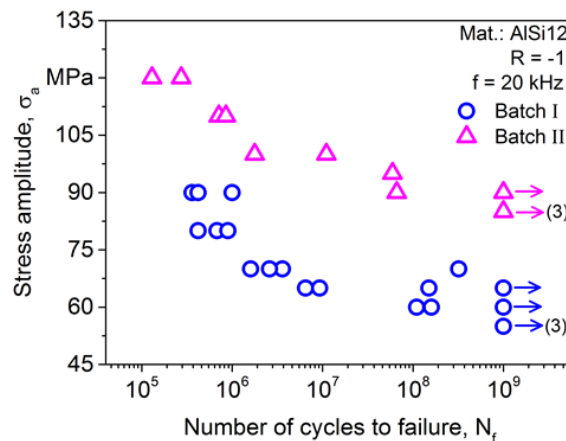
where  $s$  is the VCHF strength at  $10^9$  cycles,  $C_1$  is a constant parameter depending on the defect location,  $HV$  is the Vickers hardness,  $R$  is the stress ratio, and  $\sqrt{a_c}$  is the equivalent critical defect size. This last term includes the effect of size, shape, and orientation of the critical defect. The location of the defect is considered by the constant term  $C_1$ . If the defect is located superficially this term is 40% smaller compared to internally located one. Lastly, the distribution of the defects, i.e.  $\sqrt{a_c}$ , inside the material is proved to follow the Largest Extreme Value Distribution (LEVD) as reported in<sup>31</sup>.

#### 1.1 VHCF response of AlSi12: experimental results

In<sup>32-36</sup> the VHCF response of AlSi12 alloy processed by AM technologies is investigated. The specimens subjected to VHCF tests are manufactured by using an SLM 250 HL machine, by SLM Solution company, with a laser power of 400 W. A chessboard scanning strategy with 90° degrees between the chess fields is used. The orientation field is rotated of 79° layer by layer to obtain isotropic properties in the final component. The fast-cooling rate of SLM additive process leads to a microstructure characterised by fine Si columnar

dendrites along the scanning direction immersed in  $\alpha$ -Al matrix. For a deeper analysis of the employed process parameters, scanning strategies and resulting AlSi12 microstructure the reader is referred to Siddique<sup>37</sup>. The AlSi12 powder contains (wt-%) 11.62 Si, 0.72 Fe, 0.03 Cu and bal. with a mean diameter of 33  $\mu\text{m}$ . Hourglass specimens with gage diameter of 3 mm are produced in vertical direction and thereafter machined before testing. Fully reversed tension compression VHCF tests up to  $10^9$  cycles are experimentally assessed by using the ultrasonic fatigue testing system USF-2000 produced by Shimadzu at a loading frequency of 20 kHz.

In<sup>33</sup> two different batches of specimens are compared, named B1 and B2. B1 is obtained without Base plated heating (BPH) and B2 is obtained by heating the plate up to 200°C. Both undergo to a stress relief (SR) post-process at 240°C followed by oven cooling. Figure 1 displays the S-N plot of the experimental data obtained in [7]. The VHCF strength up to  $10^9$  cycles for specimens belonging to B2 (with BPH) is 45% higher than that for specimens in B1, being  $88.7 \pm 3.3$  MPa for B2 and  $60.5 \pm 4.7$  MPa for B1, indicating a significant influence of the BPH. The Scanning Electron Microscope (SEM) investigation of the fracture surfaces demonstrates that surface and internal defects are the main cause of crack initiation. In B1, crack initiation starts from surface as well as from subsurface pores; in B2, the crack initiation occurs mainly at the surface. The BPH permits to slightly reduce the residual porosity in the material and eliminate the largest process-induced pores thanks to the reduction in thermal gradients, enhancing the VHCF response.



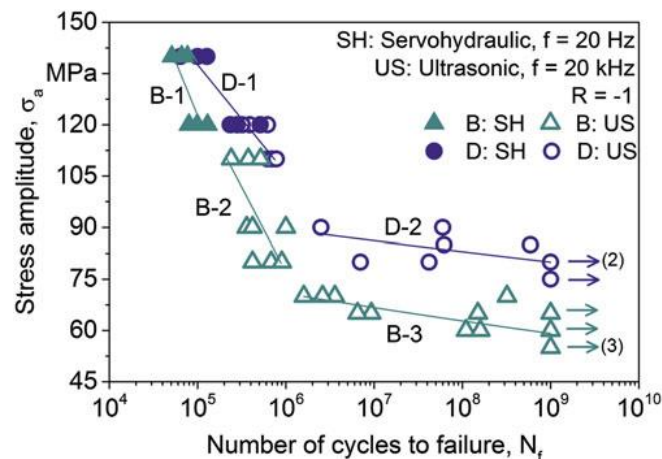
**Figure 1** - S-N plot of batch B1 (no BPH) and batch B2 (with BPH) from Siddique<sup>33</sup>

In<sup>34</sup> the influence of the BPH and the SR process on the fatigue life of AlSi12 is investigated. In particular, this study follows and confirms the results already obtained in<sup>33</sup>, focusing mainly on the defect originating failures. Four batches, named A, B, C, and D are manufactured with the process parameters reported in Table 1. The microstructure is characterised by a very fine dendritic structure, with width range 0.3-0.6  $\mu\text{m}$  and Si precipitates. The porosity for all the samples results to be inferior to 0.3 %: for batch A, 0.26%; for batch B, 0.38%; for batch C, 0.13%; for batch D, 0.29%. Hence, the BPH decreases the overall porosity. On the other hand, the analysis on the defect size points out that the SR induces an expansion of the existing pores due to the internal pressure of trapped gas at high temperatures. Furthermore, the absence of BPH causes a non-uniform distribution of pores in terms of size and location. Overall, the BPH enhances the VHCF response due to inferior pores fraction and pore size. This effect is less evident in the HCF regime where even small pores, present in all the batches, become critical. It can be concluded that the VHCF strength is mainly controlled by micro-porosity, since only in a small number of failures the crack initiates from other material defects.

Batch	Base plate heating (BPH)	Stress relief (SR)
A	No	No
B	No	240°C
C	200°C	No
D	200°C	240°C

**Table 1** – Parameters of the investigated batches in Siddique<sup>34</sup>

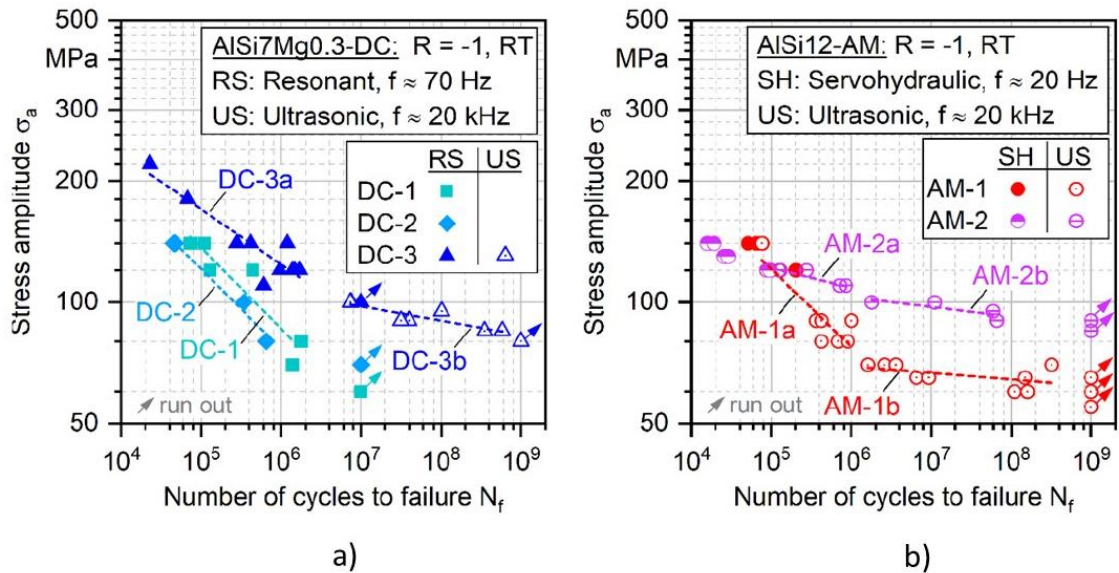
In Siddique<sup>35</sup>, a comparison between the VHCF response of “pure” AlSi12 and “hybrid” AlZn4.5Mg1 + AlSi12 specimens is carried out. The hybrid batch is obtained by using the AlZn4.5Mg1 as solid base material in the build chamber and by melting AlSi12 on the top of this base material as in the classical SLM process. The specimens are produced without heating the building platform and are subjected to the SR post-process at 240°C followed by oven cooling at the end of the AM process. The hybrid specimens are built such that the interface between the two different materials lies in the middle of the longitudinal axis. The experimental tests are carried out up to  $10^9$  cycles: in particular, fatigue tests with a servo hydraulic testing machine (loading frequency of 20 Hz) are carried out up to  $10^6$  cycles; whereas ultrasonic fatigue tests are carried out to investigate to VHCF region. Figure 2 shows the S-N plot of the experimental data: Batch B refers to pure specimens, whereas Batch D to hybrid specimens. The VHCF strength up to  $10^9$  cycles of the hybrid material is equal to 80.7 MPa according to the S-N plot, considerably larger than that of the pure AlSi12 (about 59 MPa). The fatigue fracture in the hybrid specimens mainly occurs in the SLM part, however the global fatigue limit is enhanced due to the progressive hardening which occurs in the material.



**Figure 2** - S-N plot of batch B (pure AlSi12) and batch D (hybrid AlZn4.5Mg1 + AlSi12) from Siddique<sup>35</sup>

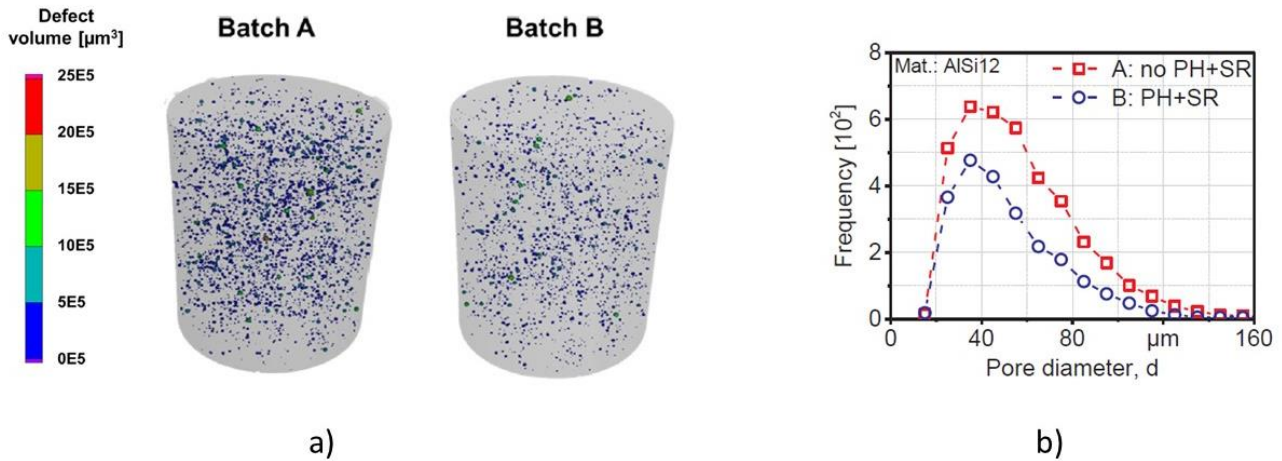
In Tenkamp<sup>32</sup> a comparison between the VHCF strength of die-cast alloy AlSi7Mg0.3 and additively manufactured AlSi12 is carried out. Die-cast (DC) samples are manufactured by gravity die-casting, including alloying procedure with AlTi5B1 and Sr. The melt is heated up to 730°C and the steel mould is pre-heated to 450°C. Three batches are produced: DC1 as cast, DC2 with melt cleaning and DC3 with melt cleaning and HIP process by Bodycote Densal™. All DC batches are afterwards subjected to a T6 process. AM AlSi12 specimens are produced including BPH. Two different batches are manufactured: batch AM1 including as-built specimens and batch AM2 including specimens subjected to the SR post-process at 240°C in oven. Before testing, the gage diameter of all batches is ground and polished up to 1  $\mu$ m diamond paste. The microstructure of the DC specimens shows a dendritic width about 2 order of magnitude larger than that of the AM specimens. In addition, the pores in DC1 and DC2 are in average one order of magnitude bigger and less spherical than DC3, AM1 and AM2. Figure 3a shows the S-N plot of the DC batches; whereas Figure 3b shows the S-N plot of the AM batches. As for the DC specimens, the DC3 is characterised by the highest VHCF

strength up to  $10^9$  cycles, approximately 83 MPa by graphical extrapolation, outperforming the other DC batches. For the AM specimens, the AM2 batch (SLM with BPH and SR) is characterized by the largest VHCF strength (approximately 88 MPa for the AM2 and 62 MPa for the AM1). The enhancement of the VHCF response of AM2 specimens is justified by the Authors by considering that the SR process increases the sphericity of the internal pores, reducing the influence of internal micro-notches. Moreover, the combination of BPH and SR in AM AlSi12 permits to obtain a VHCF strength larger than that of traditional die cast AlSi7Mg0.3 alloy.



**Figure 3** - S-N plot of: a) batch DC (die cast process) and b) AM (1-no SR, 2-with SR) from Tenkamp <sup>32</sup>

In Awd <sup>36</sup> the influence of process parameters and post-processes on the microstructure and VHCF response behaviour of AlSi12 and AlSi10Mg alloy is investigated. Two different batches are compared, named A and B. A is obtained without Base plated heating (BPH) and B is obtained by heating the plate up to 200°C. Both undergo to a stress relief (SR) post-process at 240°C followed by oven cooling. The process parameters are the same detailed at the beginning of this paragraph. Micro Computed Tomography (micro-CT or  $\mu$ -CT) analyses have been carried out to compare the porosity of specimens. The results obtained by the Authors are shown in Figure 4a. As it can be noticed, BPH (batch B) reduces consistently the pore density in the specimen. Figure 4b shows the frequency of occurrence of pore diameters: in both the batches the mean value of the pore diameter is approximately 40  $\mu$ m. However, the BPH remarkably decreases the pore density, according to Figure 4b.



**Figure 4** - a) Micro-computed tomography and b) frequency-pore diameter plot of batches A and B from Awd<sup>36</sup>

*Note: (PH: plate heating)*

This analysis confirms the experimental results found in previous research: the BPH significantly diminishes the micro-porosity and fosters the VHCF strength. The results on the AlSi10Mg alloy are analysed in **Section 2**.

## 1.2 VHCF strength of AlSi12: discussion

In this Section the experimental data on the VHCF response of AM AlSi12 alloy obtained in the literature are critically compared. The analysis of the fracture surfaces reveals that the VHCF response of AlSi12 alloy is driven by process induced porosity in the material. The threshold stress level which separates the HCF from the VHCF regime is approximately 100 MPa. The influence of BPH and of SR treatment on the VHCF response is mainly investigated.

The BPH reduces the final porosity fraction in the material, increasing by about the 45% the VHCF strength, even if the mean pore diameter is left unaltered: indeed, BPH permits to reduce the size of the largest pores. The BPH with the SR produces a reduction in the material hardness of about the 15% compared to as-built samples. For more, in specimens produced without BPH the scatter associated to the VHCF strength is larger due to a less uniform distribution of the pores, in terms of size and location. An SR treatment increases the dimension of the pores due to the gas expansion in the heating process, but in contrast, it also enhances the VHCF strength of about 20% by increasing the sphericity of the pores and limiting internal micro-notches due to irregular shapes. The VHCF response of hybrid material AlZn4.5Mg1 + AlSi12 is 25% larger than that of pure AlSi12.

In conclusion, BPH positively affects the VHCF response, since it permits to reduce the porosity density and avoid the formation of very large pores. On the other hand, an SR treatment could have a positive influence on the VHCF response, since it enhances the defect sphericity, but it can also induce an increment of the pore size, which should be avoided in case of failures from defects. Compared to the traditional die cast AlSi7Mg0.3 alloy, the additively manufactured AlSi12 outlines a 7% improved VHCF strength. Table 2 summarises the experimental results on the VHCF response of AlSi12, reporting the main process parameters investigated in the literature.



<b>Pure AlSi12</b> (SLM process and final machining)					
Specimen type	Building orientation	Base plate heating (BPH)	Stress relief (SR) and oven cooling	Runout	VHCF strength [MPa]
Hourglass <sup>33</sup>	Vertical	\	240°C	10 <sup>9</sup>	60.5 ± 4.7
Hourglass <sup>33</sup>	Vertical	200°C	240°C	10 <sup>9</sup>	88.7 ± 3.3
Hourglass <sup>35</sup>	Vertical	\	240°C	10 <sup>9</sup>	59
Hourglass <sup>32</sup>	Vertical	200°C	\	10 <sup>9</sup>	62
Hourglass <sup>32</sup>	Vertical	200°C	240°C	10 <sup>9</sup>	88
<b>Hybrid AlZn4.5Mg1 + AlSi12</b> (SLM process and final machining)					
Hourglass <sup>35</sup>	Vertical	\	240°C	10 <sup>9</sup>	80.7

**Table 2** – VHCF strength of AM AlSi12 depending on process setup.

### 1.3 VHCF strength of AlSi12: summary

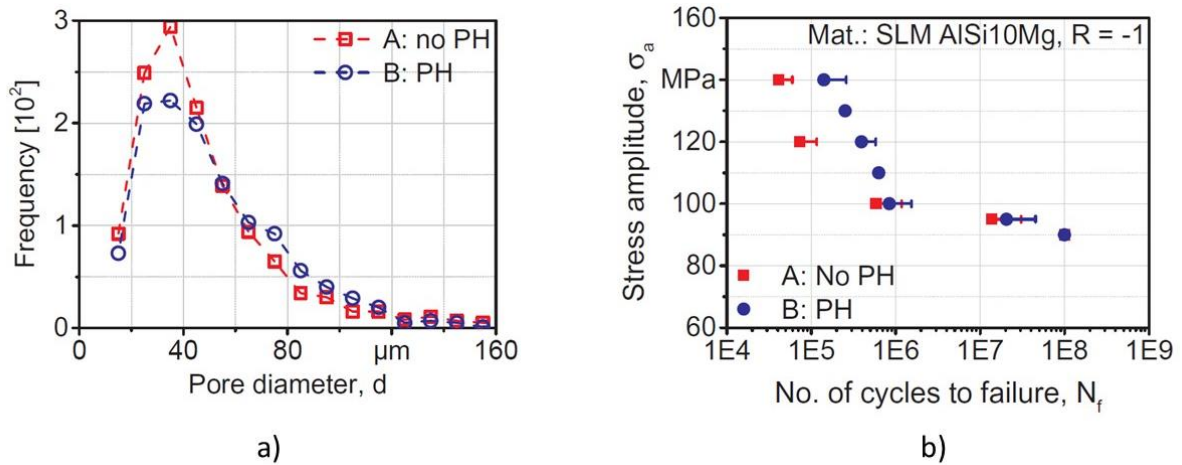
For the AlSi12 specimens, the influence of the base plate heating (BPH) and of a stress relief (SR) heat treatment are mainly investigated:

- The BPH (200°C) reduces the final porosity and increases the VHCF strength.
- SR (240°C – 2h) treatment enhances the VHCF strength.
- The hybrid material AlZn4.5Mg1 + AlSi12 presents an improved VHCF performance.

### 2.1 VHCF strength of AlSi10Mg and AlSi7Mg: experimental results

The effect of BPH on the VHCF response of AlSi10Mg specimens produced through SLM is experimentally assessed in<sup>38</sup>. A modified SLM 250 HL system with an external laser source with laser power of 1 kW is employed and a two-step scanning strategy consisting of two contour scans followed by the core bidirectional scanning is employed to ensure high-quality parts and porosity less than 0.5%. Fully reversed tension-compression tests at 20 kHz up to 10<sup>8</sup> cycles are carried out with a Shimadzu USF-2000A ultrasonic testing system on hourglass specimens built in vertical direction and thereafter machined to the final geometry. Two different batches are analysed: batch A is obtained without BPH and batch B is obtained by heating the plate up to 200°C. No SR post-process is applied, in order to isolate the effect of the BPH. The microstructure of the specimens in both batches is characterized by an Al matrix with embedded network of equiaxed cellular Si boundaries parallel to the build platform and Si columnar dendrites along the building axis, with local coarsening along the melt pool track overlaps. However, the microstructure of the batch B (with BPH) presents a coarser morphology, a greater dispersion and supersaturation of Si in the Al matrix, fewer Si semi-coherent particles and a larger number of Mg<sub>2</sub>Si precipitates at the grain boundaries. The manufacturing defects have been also analysed: the average pore diameter is similar in both the batches (Figure 5a) even if the pores in batch B are characterised by a more spherical shape. Batch B shows an enhanced quasi-static, LCF and HCF strength due to the strengthening effect induced by these microstructure modifications. On the other hand, batches A and B show similar VHCF strength up to 10<sup>9</sup> cycles, being about 90 MPa, according to the S-N plot in Figure 5b. Indeed, the VCHF response is found to be driven by the micro-porosity in both batches: more specifically, in batch A the crack initiation starts from a single critical defect while in batch B multiple crack initiations occur. The BPH reduces the largest pore size but increases the pore number in the

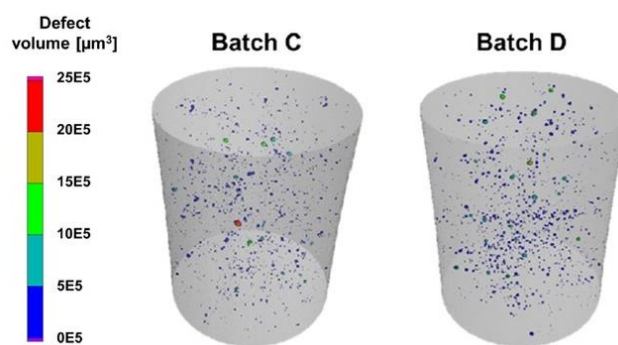
material. The Authors conclude that the fatigue strength of platform heated specimens is higher in the VHCF regime due to the influence of defect morphology and microstructure. However, the increment is not significant due to the large scatter of the experimental data. The main effect of BPH is the reduction of the scatter associated to the VHCF results.



**Figure 5** –Experimental results in <sup>38</sup>: a) Frequency of occurrence with respect to pore diameter; b) Fatigue life mean and standard deviation on a S-N plot.

*Note: (PH: plate heating)*

Awd, in <sup>36</sup>, further analyses the experimental results in <sup>38</sup>, in particular by considering the defect density and distribution. The  $\mu$ -CT of the samples obtained in [11] is reported in Figure 6. As it can be noticed, the  $\mu$ -CT proves that the BPH has not a remarkable effect on the final porosity. This confirms why BPH does not affect the VHCF response <sup>38</sup>. Spherical gas porosities located sub-superficially result to be the most critical defect for the VHCF response.

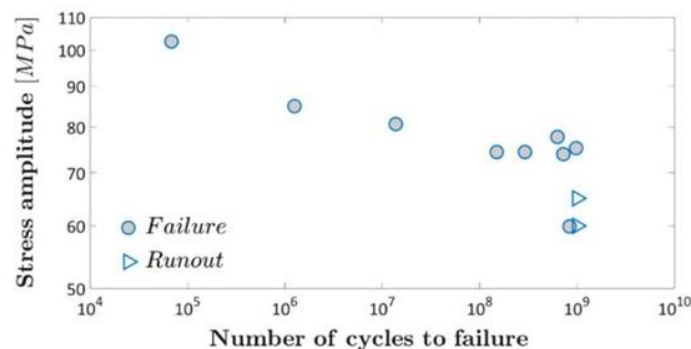


**Figure 6** - Micro-computed tomography of batches C and D from Awd <sup>36</sup>

In References <sup>39–44</sup>, fully reversed ultrasonic tension-compression VHCF tests up to  $10^9$  cycles on AISi10Mg Gaussian specimens are carried out. The tested Gaussian specimens are characterized by a risk-volume <sup>31</sup> (above  $2000 \text{ mm}^3$ ) significantly larger than that of hourglass specimens commonly tested in the literature, permitting to take into account the dependency between the defect size distribution and the risk-volume. The SLM specimens are produced with an SLM Solutions system (model 500 HL with 4 continuous fibre lasers

with power of 400W). Standard industrial process parameters by the SLM system are selected to manufacture high-quality parts (including BPH at 150°C). Spherical gas atomized powder with a mean diameter of 45  $\mu\text{m}$  is used. The manufactured specimens are manually polished to reduce the roughness of the surface and remove a layer of surface material thinner than that removed with other processes (e.g., machining). The microstructure is characterised by columnar solidification structure following the shape of localised melting pools. A complete and detailed description of the AlSi10Mg microstructure by SLM process is provided in <sup>45,46</sup>.

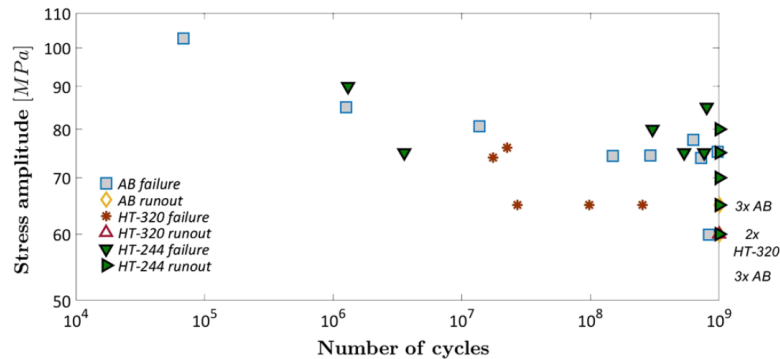
In Tridello <sup>39</sup>, the VHCF strength of horizontally as-built AlSi10Mg samples is investigated. Figure 7 shows the S-N plot of the experimental dataset: on the abscissa axis the stress amplitude close to the defect location,  $s_{local}$ , is reported. The smallest  $s_{local}$  at runout is found at 68 MPa. Surface and sub-surface defects are found to be responsible for the crack initiation in all the experimental failures. The reasons for this behaviour are mainly three. First, defects close to the surface are known to be more detrimental than internal ones (i.e., the Stress Intensity Factor (SIF) of surface defects is approximately 32% larger than the SIF of internal defects <sup>31</sup>). Second, residual stresses tend to be larger close to the surface due to higher thermal gradients. Third, the horizontally built specimens need to be supported in the overhang portion of the surface. The afterwards support removal creates many superficial defects which are detrimental for the fatigue response. The critical defect type is also analysed, being in the 67% of the cases cluster of pores, in the 22% surface defects and in the 11% lack of fusion (LOF) defects. According to these analyses, the Authors conclude that the removal of the most external layer of material, e.g., through a machining process, would eliminate the most defect-dense zone and would likely enhance the VHCF strength.



**Figure 7** - S-N plot of the experimental data obtained considering the local stress close to the defect size  $s_{local}$  from <sup>39</sup>.

In Tridello <sup>40</sup> the influence of SR heat treatments on the VHCF response of AlSi10Mg specimens is investigated. Two SR heat treatments characterized by two different heating temperatures are considered. The first one, HT-320, is suggested by the AM system manufacturer and involves annealing process for two hours at 320°C followed by oven cooling. The second one, HT-244, is optimized by the Authors and involves annealing at 244°C for 2 hours and then oven cooling. The investigated SR treatments are applied to specimens built in horizontal direction and their effect on the VHCF response is investigated by considering the experimental results on as-built (AB) specimens in <sup>39</sup>. Microstructural analyses are at first carried out with the SEM; in particular it is found that a heating temperature of 320° C modifies the final microstructure. Indeed, the continuous network of eutectic Si, typical of AB specimens, is modified and the final microstructure is characterized by embedded spherical Si particles within the Al matrix. The microstructural change induces a significant reduction of the Vickers hardness, about 42%, and of the tensile strength (about 50%), whereas the elongation to failure significantly increases (2.4 times). On the contrary, the HT-244 post-process induces

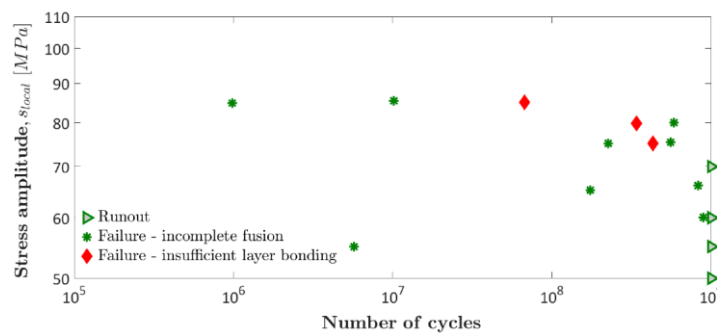
a coarsening of the Si network, but it does not modify its morphology. Consequently, the reduction of the HV hardness is limited to 13% with respect to AB specimens and the tensile properties are close to that of AB specimens. The S-N plot of experimental data with respect to stress amplitude close to the defect location  $s_{local}$  is shown in Figure 8.



**Figure 8** - Stress-life (S-N) plot of the experimental data obtained considering  $s_{local}$  from Tridello <sup>41</sup>

The defect originating the fatigue failures are analysed. As for AB specimens, surface and sub-surface cluster and single pores are the most critical defects in HT-320 specimens. In HT-244 the most critical defects are LOF, differently from AB and HT-320. The graphically extrapolated VHCF strength at  $10^9$  cycles for AB, HT-320 and HT-244 is 68 MPa, 61 MPa and 76 MPa respectively. The VHCF response of HT-244 specimens is improved due to the minimization of the residual stresses: on the other hand, the microstructural changes and the consequent Vickers hardness reduction in HT-320 significantly affects the VHCF response, even if the residual stresses are minimized.

In Paolino <sup>42</sup> the VHCF response of AlSi10Mg as-built samples built in vertical direction is investigated. As for horizontally built specimens [14], the microstructure is characterized by a continuous eutectic Si network embedded in a supersaturated Al matrix, elongated in the building direction. The S-N plot of experimental data with respect  $s_{local}$  is shown in Figure 9. The VHCF response shows a large scatter, mainly associated to the large dispersion associated to the critical defect size. The VHCF strength at  $10^9$  cycles is graphically estimated as approximatively 63 MPa. The SEM inspection of fracture surface reveals that two different types of sub-superficial defects initiate the crack: LOF defects and defect due to improper layer bonding (ILB defects). Most of the critical defects are found to be smaller than  $200 \mu\text{m}$ , but three significantly larger defects, with size larger  $200 \mu\text{m}$  are also found. This contributes to the large experimental scatter: for example, the specimens containing the largest defect (equivalent size of  $286 \mu\text{m}$ ) failed at a number of cycles smaller than  $10^7$  for a stress amplitude of 58 MPa. On the other hand, a specimen tested at the same stress amplitude, did not fail at  $10^9$  cycles.



**Figure 9** - Stress-life (S-N) plot of the experimental data obtained considering  $s_{local}$  from In Paolino <sup>42</sup>

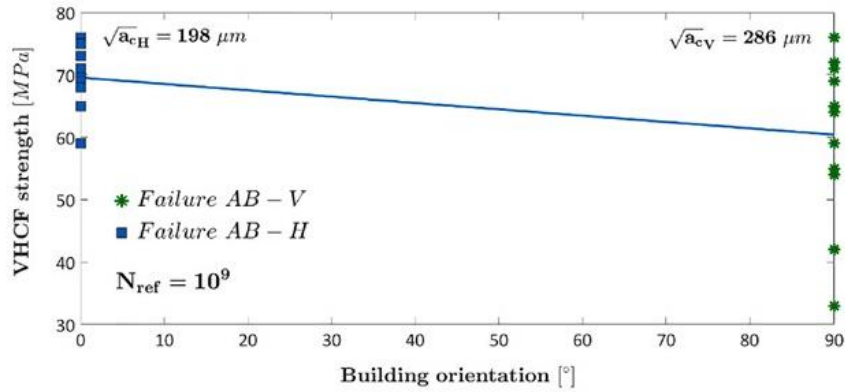
In <sup>43</sup> a critical analysis on the effect of microstructure, residual stress and building orientation on VHCF response of AlSi10Mg is carried out by analysing the experimental data in <sup>39-42</sup> with models available in the literature and statistical models proposed by the Authors. In particular, the VHCF strengths at  $10^9$  cycles for the defect with median size in each group are at first compared (being  $s_{AB-H}$ ,  $s_{HT-320}$  and  $s_{HT-244}$  the VHCF strength at  $10^9$  cycles for AB, HT320 and HT244 specimens, respectively) according to a statistical procedure reported on the paper. The ratios  $s_{AB-H}/s_{HT-320}$ ,  $s_{HT-244}/s_{HT-320}$  and  $s_{HT-244}/s_{AB-H}$  and the corresponding confidence intervals are computed. This analysis confirms that the difference between  $s_{HT-244}$  and  $s_{HT-320}$  is statistically significant, proving that the microstructure plays a significant role and significantly affects the VHCF response (i.e., in both specimens the residual stresses are minimized and the difference is mainly due to the microstructure). On the other hand, no statistical difference between  $s_{AB-H}$  and  $s_{HT-320}$  and between  $s_{AB-H}$  and  $s_{HT-244}$  is found, even if the median  $s_{HT-244}$  is found to be the largest. By equalling the ratio  $s_{HT-244}/s_{AB-H}$  computed with Equation 1 with the same ratio computed by considering the median VHCF strength at  $10^9$  cycles, the residual stresses in AB specimens can be computed. Table 3 summarises the main experimental data employed in the model and the estimated residual stresses for AB specimens.

Batch (from <sup>41</sup> )	Critical defect size (mean) $\sqrt{a_c}$ [ $\mu\text{m}$ ]	HV hardness	Residual stress (estimated) [MPa]	Stress Ratio (estimated)	VCHF strength $s_{10^9}$ [MPa]
AB	167	132	About 100	About 0.2	68
HT-320	88	76	0	-1	61
HT-244	183	115	0	-1	76

**Table 3** – Parameters for Murakami model in Tridello <sup>43</sup> and estimated residual stresses in AB specimens.

According to Table 3, residual stresses in AB are found to be close to 100 MPa, in agreement with literature results in <sup>47</sup> where the maximum principal residual stresses are experimentally assessed. These analyses justify the experimental results: indeed, no statistical difference is found between  $s_{AB-H}$ ,  $s_{HT-320}$ , since the reduction of the Vickers hardness counterbalances the larger median defect and the presence of residual stresses in AB specimens. On the other hand, the larger median defect in HT-244 specimens is balanced by the high residual stresses in AB specimens: for this reason, the median VHCF strengths at  $10^9$  cycles in AB and HT-244 do not significantly differ. A similar analysis is carried out by the Authors by comparing the VHCF strengths at  $10^9$  cycles for the largest defect found experimentally ( $\sqrt{a_c}$  equal to 365  $\mu\text{m}$ ), proving that there is an interaction between defect size and residual stresses and that the detrimental effect of residual stresses increases with the defect size.

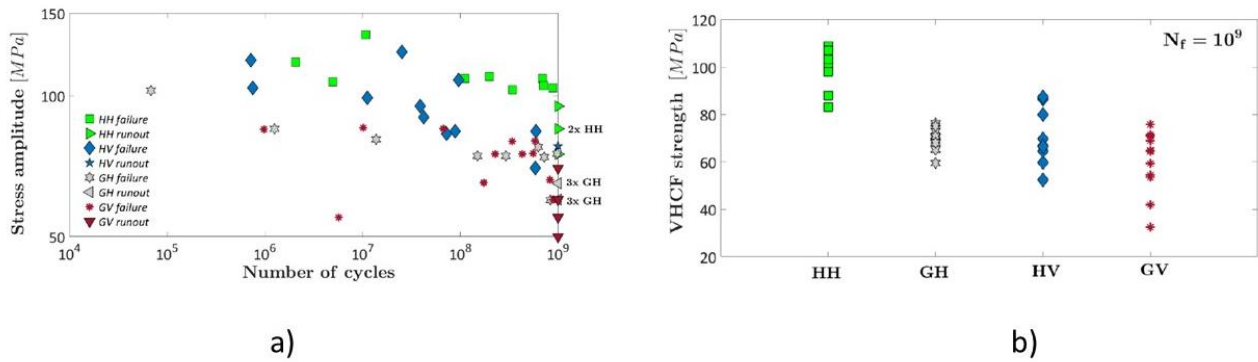
The effect of the building orientation is also investigated in <sup>43</sup>. Figure 10 depicts the VHCF strengths of horizontal as-built specimens (AB-H from <sup>39</sup>) and vertical as-built specimens (AB-V from <sup>42</sup>) at  $10^9$  cycles with respect to the building angle ( $0^\circ$  corresponds to the horizontal building orientation, whereas  $90^\circ$  corresponds to the vertical building orientation). The VHCF strength at  $10^9$  cycles is computed by considering the largest defect in AB-H and in AB-V. As shown in Figure 10, the VHCF strength is smaller for AB-V specimens, characterised by more dispersed experimental failures due to a large defect size range. The VHCF strength decrement with respect to the building angle is properly modelled with a linear interpolation.



**Figure 10** - VHCF strength at  $10^9$  cycles for as-built specimens with respect to the building orientation: experimental data and linear regression from Tridello <sup>43</sup>.

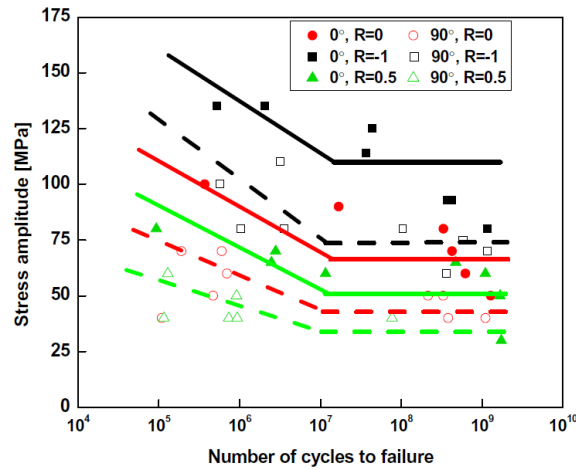
Note: ( $\sqrt{a_{cH}}$ : largest critical defect size horizontal specimens,  $\sqrt{a_{cV}}$ : largest critical defect size vertical specimens,  $N_{ref}$  reference number of cycles )

In Tridello <sup>44</sup>, the effect of the specimen volume (size-effect) on the VHCF response of AlSi10Mg is investigated. The experimental results obtained by testing hourglass specimens, with a risk volume of 250 mm<sup>3</sup>, and Gaussian specimens, with a significantly larger risk-volume of 2330 mm<sup>3</sup>, are analysed. Hourglass and Gaussian specimens are manufactured in horizontal and vertical direction to assess the influence of the building orientation on size-effect. In the paper, the specimen type is identified by two capital letters: the first letter refers to the specimen type (H for the Hourglass and G for the Gaussian) and the second letter refers to the building orientation (H for the horizontally built specimens and V for the vertically built specimens). Figure 11a shows the S-N plot of the experimental data, whereas Figure 11b shows the estimated VHCF strengths at  $10^9$  cycles. The graphically estimated mean VHCF strengths at  $10^9$  cycles can be considered approximatively as  $97 \pm 10$  MPa for HH,  $68 \pm 6.2$  MPa for GH,  $68 \pm 13$  MPa for HV and  $55 \pm 16$  MPa for GV. As already reported in Paolino <sup>42</sup>, the vertical building direction induces larger scatter and smaller VHCF strength. Similarly, the VHCF strength of Gaussian specimens is smaller than that of hourglass specimens due to size-effects. Considering the horizontal batches, all the critical defects are superficial pores located in the support-removal zones. Indeed, this region is characterized by a local smaller Vickers hardness and a larger pore density regardless of the sample geometry. Therefore, in the horizontal specimens, the size-effect is mainly related to the pore size, with the largest critical defect in Gaussian specimens 1.6 times larger than the largest critical defect in hourglass specimens. For the vertical specimens, the critical defects in the hourglass specimens are mainly superficial pores. On the other hand, larger LOF and ILB critical defects are found in Gaussian specimens. Therefore, size-effects not only affect the pore size, but also the pore type. Experimental results confirmed that size-effects strongly affect the VHCF response, regardless of the building orientation and must be carefully taken into account when the AM parts have to be designed.



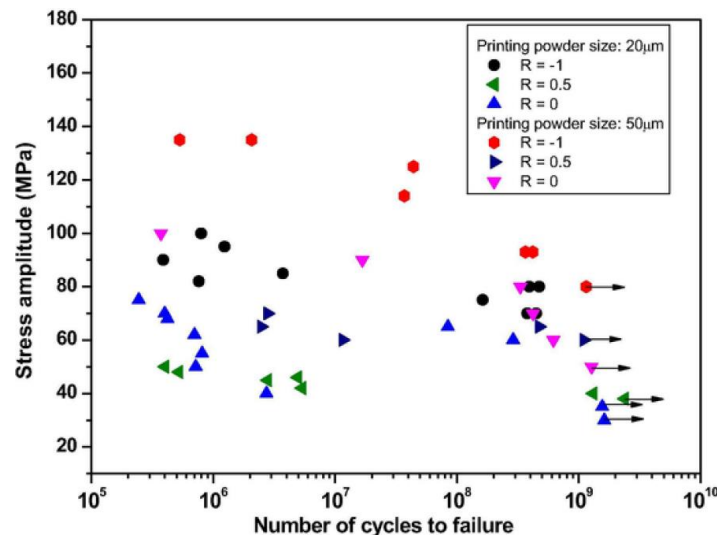
**Figure 11** - a) Stress-life (S-N) plot of the experimental data obtained considering  $s_{local}$  and b) estimated VHCF strength at  $10^9$  cycles from Tridello <sup>44</sup>

In Qian <sup>48</sup> the influence of building orientation and mean stress on the VHCF response of AlSi10Mg alloy specimens manufactured by SLM process is investigated. Hourglass specimens with gage diameter of 3.5 mm are manufactured with an SLM machine with a laser power of 370 W and with a preheating temperature of 35°C. Alloy powder with a mean diameter of 50  $\mu\text{m}$  containing (wt-%) 9.75 Si, 0.22 Mg, 0.01 Cu, 0.01 Cr, 0.01 Ni, 0.092 Fe, 0.01 Mn, 0.01 V, 0.01 Zn, 0.011 Ti and bal. is used. The average surface roughness is 2.526  $\mu\text{m}$ . The VHCF tests are carried out up to  $10^9$  cycles with a Lasur GF20-TC ultrasonic testing system (loading frequency of 20 kHz) integrated in a tensile machine, thus allowing to vary the mean stress. The effect of the building orientation is also investigated, by testing specimens produced in vertical and horizontal direction. The microstructure, in agreement with previous research, shows a fine Si network embedded in Al matrix in the plane parallel to the building plate whereas a dendritic elongated structure perpendicularly to the building plate. Horizontal and vertical specimens are tested at stress ratios,  $R$ , equal to -1, 0 and 0.5. The experimental results are reported in the S-N plot in Figure 12, which shows that horizontally built specimens are characterized by a better VHCF response for the same stress ratio. As already found in <sup>42</sup>, the vertically built samples are characterized by a lower VHCF strength mainly due to the presence of larger defects. In addition to the increased defect size, in vertical built specimens the crack propagation occurs in the same layer, facing fewer obstacles, and it is fostered by possible existing discontinuities. The effect of the applied mean stress is clear, with the VHCF strength decreasing with the stress ratio. The critical defects are: superficial, sub-superficial and internal LOF, internal pores, unspecified defects from the rough surface and tearing defects. The graphically estimated mean VHCF strengths at  $10^9$  cycles for horizontal specimens and  $R$  equal to -1, 0 and 0.5 are 110 MPa, 66 MPa and 51 MPa respectively. For batch B the counterpart values are 74 MPa, 43 MPa and 34 MPa.



**Figure 12** - S-N plot of batches A and B under different mean stress from Qian <sup>48</sup>

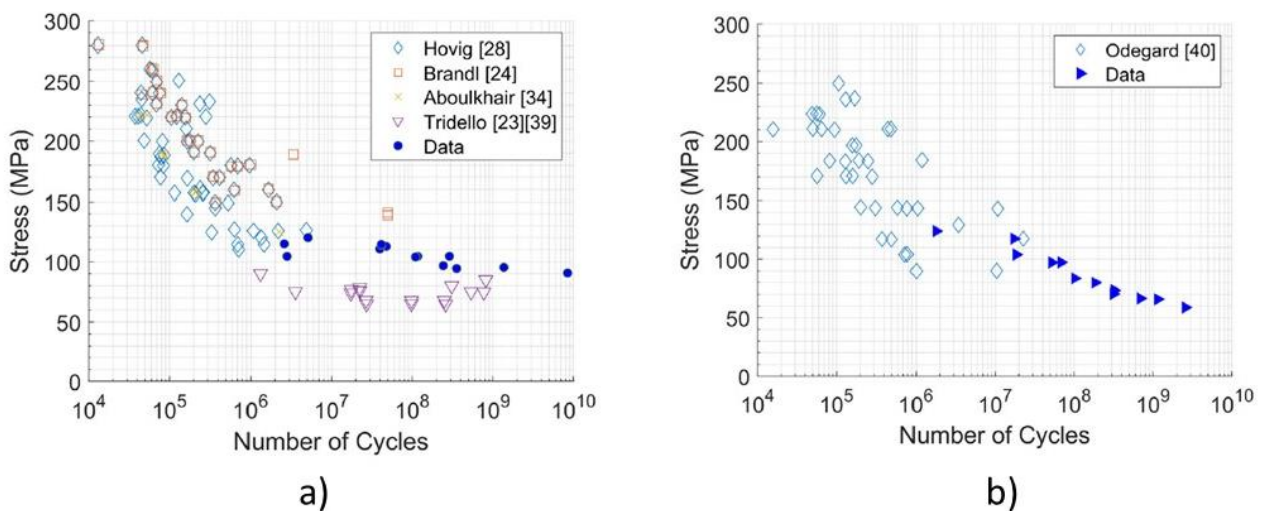
In Jian <sup>49</sup> the effect of different powder sizes on the VHCF response of AlSi10Mg alloy manufactured by SLM process is investigated. Hourglass specimens with gage diameter of 3.5 mm are manufactured with an SLM machine with a laser power of 370 W and with a preheating temperature of 35°C. The chemical composition of the powder is the same as in <sup>48</sup>. The specimens are machined and polished to the final geometry without heat treatment. The VHCF tests are carried out with an ultrasonic testing system at 20 kHz up to 10<sup>9</sup> cycles. The specimens are produced by using a powder with mean size 20 μm (batch A) and a powder with mean size of 50 μm (batch B). The tests are carried out at different stress ratios (-1, 0 and 0.5). As for the microstructure, the results are in line with the previous described SLM AlSi10Mg samples, the morphology of the melt pool shows a fine grain area with smaller size grains and a coarser grain area at the boundary. The overlapping of the laser scan leaves a heat-affected zone with sparse grain distribution. Figure 13 shows the S-N plot of the experimental data. Two different conclusions are drawn by the Authors: first, the increase of the stress ratio diminishes the VHCF strength as already found in <sup>48</sup>; second, the powder size strongly influences the VHCF response. The specimens produced with a smaller powder size are characterized by higher roughness, smaller hardness and the critical defects are larger. The main reasons are the worse powder fluidity and the larger powder surface area that fosters gas inclusion. In batch A the critical defects are mostly sub-superficial and internal unmelt-powder. The largest defect size in batch A is almost double than the largest in batch B. The graphically estimated mean value of VHCF strength at 10<sup>9</sup> cycles and R equal to -1 for batch A is approximately 87 MPa, whereas it is 99 MPa for batch B.



**Figure 13** - S-N plot of batches A and B under different mean stress from Jain <sup>49</sup>



In Lesperance<sup>50</sup> the VHCF response of machined and heat-treated SLM AlSi10Mg and AlSi7Mg alloys is investigated. Cylindrical specimens are manufactured vertically with an SLM machine (EOS M290) with a laser power of 370 W and with a preheating temperature of 180°C, following ISO 4497 standards. AlSi10Mg and AlSi7Mg alloy powders with diameter smaller than 63 µm containing mainly (wt-%) 9.0-11.0 Si, 0.2-0.45 Mg and 6.7-7.3 Si, 0.5-0.6 Mg are used, respectively. The AM cylindrical specimens are subjected to a T6 heat treatment, namely exposure at 535°C for 2 hours followed by water quenching and ageing at 170°C for 6 hours. Finally, the specimens are machined to the final dog-bone (gage diameter 4 mm) geometry for fatigue testing. An ultrasonic fatigue testing machine developed in-house by the Authors is used with a frequency of 20kHz. More details about this VHCF testing system is provided in<sup>51</sup>. Figure 14a shows the S-N plot of the AlSi10Mg experimental data compared with literature heated-treated AlSi10Mg data. At 10<sup>9</sup> cycles, the VHCF strength is graphically estimated to be around 94 MPa. The Authors highlight the difficulties in comparing experimental data from different experimental setups and claim that the effect of heat treatment is crucial, and it needs further investigations for practical applications. Figure 14b shows the S-N plot of the AlSi7Mg experimental data compared with literature cast AlSi7Mg data.



**Figure 14** - S-N plot of: a) AlSi10Mg batch compared with literature AlSi10Mg heat-treated data and b) AlSi7Mg batch compared with literature cast AlSi7Mg data from Lesperance<sup>50</sup>

*Note: The legend in the plot link to literature VHCF data which are in this paper renumbered as Tridello<sup>43</sup> (Tridello [23] in the plot) Tridello<sup>40</sup> (Tridello [39] in the plot). For the other references about HCF data, the reader is referred to Ref.*

<sup>50</sup>.

The results show a reasonable agreement and the VHCF strength up to 10<sup>9</sup> cycles is graphically estimated to be approximately 65 MPa. Compared to the AlSi10Mg batch, the VHCF strength is about 30% smaller, which means the AlSi7Mg alloy has a worse VHCF response under these conditions.

## 2.2 VHCF strength of AlSi10Mg and AlSi7Mg: discussion

Three research groups have focused their experimental activity on the assessment of the VHCF response of AlSi10Mg specimens. The main objectives are the investigation of the influence of the BPH<sup>38</sup>, of the heating temperature on residual stresses in SR heat treatments<sup>40</sup>, of size-effects<sup>44</sup>, of the stress ratio<sup>48</sup>, of the building orientation<sup>42,44,48</sup>, and of the powder size<sup>49</sup>. In all the cases, the defects control the VHCF response and are responsible for the fatigue crack initiation. All the experimental results agree that the process parameters or the post processes allowing for a reduction of the defect size have a positive effect on the

VHCF response. For example, BPH does not have a remarkable effect on the final porosity and, therefore, it mainly contributes to limit the scatter associated to the VHCF response and fosters the microstructure homogenisation, but it does not permit a significant enhancement of VHCF response. Together with the defect size distribution, the microstructure can play a significant role. Indeed, the as-built microstructure can be modified by varying the heating temperature in SR heat treatment. A heating temperature of 320° C in SR induces the rupture of the Si network and its spheroidization, with a consequent significant reduction of the Vickers hardness (about the half of the HV of as-built specimens). The reduction of the HV induces a reduction of the VHCF strength, highlighting that also the microstructure is important when dealing with the VHCF response of material. On the other hand, by limiting the heating temperature in ST to 244°C, the Si network remains intact and the residual stresses are minimized, enhancing the VHCF response. Residual stresses are also important: SR heat treatment permitting to minimize residual stresses are beneficial, provided that the microstructure is not significantly altered. Moreover, all the researchers agree that the building orientation affects the VHCF response, which is largest for horizontally built specimens and decreases with the building angle. The reason is the different defect size and the larger defect range and scatter in vertically built specimens. According to <sup>42</sup>, defects in horizontally built specimens are mainly concentrated in the region where supports are removed: therefore, the removal of the external layer of the samples through a post treatment (e.g., machining) eliminates the most defect-dense zones of the specimens, enhancing the final VHCF response. Size-effect is found to be relevant for both horizontally built and vertically built specimens: the experimental results in <sup>44</sup> warns on the need of properly modelling size-effects when components are to be designed. As expected, moreover, the stress ratio affects the VHCF response, which decreases as the stress ratio increases. The powder size is another important parameter <sup>49</sup>. Indeed, the VHCF strength of specimens with larger powder size is larger than that of specimens manufactured with smaller powder size. The difference is related to the different size of the pores. As for the defect population, in the case of non-machined specimens, superficial and sub-superficial defects (pores, pore clusters and LOF) are the most critical. In vertically built specimen, ILB defect type is critical too. On the other hand, for machined specimens, the VHCF failure can also initiate from internal defects, mainly pores with a critical diameter of around 40 µm. Typically, the machining process removes 1 mm of external layer. For AM components, in the external layer of material the presence of defects is higher, and the defect size is larger compared to the internal defects. The presence of these large superficial defects drives the VHCF response, lowering the VCHF strength and leading to a premature failure. For this reason, machining is always suggested, when possible, to enhance the VHCF strength of the material. However, it must be noticed that for many real complex-shaped AM components, machining is not feasible and process-induced superficial defects must be considered during the fatigue design. Indeed, a difference in magnitude of about 50% can be present between VHCF strength of machined <sup>38</sup> and non-machined <sup>42</sup> specimens. In addition, the study shows that AlSi10Mg has better thermodynamic properties during the AM process providing a much more pore-free structure compared to AlSi12. The presence of Mg induces a grain refinement, the supersaturation of Si in Al matrix and the Mg2Si precipitation. In VHCF regime, finer grains and reduced porosity enhance related VHCF strength since they are the most effective combination to resist crack initiation. In <sup>50</sup>, a traditional T6 heat-treatment is applied on machined specimens of AlSi10Mg. This post-process seems to improve the VHCF response of the material; however, further investigations about its effect on the microstructure and defect population is required. Finally, in <sup>50</sup> the VHCF response of AlSi7Mg alloy is investigated. The AlSi7Mg VHCF strength is smaller compared to its AlSi10Mg counterpart; however, the available experimental data is too small to draw reliable conclusions. Table 4 summarises the VHCF strength for each analysed batch together with main process and testing parameters.

Specimen type	Building orient. <sup>1</sup>	Mechanical post-process	Mean powder size [μm]	BPH [°C]	SR - oven cooling	Runout	Stress ratio <i>R</i>	VHCF strength [MPa]
H <sup>38</sup>	90°	mac	\	\	\	10 <sup>8</sup>	-1	90
H <sup>38</sup>	90°	mac	\	200	\	10 <sup>8</sup>	-1	90
G <sup>39</sup>	0°	pol	45	150	\	10 <sup>9</sup>	-1	68
G <sup>40</sup>	0°	pol	45	150	320°C - 2h	10 <sup>9</sup>	-1	61
G <sup>41</sup>	0°	pol	45	150	244°C - 2h	10 <sup>9</sup>	-1	76
G <sup>42</sup>	90°	pol	45	150	\	10 <sup>9</sup>	-1	63
G <sup>44</sup>	0°	pol	45	150	\	10 <sup>9</sup>	-1	68±6.2
G <sup>44</sup>	90°	pol	45	150	\	10 <sup>9</sup>	-1	55±16
H <sup>44</sup>	0°	pol	45	150	\	10 <sup>9</sup>	-1	97±10
H <sup>44</sup>	90°	pol	45	150	\	10 <sup>9</sup>	-1	68±13
H <sup>48</sup>	0°	\	50	35	\	10 <sup>9</sup>	-1	110
H <sup>48</sup>	0°	\	50	35	\	10 <sup>9</sup>	0.5	51
H <sup>48</sup>	0°	\	50	35	\	10 <sup>9</sup>	0	66
H <sup>48</sup>	90°	\	50	35	\	10 <sup>9</sup>	-1	74
H <sup>48</sup>	90°	\	50	35	\	10 <sup>9</sup>	0.5	34
H <sup>48</sup>	90°	\	50	35	\	10 <sup>9</sup>	0	43
H <sup>49</sup>	0°	mac + pol	50	35	\	10 <sup>9</sup>	-1	99
H <sup>49</sup>	0°	mac + pol	20	35	\	10 <sup>9</sup>	-1	87
DG <sup>50</sup>	90°	mac	<63	180	T6	\	-1	94
<b>AlSi7Mg (SLM process)</b>								
DG <sup>50</sup>	90°	mac	<63	180	T6	\	-1	65

**Table 4** – VHCF strength of AM AlSi10Mg and AlSi7Mg depending on process setup

*Note: mac – final machining, pol – final polishing, H – Hourglasses, G-Gaussian, DG-Dog Bone,<sup>1</sup> – angle between building plate and specimen axis, i.e., 90° vertical and 0° horizontal, T6-traditional heat treatment T6 for aluminium alloy.*

### 2.3 VHCF strength of AlSi10Mg and AlSi7Mg: summary

For the AlSi10Mg specimens, the influence of the BPH, of the heating temperature in SR heat treatments, of the building orientation, of residual stresses and of the powder size is investigated:

- 2.1 BPH (200°C) is suggested to avoid unexpected premature failures and mainly contributes to limit the scatter associated to the VHCF results, but it does not permit a significant enhancement of VHCF response.
  - SR (240°C – 2h) improves the VHCF strength, whereas SR (320°C – 2h) reduces it.
  - Microstructure, together with defect size and residual stresses, influences the VHCF response: heat treatments, which minimize the residual stresses but also involve a significant reduction of the Vickers hardness, could not be effective and could have a negative influence on the VHCF response.

- The building orientation influences the VHCF response: horizontal specimens are characterized by a VHCF response better than that of vertically built specimens.
- The powder size influences the VHCF response: the VHCF strength of specimens produced with powder with mean size of 20  $\mu\text{m}$  is smaller than that of specimens produced with powder with mean size of 50  $\mu\text{m}$ .
- A traditional T6 heat-treatment, together with BPH and machining, improves the VHCF response even if more analyses about the resulting microstructure and defect population are needed.

For the AlSi7Mg a limited number of experimental data are available, and it is difficult to draw reliable insights about the VHCF response of this alloy produced through AM. However, at a first glance, AlSi7Mg parts seem to have a VHCF strength smaller than that of AlSi10Mg parts.

### 3.1 VHCF strength of Ti6Al4V: experimental results

In this Section the results of experimental tests on the VHCF response of AM Ti6Al4V specimens are analysed. In Wycisk<sup>52</sup> the effect of different heat-treatment and post processing on Ti6Al4V alloy produced by means of SLM technology is investigated. A commercially available M270 manufacturing system (laser power 200 W) and an alloy powder with a mean diameter of 41  $\mu\text{m}$  by EOS company are employed for the production of the specimens. Two different batches are produced, named in this paper batch A and batch B, subjected to two different post-processes. Batch A undergoes to an SR at 650°C for 3 hours in vacuum, batch B is HIPed at 920°C and 1000 bar for 2 hours. All the specimens are produced vertically and finally machined and manually polished. Fully reversed tension compression VHCF tests are carried out with an ultrasonic fatigue testing system, USF-2000 by Shimadzu, at a frequency of 20 kHz. The microstructure of batch A consists of fine  $\alpha$ -phase and  $\alpha'$ -martensite lamellae with size inferior to 1  $\mu\text{m}$  and columnar grains of  $\beta$ -phase along the building direction. On the other hand, the HIP procedure removes the columnar structure, makes the  $\alpha$ -lamellae (up to 4  $\mu\text{m}$ ) coarser and induces the formation of a small amount of  $\beta$ -phase within the  $\alpha$ -lamellae. More important, in batch B prior observed gas porosities and process inherent defects are minimised or removed. Furthermore, a reduction in the yield strength and an increase in ductility is observed. The S-N plots of experimental data for batch A and batch B are shown in Figure 15a and in Figure 15b, respectively. As it can be noticed, the datasets show a monotonic decreasing trend, without a clear fatigue limit. In all the specimens, the crack initiates from internal defects. As for batch B, the reduction in defect population together with the homogenization of the microstructure leads to a highly improved VHCF strength and less scattered results. The graphically estimated VHCF strength up to  $10^9$  cycles of batch B is about 483 MPa, comparable with that of a traditional manufactured Ti alloy, whereas it is about 205 MPa for batch A, about 60% smaller. Overall, the HIP treatment is found to be extremely efficient in improving the VHCF response of the AM Ti6Al4V alloy.

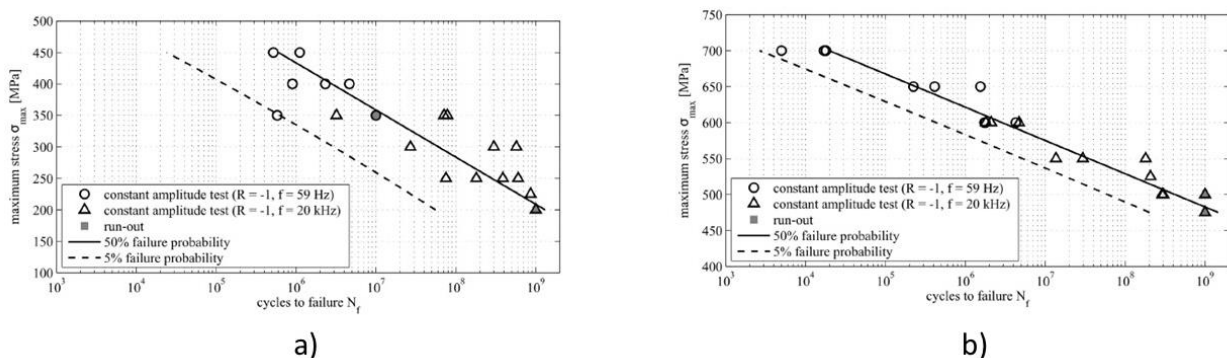
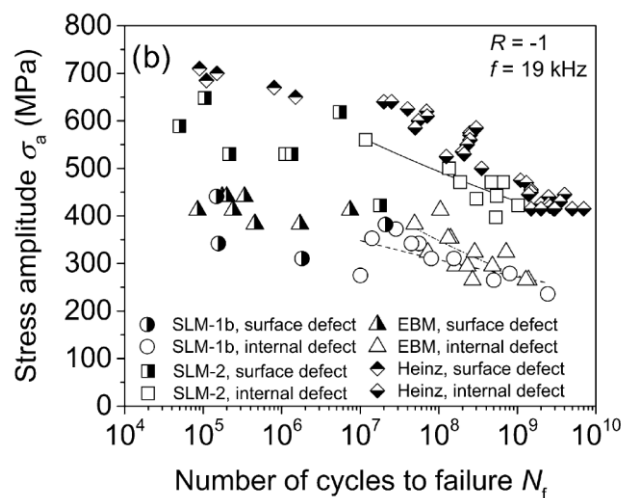


Figure 15 - S-N plot of: a) batch A and b) batch from Wycisk<sup>52</sup>.

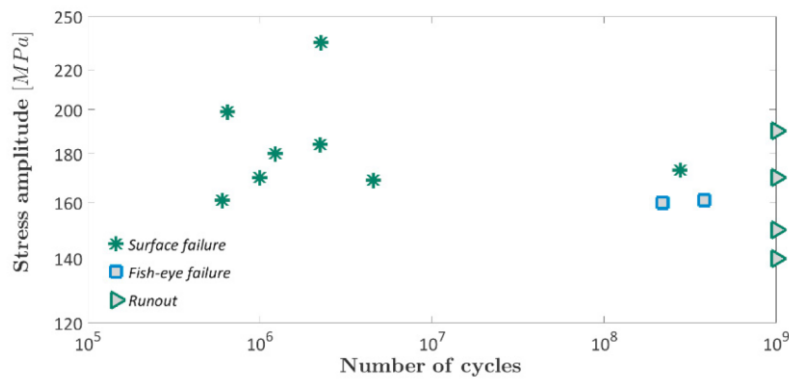
In Gunther<sup>53</sup> the VHCF strength of additively produced Ti6Al4V alloys through SLM and EBM process is investigated. Three different batches, named SLM-1b, SLM-2 and EBM, are produced. The first two batches are produced with an SLM process (SLM system by SLM Solution company). The alloy powder is composed of (wt-%) 7.1 Al, 4.3 V and bal. with a mean diameter of around 42  $\mu\text{m}$ . The building chamber (BPH) is heated up to 200°C and a laser power of 400 W is employed. SLM-1b is heat treated up to 800°C for 2 hours, whereas SLM-2 is HIPed at 920°C and 1000 bar for 2 hours in Ar atmosphere. As for the EBM batch, an Arcam A2X system by Arcam company is employed to manufacture the samples. The alloy powder is composed of (wt-%) 7.3 Al, 4.5 V and bal. with a mean diameter of around 73  $\mu\text{m}$ . No post heat treatment is carried out on EBM samples. Raw cylinders are built vertically (SLM batches) and horizontally (EBM batch) and thereafter machined to the final tested hourglass geometry (gage diameter 4 mm). Lastly, all the samples are ground and polished before testing. The SEM investigation shows that the microstructure of SLM-1b and EBM is composed by a fine lamellar  $\alpha$ -phase and minor amount of  $\beta$ -phase, whereas SLM-2 is composed by coarser partially spheroidized  $\alpha$ -lamellae and interlamellar  $\beta$ -phase. Fully reversed tension compression VHCF tests are carried out with an ultrasonic fatigue testing system at a frequency of 19 kHz. Figure 16 shows the S-N plot of the experimentally tested batches together with literature results of tests carried out on Ti6Al4V specimens produced with traditional processes<sup>54</sup>. According to Figure 16, the graphically estimated VHCF strength up to  $10^9$  cycles of batch SLM-1b and EBM is very similar, around 275 MPa. Indeed, these two batches share almost the same microstructure and defect population. The critical defects are internal LOFs and pores. As for the SLM-2 batch, the graphically estimated VHCF strength up to  $10^9$  cycles is highly improved and nearly doubled compared to that of SLM-1b and EBM, around 435 MPa. The main reason is the homogenization of the structure and the minimization of the internal LOFs and pores through the HIP treatment, as already found in<sup>52</sup>. Accordingly, the crack mainly initiates at  $\alpha$ -phase cluster and facets: in both cases, the applied stress for crack propagation is larger, thus justifying the enhanced VHCF response.



**Figure 16** - S-N plot of batch SLM 1b (SR), SLM-2 (HIP), EBM and conventional process by Heinz<sup>54</sup> batch from Gunther<sup>53</sup>.

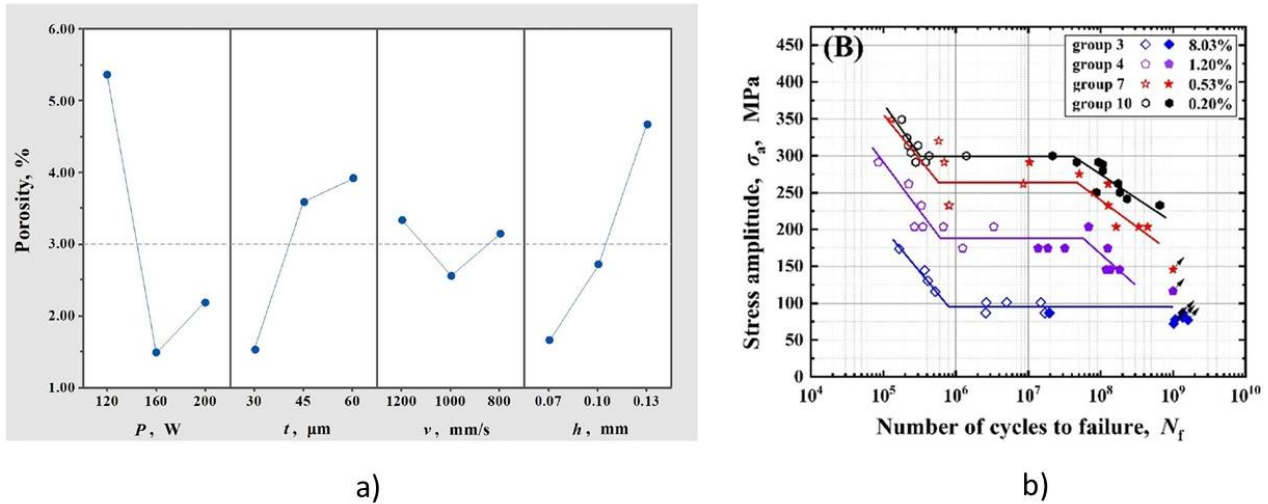
In Tridello<sup>55</sup> the VHCF response of heat treated SLM Ti6Al4V is experimentally assessed. A Renishaw AM400 system with laser power of 400 W, standard process parameters and Ti6Al4V powder (mean diameter of 45  $\mu\text{m}$ ) are used to manufacture the tested as-built Gaussian specimens<sup>56</sup> (risk-volume of 2300  $\text{mm}^3$ ). The specimens are vertically produced, then heat-treated at 850°C for 1 hour followed by cooling in flowing Ar atmosphere and thereafter manually polished to reduce the surface roughness and avoid premature HCF failures due to large surface defects. Two self-assembled Ultrasonic Fatigue Testing Machines (UFTMs) are employed for the VHCF tests up to  $10^9$  cycles. The microstructure is mainly composed of  $\alpha$ -lamellae and local

$\beta$ -phase. The S-N plot of experimental data with respect to  $s_{local}$  is shown in Figure 17. The results show a large scatter and a clear monotonic decreasing trend is not visible and the VHCF strength is concentrated in the range 145 - 200 MPa. Critical defects are ILBs and LOFs, 80% superficial and sub-superficial and 20% located internally. Depending on the defect location, the VHCF response changes remarkably: according to Figure 17, the two failures originated from internal defects are characterized by a fatigue life significantly larger (two order of magnitude) than the fatigue life of the failure originated at a surface defect and subjected to the same applies stress amplitude. Therefore, the removal of an external layer of material can enhance VHCF response.



**Figure 17** - Stress-life (S-N) plot of the experimental data obtained considering  $s_{local}$  from Tridello<sup>55</sup>.

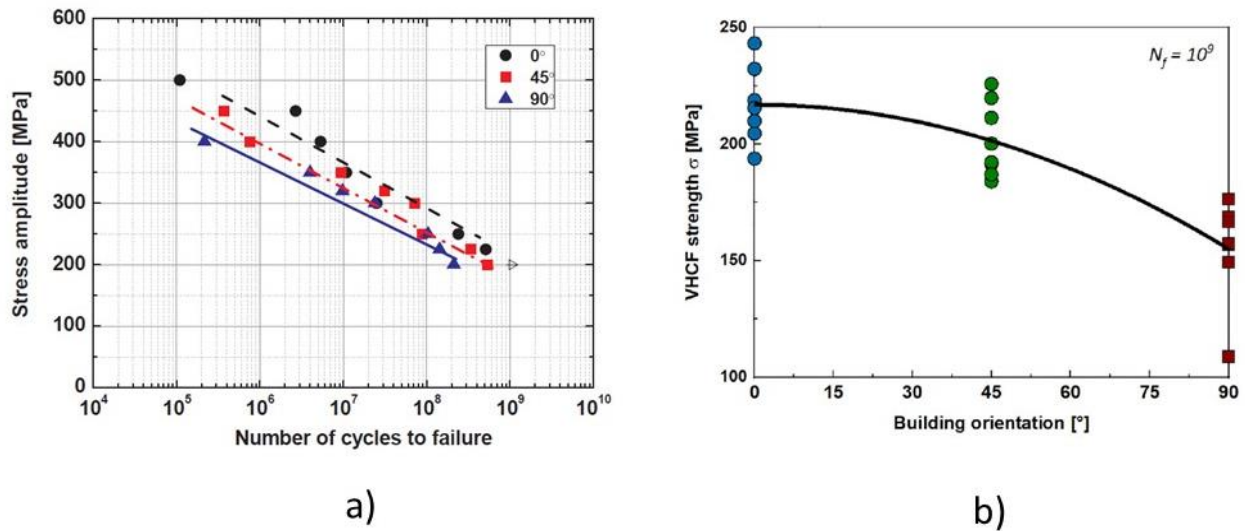
Du<sup>57</sup> investigates the influence of the SLM process parameters on the VHCF response of Ti6Al4V. A commercially available HBD-200 SLM machine with a maximum power of 200 W is employed for the specimen production. The scanning procedure firstly defines the section contour in the layer and then proceeds with infilling the inner areas. The laser travel direction for the infill scanning is rotated by 60° in each subsequent layer. An SR at 600°C for 2 hours followed by flowing Ar cooling is applied to all specimens, which are finally polished to reduce roughness. The alloy powder is composed of (wt-%) 6.40 Al, 4.1 V, 0.01 C, 0.19 Fe, 0.01 N and bal. with a mean diameter of around 34  $\mu$ m. Vertical hourglasses specimens (gage diameter 3.5 mm) are subjected to fully reversed tension compression VHCF tests up to 10<sup>9</sup> cycles with a Lasur GF20-TC ultrasonic testing system (loading frequency of 20 kHz). The achieved microstructure of SLM Ti6Al4V is deeply detailed in<sup>58,59</sup>. An Orthogonal Design of Experiments (OED) with 9 different batches is carried out to assess the effect of the main four process parameters, namely laser power  $P$  [W], hatching distance  $h$  [mm], layer thickness  $t$  [ $\mu$ m], and scan speed  $v$  [mm/s]. Figure 18a shows the relationship between the final porosity and the selected process parameters, whose influence follows the order: laser power > hatching distance > layer thickness > and scan speed. The optimal parameters combination, used for the production of an additional validation batch numbered 10, results to be: laser power of 160 W, hatching distance of 0.07 mm, layer thickness of 30  $\mu$ m and scan speed of 1000 mm/s. As for the VHCF response, Figure 18b shows the S-N plot of some of the most significant batches. For the highest values of porosity (8.03%, worst-case scenario) the graphically estimated VHCF strength up to 10<sup>9</sup> cycles is the lowest (about 95 MPa), whereas in the validation group, characterized by the lowest porosity density, (0.20%, best case scenario), the graphically estimated VHCF strength up to 10<sup>9</sup> cycles is the highest, about 215 MPa. The most critical defects in the VHCF regime are sub superficial LOFs, which form mainly due to a bad combination of process parameters, such as insufficient laser power for a selected hatching distance or layer thickness.



**Figure 18** – VHCF response from Du <sup>57</sup>: a) Porosity vs. process parameters ; b) stress-life (S-N) plot.

*Note: b) full dense marks indicate crack initiation from subsurface or internal defects, hollow marks from surface defects and with arrows are run outs. Group 10 is the reference batch and the value in percentage indicates the related porosity.*

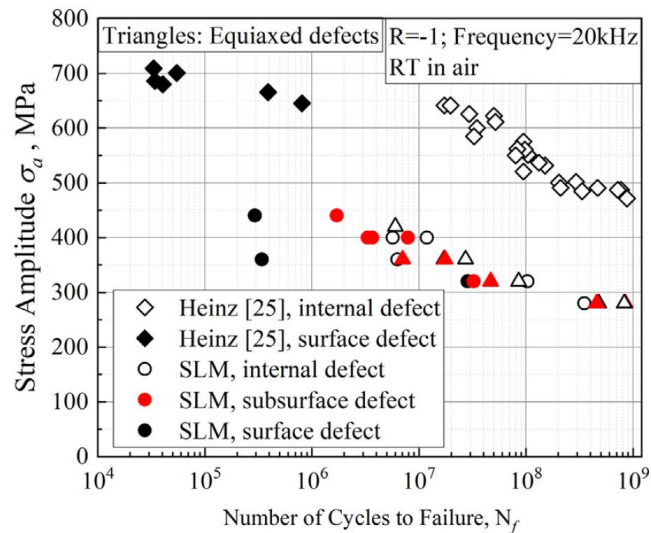
Qian, in <sup>60</sup>, studies the VHCF response of SLM Ti6Al4V samples built with different orientations. A Lasur GF20-TC ultrasonic testing system (loading frequency of 20 kHz) is used for fully reversed tension compression VHCF tests. Raw cylinders are manufactured in three directions with respect to the building plate, namely 0°, 45° and 90°. Afterwards, all the specimens are machined to the final hourglass geometry (gage diameter 3.5 mm) for VHCF tests. The typical lamellar microstructure of Ti-6Al-4V processed by SLM is obtained. Figure 19a shows the S-N plot of different batches. As it can be noticed, specimens built with 0° building orientation are characterized by the best VHCF response, whereas the VHCF response of specimens built at 90° is the worst. The slope of the three curves is similar and no evident stepwise type of S-N data is present. The crack initiates at internal LOF defects or pores caused by the SLM process both in HCF and VHCF regime. The effect of different building orientations is more evident in the HCF regime. The difference in fatigue strength is due to the presence of larger defects in the 90° batch. The median value of the VHCF strength at  $10^9$  cycles is 217 MPa for the 0° batch, followed by 210 MPa for the 45° batch and finally 155 MPa for the 90° batch, with a reduction of 40% from 0° to 90°. In Figure 19b the Authors analyse the VHCF strength decrement at  $10^9$  cycles with respect to the building orientation angle. A quadratic model is used to fit the experimental data.



**Figure 19** – VHCF response from Qian<sup>60</sup>: a) Stress-life (S-N) plot and b) effect of building orientation on fatigue strength at  $10^9$  cycles.

In Liu<sup>61</sup>, the VHCF response and the defect population of as-built SLM Ti6Al4V specimens is investigated. The specimens are produced with a laser power of 500 W and BPH at 60°C. The alloy powder is composed of (wt-%) 5.99 Al, 3.49 V and bal. and has a mean diameter of around 33  $\mu\text{m}$ . Fully reversed tension compression VHCF tests at 20 kHz are carried out with an USF-2000 (Shimadzu) ultrasonic testing system. Raw cylinders are at first produced in vertical direction and afterwards machined and polished to the hourglass geometry (gage diameter 3 mm). The microstructure is mainly composed of acicular  $\alpha$  and  $\alpha'$  lamellae and columnar  $\beta$ -phase along building axis. Figure 20 shows the fatigue response (S-N plot) of the specimens. As it can be noticed, an evident VHCF limit is not observable, whereas a steady decrease in strength with increasing lifetime to failure is present. The VHCF strength is assessed to be around 285 MPa at  $10^9$  cycles. For more, VHCF failures mainly originate from sub-superficial or internal LOF defects due to local process instabilities. Compared to the traditionally manufactured alloy in Heinz<sup>62</sup>, the VHCF is almost halved due to these internal SLM-induced defects. In the VHCF regime, the defect shape is usually regular and spherical.





**Figure 20** – Stress-life (S-N) plot of experimental data from Liu <sup>61</sup>.

Note: Heinz [25] in the legend refers to traditionally manufactured Ti6Al4V alloy. In the present paper the reference is <sup>62</sup>.

In Liu <sup>63</sup> the EBM induced anisotropy and the related VHCF response of Ti6Al4V are investigated. A Q10 system by Arcam company is employed to manufacture the samples. The building chamber is heated at 587°C and it is kept under vacuum at  $7.8 \cdot 10^{-3}$  mbar. Two different batches are built, named in the paper *xoy* and *xoz*. Raw cuboids *xoy* and *xoz*, produced through EBM in horizontal and in vertical directions, respectively, are machined and polished to the tested hourglass geometry (gage diameter 3 mm). The surfaces parallel to the vertical direction present a very high roughness, about  $R_a$  equal to 35  $\mu\text{m}$ , due to the large presence of non-melt particles partly sintered to the surface. Due to the high temperature in the chamber the as-built structure with fine acicular  $\alpha$ -phase and  $\alpha'$ -martensite with small amount of  $\beta$ -phase is transformed in a coarser duplex phase  $\alpha + \beta$ . Elongated columnar grains of  $\beta$ -phase are observable along the building direction, improving the ductility in this direction. However, these microstructural changes are not effective for top-layers that present a microstructure closer to that as-built. Indeed, these layers spend a smaller time in the heated chamber before the process end. Figure 21 shows the S-N plot of the *xoy* batch, i.e., horizontally built. A transition stress between HCF and VHCF failures at about 450 MPa is observable in the S-N plot and the graphically estimated VHCF strength up to  $10^9$  cycles is close to 260 MPa. At number of cycles smaller than  $10^7$ , the critical defects are superficial, whereas internal pores are the responsible for the crack initiation in the VHCF region. In this case, the porosity is not caused by shield gas inclusion, because the EBM process is carried out under high vacuum. The pores are instead produced by the Al vaporisation present in the alloy powder during the process.

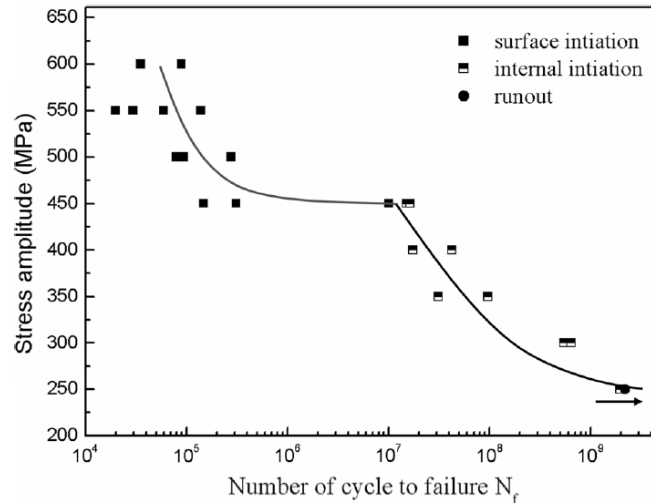


Figure 21 – Stress-life (S-N) plot of experimental data from Liu <sup>63</sup>.

### 3.2 VHCF strength of Ti6Al4V: discussion

Table 5 lists the literature results on the VHCF response of AM Ti6Al4V alloys.

Ti6Al4V					
Specimen type	Process	Building orient. <sup>1</sup>	Mechanical post-process	Thermal post-process	VHCF strength [MPa]
H <sup>52</sup>	SLM	90°	mac + pol	SR 650°C- 3h	205
H <sup>52</sup>	SLM	90°	mac + pol	HIP 920°C - 100 MPa -2h	483
H <sup>53</sup>	SLM (BPH 200°C)	90°	mac + pol	SR 800°C- 3h	240
H <sup>53</sup>	SLM (BPH 200°C)	90°	mac + pol	HIP 920°C - 100 MPa -2h	435
H <sup>53</sup>	EBM	0°	mac + pol	\	275
G <sup>55</sup>	SLM	90°	pol	SR 850°C- 1h	150
H <sup>57</sup>	SLM (OCS)	90°	pol	SR 600°C- 2h	215
H <sup>57</sup>	SLM (WCS)	90°	pol	SR 600°C- 2h	95
H <sup>60</sup>	SLM	0°	mac	\	217
H <sup>60</sup>	SLM	45°	mac	\	210
H <sup>60</sup>	SLM	90°	mac	\	155
H <sup>61</sup>	SLM (BPH 60°C)	90°	mac + pol	\	285
H <sup>63</sup>	EBM (BPH 587°C)	0°	mac + pol	\	260

Table 5 – VHCF strength of AM Ti6Al4V depending on process setup

*Note: mac – final machining, pol – final manual polishing, OCS– optimal case scenario, i.e., optimised choice of process parameters, WCS – worst case scenario,<sup>1</sup> – angle between building plate and specimen axis, i.e., 90° vertical and 0° horizontal, H – Hourglasses, G-Gaussian.*

In two works, the influence of the HIP treatment is investigated, proving that it has a positive influence on the VHCF response. Indeed, the HIP procedure homogenises the structure and closes or removes most of the process-induced defects. The graphically estimated VHCF strength up to  $10^9$  cycles of specimens subjected to HIP process is within the range 400 - 500 MPa and it is close to that of traditionally manufactured Ti alloy. As-built EBM samples and SLM + SR treated samples share almost the same microstructure and defect population. As a matter of fact, the SR effect is partly included in the EBM process where the building chamber is heated up to 500 - 600 °C. Nevertheless, the EBM process induces higher anisotropy along the building direction due to different time of exposure to high temperature of melt layers. Anyway, the VHCF strength of SLM + SR and EBM is within the range of 150 - 285 MPa, almost half of HIPed specimens. The most critical defects are mainly internal LOFs and pores caused by process instabilities. Indeed, a bad combination of process parameters can increase the defect population and lower of about 40% of the VHCF strength. Similarly, the building direction can reduce the VHCF strength by about 40% from horizontally to vertically built specimens. The SR treatment and machining post-process increase the VHCF strength, but it seems not to be that effective as for the AISI10Mg alloy. The use of Gaussian specimens leads to a more conservative estimation of VHCF strength due to larger employed risk volume. Overall, the most effective way to enhance the VHCF strength of Ti6Al4V is by means of HIP post process<sup>53</sup>, together with a proper selection of process parameters, as investigated in<sup>57</sup>. The optimal SLM process parameters combination results to be laser power of 160 W, hatching distance of 0.07 mm, layer thickness of 30 µm and scan speed of 1000 mm/s.

### **3.3 VHCF strength of Ti6Al4V: summary**

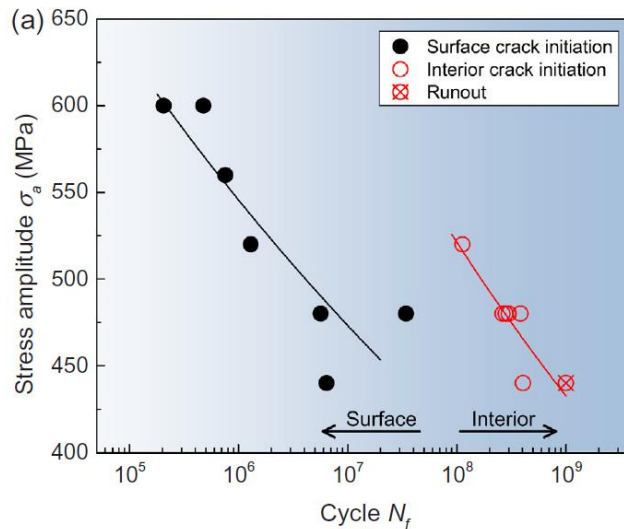
For the Ti6Al4V specimens, the influence of the production processes (SLM and EBM), of heat treatments and of the HIP process is mainly investigated:

- HIP enhances the VHCF strength, becoming comparable to that of traditionally manufactured Ti alloy.
- As-built EBM samples and SLM + SR treated samples share almost the same VHCF strength.
- EBM process induces higher anisotropy along the building direction.
- The VHCF strength of vertically built specimens is in average 40% lower than that of horizontally built specimens.

### **4.1 VHCF response of Inconel 718: experimental results**

In Yang<sup>64</sup>, the VHCF response of SLM Inconel 718 in the as-built condition is investigated. An SLM system M290 by EOS GmbH company with laser power of 285 W is employed, for manufacturing, in vertical direction, raw cylinders, which are thereafter machined and polished to the final tested hourglass geometry (gage diameter 3 mm). Fully reversed tension compression VHCF tests are carried out with an ultrasonic fatigue testing system, USF-2000 by Shimadzu, at a loading frequency of 20 kHz. The microstructure is characterised by a fine dendritic structure with Dendritic Arm Spacing (DAS) of about 0.8 µm, elongated in the building direction. The interdendritic regions are mainly constituted by a large amount of granular precipitates composed of Laves phases. Figure 22 shows the S-N plot of the experimental data. As it can be noticed, a clear fatigue limit does not exist as in traditional manufactured Inconel 718<sup>65</sup>. Two main slopes are clearly visible, discriminating between two failure modes: failures from surface defects in the HCF region (black curve on the left) and failures from internal defects in the VHCF region (red curve on the right). The graphically estimated VHCF strength at  $10^9$  cycle is approximately 432 MPa. Three main types of defects are observed through SEM analysis, namely pores, LOFs and crystallographic facets. However, LOF defects are rare and mainly formed due to bad process parameters combination, according to the Authors. On the other hand,

pores are quite frequent and together with crystallographic facets simultaneously cause the crack initiation in most of the cases.



**Figure 22** – Stress-life (S-N) plot of experimental data from Wang <sup>64</sup>.

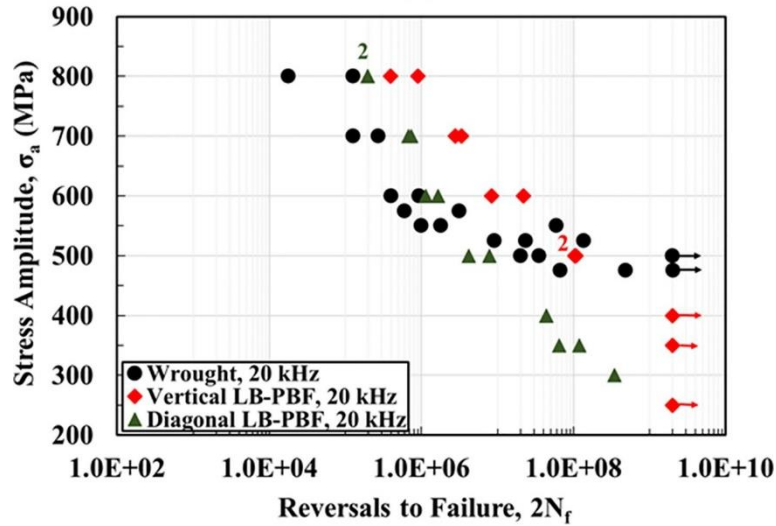
The VHCF response of wrought and SLM heat treated Inconel 718 is investigated in Muhammad <sup>66</sup>. The study focuses on the influence of test frequency and building orientation. An SLM system M290 by EOS company with laser power of 285 W is employed. Hourglass specimens are manufactured vertically and at 45° with respect to the base plate, with material allowance for the subsequent machining process. Afterwards, the as-built specimens are heat treated for 10 minutes at 954°C, then for 8 hours at 718°C and finally for 10 hours at 621°C. Before testing, the specimens are finally machined and polished (gage diameter of 3 mm). A second batch of wrought specimens is tested: in this case, the same hourglass geometry is obtained by machining round bars processed and heat treated according to SAE ANS2774E. Fully reversed tension compression VHCF tests are carried out with an ultrasonic fatigue testing system, USF-2000 by Shimadzu, at a loading frequency of 20 kHz. Figure 23 shows the experimental data of the three tested batches, namely vertically built (V-SLM), diagonally built (D-SLM) and wrought (W). The SLM specimens are characterised by a fatigue response at high stress levels (in the HCF region) better than that of W specimens. At lower stress levels (in the VHCF region), the difference tends to diminish and the VHCF strength of V-SLM specimens is larger than that of D-SLM specimens. The difference between the fatigue strength of W and SLM specimens is due to the large static strength and hardness of SLM specimens. The different fatigue response of the two tested SLM batches (i.e., manufactured with different building direction) can be explained by considering the microstructures and crack initiation sites. In both the SLM microstructures, a dendritic structure elongated in the building direction within coarse columnar grains, similarly to the as-built condition found in <sup>64</sup>, is present. The crack originates from persistent slip bands (PSBs) within the elongated grains oriented at around 45° with respect to the loading axis. The probability of slip planes in that direction for D-SLM specimens is larger, being detrimental for the fatigue life. Interestingly, even if pores are present (35 pores per mm<sup>3</sup>, mostly with diameter < 15 µm) the critical defects are PSBs. The small defect density could be also due to the small risk volume of the tested hourglass specimens. The graphically estimated mean VHCF strength at 10<sup>9</sup> cycles for V-SLM and D-SLM is approximately 334 MPa and 245 MPa, respectively.

The influence of the loading frequency on the tested Inconel 718 is also verified in the paper, showing that ultrasonic testing can overestimate the fatigue strength of the material. Indeed, in ultrasonic tests the specimen is considered to behave following the linear elastic model. However, during the test, plasticity phenomena can appear, and the estimated resonance frequency may not be accurate anymore, according

to the explanation of the Authors. The Authors therefore proposed the following correction factor for test results obtained with the ultrasonic testing machines:

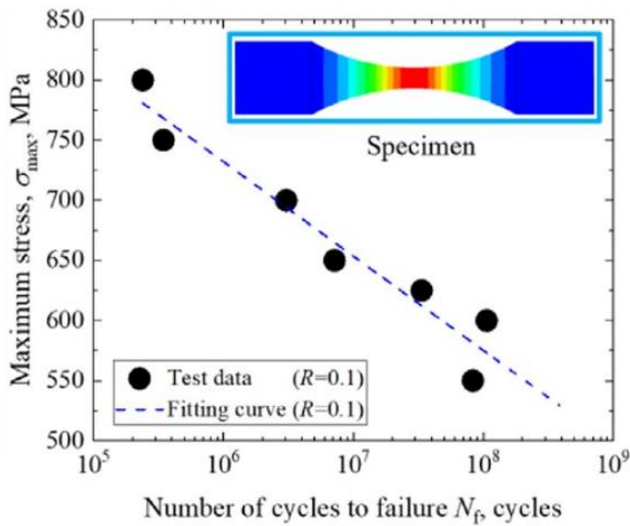
$$\sigma_{UF}^C = 0.85 \times \sigma_{UF}, \quad (2)$$

where  $\sigma_{UF}^C$  is the corrected stress amplitude and  $\sigma_{UF}$  is the measured one.

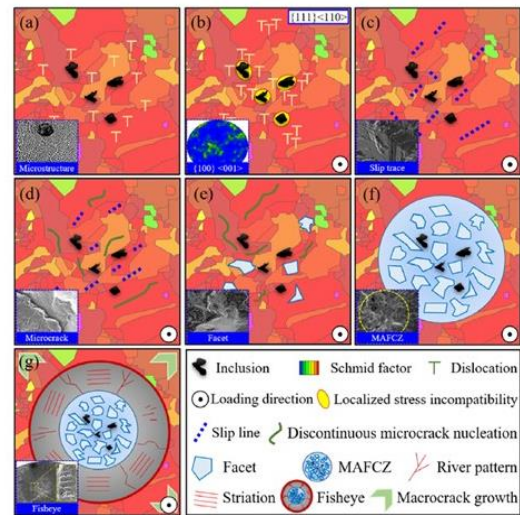


**Figure 23** – Stress-life (S-N) plot of experimental data from Muhammad <sup>66</sup>.

The subsurface cracking mode with fisheye pattern of SLM Ni-based superalloy (Inconel 718) in the VHCF regime is investigated by Li <sup>67</sup>. An SLM system M280 by EOS company with laser power of 280 W is employed. The Ni-based superalloy powder is composed of (wt-%) 52.16 Ni, 19.34 Cr, 5.07 Nb, 2.96 Mo, 0.89 Ti, 0.51 Al, 0.049 Co, 0.0041 C and 0.0021 Mn with a maximum particle size below 40  $\mu\text{m}$ . Raw cylinders are manufactured vertically without further heat treatment. Afterwards, the cylinders are machined and grounded to the final hourglass testing shape (gage diameter 4 mm). The VHCF strength is evaluated with axial loading fatigue tests at stress ratio  $R$  equal to 0.1 by using an electromagnetic resonant testing machine at a frequency of 100 Hz. The as-built microstructure is similar to that found in <sup>64</sup>, characterized by a very fine cellular structure in the plane parallel to the building plate. Figure 24a shows the S-N plot up to  $10^8$  cycles together with a linear fitting curve. The graphically estimated VHCF strength at  $10^9$  cycles is around 500 MPa. However, the experimental data do not show a clear transition between the HCF and VHCF regime, with a monotonic decreasing trend from the HCF to the VHCF region. VHCF failures mainly originate from internal non-metallic inclusions or pores, with a subsequent progressive crack propagation. Figure 24b shows precisely how this mechanism takes place in the SLM Ni-based superalloy.



a)



b)

**Figure 24** – a) Stress-life (S-N) plot of experimental data and b) fisheye formation process from internal defects<sup>67</sup>.

Initially, the crack initiates from grain slip assisted by inclusion along the maximum shear stress plane. Then, the microcrack grows and induces the formation of facets. Finally, this process continues forming the well know circular fisheye pattern till the specimen collapses.

#### 4.2 VHCF response of Inconel 718: discussion

AM produced Inconel 718 exhibits a very good fatigue response. In the VHCF response, some peculiar aspects can be inferred from the literature data. First of all, differently from tests on traditionally built Inconel 718 specimens, a clear fatigue limit is not present. Secondly, the critical defects are always internal and mainly non-metallic inclusions, pores and LOFs together with facets and crystallographic discontinuities. Finally, the heat treatment seems not to improve remarkably the VHCF response. In contrast with other AM materials, such as Ti6Al4V<sup>60</sup> and AlSi10Mg<sup>42</sup>, the vertically built Inconel 718 specimens do not present a decrease in fatigue performance. On the contrary, the vertical build direction guarantees a good VHCF strength at  $10^9$  cycle while it is highly reduced in diagonally built specimens. Moreover, in<sup>66</sup> two aspects are highlighted by the Authors. Firstly, the need of enlarging the tested risk-volume, since the hourglass shape does not permit to properly assess the defects population and, consequently, to reliably estimate the VHCF response (as already noticed for AlSi10Mg in<sup>44</sup>). Secondly, when ultrasonic fatigue tests are carried out possible strain rate effects and plastic phenomena should be verified and considered, in order to avoid an overestimation of the VHCF response.

Table 6 summarizes the experimental results of VHCF tests on the Inconel 718 alloy: the main process parameters and the VHCF strength are reported.

<b>Inconel 718</b> (SLM process, final machining and polishing)					
Specimen type	Building orientation <sup>1</sup>	Thermal post-process	Stress ratio $R$	Runout	VHCF strength [MPa]
Hourglass <sup>64</sup>	90°	\	-1	10 <sup>9</sup>	432
Hourglass <sup>66</sup>	90°	1) 954°C - 10min 2) 718°C - 8h 3) 621°C - 10h	-1	10 <sup>9</sup>	334
Hourglass <sup>66</sup>	45°	1) 954°C - 10min 2) 718°C - 8h 3) 621°C - 10h	-1	10 <sup>9</sup>	246
Hourglass <sup>67</sup>	90°	\	0.1	10 <sup>8</sup>	500

**Table 6** – VHCF strength of AM IN718 depending on process setup

*Note:*<sup>1</sup> – angle between building plate and specimen axis, i.e., 90° vertical and 0° horizontal.

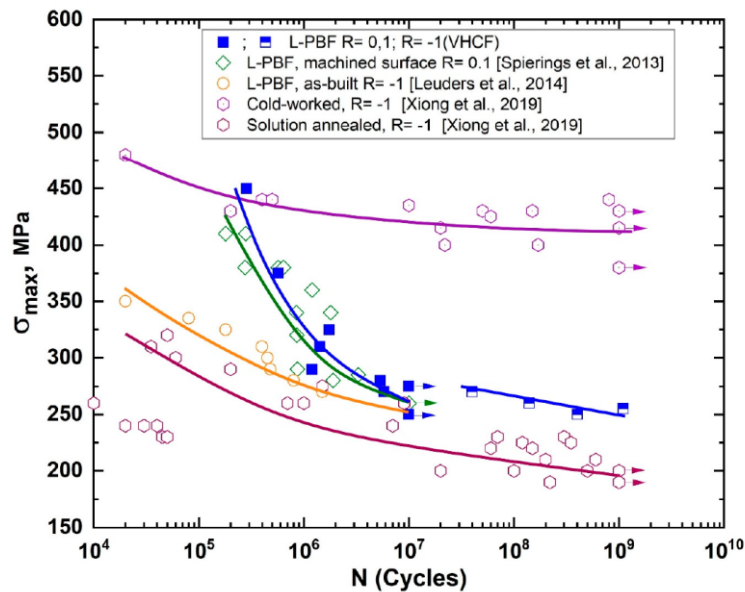
### 4.3 VHCF response of Inconel 718: summary

For the specimens produced with the Inconel 718 alloy, the influence of defects and building orientation is mainly investigated:

- Differently from traditionally built specimens, no clear VHCF limit is present, whereas, a monotonically decreasing behaviour takes place.
- The critical defects are mainly facets and crystallographic discontinuities.
- VHCF strength is highly reduced in specimens built at 45° built compared to vertically built ones.

### 5.1 VHCF response of 316L stainless steel: experimental results and discussion

The VHCF response of SLM stainless steel 316L is studied in Voloskov <sup>68</sup>. An SLM system TruPrint1000 by Trumpf company with laser power of 113 W is used. The diameter of the 316L gas atomised powders is in the range 20 - 50 µm. Raw cylinders are manufactured vertically and thereafter machined and ground to the final hourglass testing shape (gage diameter 3 mm). Fully reversed tension compression VHCF tests are carried out with an ultrasonic fatigue testing system, USF-2000 by Shimadzu, at a loading frequency of 20 kHz. Figure 25 shows the S-N plot of the experimental data together with literature results on cold-worked and heat-treated stainless steel. The VHCF response of SLM 316L is better than that of the annealed batch, but the half of that of cold-worked specimens.



**Figure 25** – Stress-life (S-N) plot of experimental data from Voloskov <sup>68</sup>

*Note: The references to Spierings 2013, Leuders 2014 and Xiong 2019 in the plot are in this paper numbered <sup>69–71</sup> respectively*

The VHCF strength at  $10^9$  cycles is around 250 MPa and the trend is monotonically decreasing without an evident fatigue limit in the VHCF region. The critical defects are internal non-metallic inclusions and LOF defects, which originate a fish-eye morphology on the fracture surface.

## 5.2 VHCF response of 316L stainless steel: summary

One work is published on the VHCF response of SLM 316L specimens, comparing the VHCF response of AM specimens with that of traditionally built specimens:

- The VHCF response of SLM 316L is better than that of traditional annealed stainless steel, but the half of that of cold-worked specimens.
- The VHCF response presents a monotonically decreasing trend without an evident fatigue limit.

## Conclusions

In the present paper, a literature review on the VHCF response of specimens produced through Additive Manufacturing (AM) processes is provided. To guarantee a safe design of the components produced through AM, the fatigue response at very high number of cycles should be also reliably assessed. The growing importance of the research on the VHCF response of parts produced through AM processes is attested by the increment of papers published on this subject in the last two years (2019-2020), being about the 60% of the total. However, more efforts should be made to understand the mechanisms of crack initiation and the influence of process parameters and post treatments on the VHCF response of AM parts.

At present, SLM and EBM production processes are used to manufacture specimens for VHCF tests. EBM samples are tested in the VHCF regime only for Ti6Al4V. The results show that the EBM process produces the final part which behaves similarly to heat-treated SLM ones with increased building direction anisotropy. Applied laser power, printing speed, hatching distance and the other process parameters are studied deeper only for Ti6Al4V. However, these parameters and their correct combination are almost always suggested by the machine producers. For most of the tested alloys, the default process setup is employed. Indeed, this



setup is often guaranteeing the lower defect density and relative porosity. As for the powder size, for all the studied alloys the mean diameter of the atomised powder should be around 50  $\mu\text{m}$ . Inferior values, induced more gas inclusions and larger defects.

For what concerns the experimental tests, they are mainly carried out with Ti6Al4V and AlSi10Mg specimens, whereas a smaller number of results are available for AlSi12, IN718 and 316L specimens. Experimental tests are mainly carried out by using ultrasonic testing machines working at a loading frequency in the range 19 kHz - 21 kHz, but experimental tests with traditionally testing machines are also carried out to assess the influence of the loading frequency. The process-induced defects, such as LOFs, pores, ILBs, control the VHCF response for every studied alloy. If defects are not present or are removed with post-treatments, the microstructural defects are responsible for the fatigue crack initiation. It has been found that the most defect-dense zone is close to the as-built surface. Therefore, the removal through machining of this zone would eliminate the largest defects and would surely have a positive effect on the VHCF response. Anyway, this operation may induce an incorrect evaluation of the fatigue limit of real parts. For example, if a part of an AM component would not be reached by machining tools, a consistent part of the surface would remain untouched. This phenomenon may be frequent since AM components are often characterised by convoluted and complex geometries. The unmachined surface would induce a premature fatigue failure compared with the data from machined samples. Hence, the final machining and polishing must be considered as crucial operations in the evaluation of the VHCF strength. Similarly, the properly estimation of the defect population and consequently of the VHCF response, requires an enlargement of the tested risk-volume to avoid underestimation of the critical defect size. Otherwise, a larger number of experimental tests should be carried out.

All considered, differently from VHCF results on specimens produced through traditional processes, the fatigue failures in VHCF parts do not originate from internal defects with a fish-eye morphology for all the tested AM materials, depending on the location of the initial defect and on the density of defects within the specimen section. To conclude, according to the analyses carried out in the paper and based on literature results, the effects of heat process parameters and post treatments are extensively investigated for Ti6Al4V and AlSi10Mg specimens. However, there is the need to fill the gap of knowledge regarding the effect of process parameters and post treatments for other commercially available materials to be produced with the AM processes. Indeed, due to the large number of process parameters involved in the specimen production, the extension of the main findings for a specific material could not be valid for other materials, making it necessary an accurate experimental validation. The research should also focus on the mechanisms of crack initiation in parts produced through AM, to provide reliable design procedures and to verify if the traditional design methodologies can be applied even for AM components.

### **Declaration of competing Interest**

The authors declared that there is no conflict of interest.

### **Statement of contribution**

R. Caivano retrieved the literature data, analysed the data, and wrote the paper. A. Tridello helps analysing the data, made additions to the paper and revised the paper. G. Chiandussi and G. Qian revised the paper. D. S Paolino revised the manuscript and supervised the work. F. Berto revised the manuscript and supervised the work.

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## Research highlights

1. The very high cycle fatigue response of additively manufactured alloys is comprehensively revised.
2. The experimental results of tests on Ti6Al4V, AlSi10Mg, AlSi7Mg, AlSi12, IN718 and 316L specimens are analysed.
3. The key factors affecting the very high cycle fatigue response of AM parts are compared and analysed.
4. The literature results are collected, summarised, and clearly organised to be straightforwardly accessible