

Fatigue behaviour of Ti6Al4V alloy components manufactured by SLM subjected to HIP and residual stress relief

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ABSTRACT

Fatigue behaviour of HIPed and stress relieved Ti6Al4V alloy specimen's produced by SLM was analysed ($R_\epsilon = -1$). The HIP process caused a microstructural transformation decreasing the hardness and monotonic properties that not allowed fatigue strength increase. A bilinear behaviour in the elastic strain-fatigue life curve was observed, because of Young's modulus decrease during the cyclic elasto-plastic tests consequence of subgrains formation. The Smith-Watson-Topper and total strain energy density models showed a good concordance between predicted and experimental fatigue lives in notched samples.

Keywords: Ti6Al4V alloy, selective laser melting, strain-life fatigue curves, fatigue life prediction, cycle fatigue properties, S-N curves.

NOMENCLATURE

R_ϵ	Strain ratio
HIP	Hot Isostatic Pressing
SLM	Selective Laser Melting
LCF	Low Cycle Fatigue
R_a	Roughness Average
ϵ_a	Strain amplitude
N_f	Number of cycles until failure
S-N	Curves: Stress vs Number of cycles until failure
σ - ϵ	Curves: Stress vs strain
wt. %	Weight percentage
H_2NO_3	Nitric acid
HF	Hydrofluoric acid
H_2O	Water
$\Delta\epsilon$	Strain range

k'	Cyclic strength coefficient
n'	Cyclic strain hardening exponent
σ_a	Stress amplitude
ϵ_a^*	Satrin amplitude of master curve
σ_a^*	Stress amplitude of master curve
k^*	Cyclic strength coefficient of master curve
n^*	Cyclic strain hardening exponent of master curve
ϵ_{a_p}	Plastic strain amplitude
ϵ_f'	Fatigue ductility coefficient
c	Fatigue ductility exponent
ϵ_{a_e}	Elastic strain amplitude
σ_f'	Fatigue strength coefficient
b	Fatigue strength exponent
E	Young's modulus
kp	Material constant of energy density
αp	Material exponent of energy density
ΔW^{e+l+l}	Tensile elastic energy range at the material fatigue limit
ΔW_T	Total strain energy density range
σ_{max}	Maximum stress
k_t	Static concentration factor
σ_n	Nominal stress
k_f	Fatigue stress concentration factor

σ_{fu} Fatigue strength for smooth specimen's

σ_{fe} Fatigue strength for notched specimen's

q Notch sensibility

σ_n Nominal stress

$\Delta \sigma_{loc}$ Maximum local stress ranges

$\Delta \varepsilon_{loc}$ Maximum local strain ranges

FEM Finite Element Method

SWT Smith-Watson-Topper model

TEM Transmission electron microscope

TKD Transmission kikuchi diffraction

α Microstructural phase of titanium

β Microstructural phase of titanium alloys

1. INTRODUCTION

The Ti6Al4V alloy have been widely applied in both aerospace and biomedical industries due to its high strength, high thermal stability, and good corrosion resistance¹⁻³. However, the machinability of titanium alloys manufactured by the traditional processing technologies presents some drawbacks, for example: sticking, blade wear-out, waste of raw material and slow manufacture times. The SLM process (Selective Laser Melting) presents great advantages such as, reduction of production steps, a high level of flexibility, a high material use efficiency, a near net shape production and also avoids the problems described previously.

The SLM process allows the manufacture of complex parts with excellent precision through the successive powder layers deposition and laser melting, in a process understood as metal 3D print. The fatigue behaviour of parts produced by SLM is strongly affected by porosity defects, surface roughness and residual stresses, all resulting from the SLM process. The influence of surface roughness and internal defects, such as insufficient fusion and surface porosity, on fatigue performance of the Ti6Al4V alloy produced by SLM have been investigated by Wycisk et al.⁴, Edwards et al.⁵, Leuders et al.⁶ and Cao et al.⁷, showing a detrimental influence of these factors on the fatigue strength. Typically, the SLM parts are submitted to a post heat treatment in order to reduce/eliminate the residual stresses induced by to the intense temperature gradients produced during the SLM process. Xuan and Nastac⁸ obtained an increase in the fatigue strength of Ti6Al4V parts produced by SLM and submitted to a post-stress relief treatment, in comparison with the as-build parts.

Other authors, as Zhang et al.⁹ and Atkinson and Davies¹⁰, proposed the utilization of HIP (Hot Isostatic Pressing) treatment to reduce the porosity defects, surface roughness and introduction of favourable residual stresses, leading to an improvement of the fatigue performance. The HIP treatment is a process in which the material is subjected to both controlled temperature and isostatic gas pressure. The theory behind this process is to create a high enough hydrostatic pressure in the vessel so that any internal porosity (not surface connected) will collapse¹¹. Chastand et al.¹² and Yu et al.¹³ applied the HIP treatment to Ti6Al4V specimen's produced by SLM, using 920°C of temperature and 120 MPa of pressure. The HIP process led to a porosity reduction that allowed a 10-15% increase in the fatigue resistance.

Wang et al.¹⁴ and Carrion et al.¹⁵ studied the fatigue behaviour of the Ti6Al4V alloy produced by conventional processes, by carrying out strain controlled fatigue tests. They obtained a cyclical softening when this alloy is subjected to strain amplitudes with a non-zero plastic component. Zhang et al.¹⁶ compared the Low Cycle Fatigue (LCF) behaviour of SLM Ti6Al4V and of wrought Ti6Al4V alloys, concluding that the wrought material showed a better fatigue performance for high strain levels due to the detrimental effect of the SLM defects, while, at low strain amplitudes, the SLM material presented a better fatigue performance. These authors also observed a cyclic softening behaviour of the materials produced by both processes.

Xu et al.¹⁷ studied the LCF behaviour of the conventional processed Ti6Al4V alloy after being heat-treated at 925°C and subsequently stabilized at 700°C. They found that the fatigue life is highly influenced by the total strain amplitude, presenting a bilinear behaviour, i.e., a discontinuity with two regions with different slopes on the Coffin-Manson curve. For total strain amplitudes above 1-1.2%, the slope of the trendline is twice less than for total strain amplitudes below those values. This indicates that this alloy can have different cyclic deformation micromechanisms for different strain amplitudes. This bilinear behaviour has already been found in other metallic materials such as, nickel superalloys (Ye et al.¹⁸ and Gopinath et al.¹⁹), aluminium alloys (Branco et al.²⁰ and Prasad et al.²¹) and titanium alloys (Gouthama and Singh,²²).

In this work, fatigue tests were performed with Ti6Al4V alloy samples produced by SLM with circular cross-section for a strain ratio $R_\epsilon = -1$. In addition, several complementary analyses were also performed, namely, metallography, hardness and tensile tests. Complementary analyses were performed in order to understand the variation in fatigue behaviour after the application of two different treatments: the stress relief treatment and the HIP treatment. Besides, fatigue life predictions were performed based on different prediction models for notched specimen's and using the fatigue properties obtained in the experimental part of this study.

2. MATERIALS AND EXPERIMENTAL PROCEDURES

Fatigue test specimen's were produced using Ti6Al4V alloy powder and the SLM process *3D Systems*, model ProX DMP 320. The thickness of each deposited layer was 30 μm using an energy density of 57 J/mm³. All layers were deposited in a reverse mesh at 45° employing metal powder of Ti6Al4V (grade 23) Titanium alloy. The specimen's

geometry was chosen following the recommendations of the ASTM E606 standard²³. Fig. 1 provides the specimen's geometries used in this research, the loading direction and the build direction. Fig. 1a) presents the smooth specimen's and Fig. 1b) shows the specimen's with a semi-elliptical notch ($a=0.8$ mm and $b=0.5$ mm).

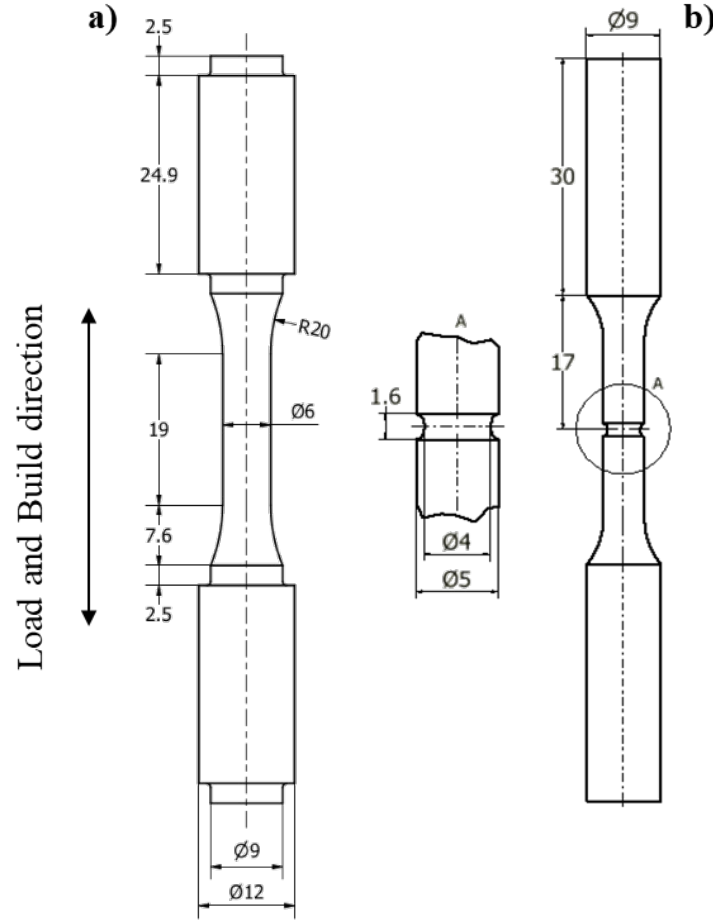


Figure 1. Specimen's geometries (mm). a) specimen's without notch, b) specimen's with semi-elliptical notch.

For the smooth specimen's, with the geometry illustrated in Fig. 1a), both HIP and residual stress relief treatments were applied, while for the notched specimen's (Fig. 1b)) only the residual stress relief treatment was applied after the machining of the notch. The heat treatment of residual stress relief was performed under gentle and controlled heating up to 670°C and keeping the temperature at 670°C±15°C for 5 hours and, lastly, a cooling in air until room temperature. The HIP process was carried out submitting the specimen's to a pressure of 100 MPa with a controlled heating up to 920 °C, followed by a period of constant temperature (2 hours) at 920 °C±15°C, finally, a cooling in air until room temperature. In order to reduce the surface roughness, the

specimen's were polished after the heat treatments obtaining a final roughness of $R_a=3.07\text{ }\mu\text{m}$. After treatment by each of the two processes, samples slices were cut from the cross section of the specimen's, and prepared considering the recommendations of ASTM E3²⁴ standard in order to study the changes in microstructure and hardness. Afterwards, Kroll's reagent (6% H_2NO_3 , 1% HF and 93% H_2O) were used to the etching process of the samples as suggested in ASTM E407²⁵ standard and subsequently observed and photographed in an optical microscope Leica DM 4000 M LED. As previously referred, the same samples were used to find the Vickers hardness according to the ASTM E384-11e1²⁶ standard, using a hardness tester Stuers Duramin 1 through 1 kg of test load and random indentation measurements.

The monotonic properties were obtained analysing tensile tests executed in an Instron mechanical tensile/compression testing machine (model 4206), at room temperature, with 2 mm/min of displacement rate using a 12.5 mm gauge length extensometer, Instron 2620-601, to record the elongation and the specimen's geometry depicted in Fig. 1a). The tensile tests were executed following the guidelines of ASTM B528-16²⁷ standard.

The fatigue tests were performed in a Dartec servo hydraulic machine, under a sinusoidal wave applying a strain-controlled conditions of $R_\epsilon = -1$ with a constant strain rate $d\epsilon/dt = 8 \times 10^{-3}\text{ s}^{-1}$ following the recommendations of ASTM E606²³ standard. The total strain amplitudes have been between 0.3-1.875%. The strain was continuously controlled and recorded with the same extensometer used in the tensile tests, clamping it directly to the gauge zone of the specimen. Fatigue results were plotted mainly as ϵ_a - $2N_f$, S-N and σ - ϵ curves, presenting the strain amplitudes (elastic, plastic and total) against the number of reversions to failure, stress amplitude against the number of cycles to failure and stress amplitude against the strain amplitude, respectively. The half-life criterion was used to analysis the cyclic stress-strain and strain-life responses of the materials in the two heat treatment conditions. The analysis of hysteresis loops at mid-life allowed to obtain the cyclical properties in the Ti6Al4V specimen's built by SLM. Also, fatigue life predictions based in local approaches were performed for stress relieved notched specimen's using the fatigue properties obtained in the tests.

3. RESULTS AND DISCUSSION

3.1. Metallographic analysis

Fig. 2 shows the microstructures obtained for each treatment. Fig. 2a) presents the microstructure of stress relieved specimen's while Fig. 2b) shows the microstructure found for the HIPed specimen's. Both images show darker zones corresponding to β phase grains and lighter areas corresponding to fine needles of the martensitic phase α (or α'). Greitemeier et al.²⁸ observed a similar morphology for the same material and manufacturing process, confirming a balance between the β and α phases in the stress relieved specimen's (Fig. 2a)). On the other hand, when comparing Fig. 2a) and 2b) can be concluded that the Fig. 2b) shows more dark areas (phase β) and less amount of the martensitic phase α (lighter areas) because the HIP process produced a microstructural transformation: phase α to phase β . Molaei et al.²⁹ also observed the same difference of microstructures after the applications of the two heat treatments. It was not possible to identify a denser structure in the HIPed specimen's when comparing Fig. 2a) and 2b).

3.2. Hardness analysis

Due to the microstructural transformation described previously, the average hardness value for the HIPed samples (350 HV1) was lower than average hardness value for the stress relieved samples (405 HV1), a decrease of 14 %. This is caused because the β phase is characterized by a lower hardness than the α phase.

3.3. Monotonic tests

The Table 1 shows the monotonic properties for each sample where can be seen that a loss of monotonic properties in samples treated by HIP due to hardness decrease described above. On the other hand, the HIPed samples showed a ductility increase around 33.5%.

Table 1. Summary of the mechanical monotonic properties.

SAMPLE	σ_{UTS} [MPa]	σ_{ys} [MPa]	ϵ_f [%]	ϵ [GPa]
Stress Relieved	1143	1107	19.7	126
HIPed	995	950	26.3	126
HIP vs Stress Relieved	-15%	-16.5%	+33.5%	-

The HIPed samples showed a greater amount of phase β provoked by the HIP treatment given that the maximum temperature achieves in the HIP process applied was 920 °C surpassing the temperature of the microstructural transformation, phase α to β phases,

which is 882 °C for pure titanium. The greater amount of phase β leads to a lower hardness and higher ductility.

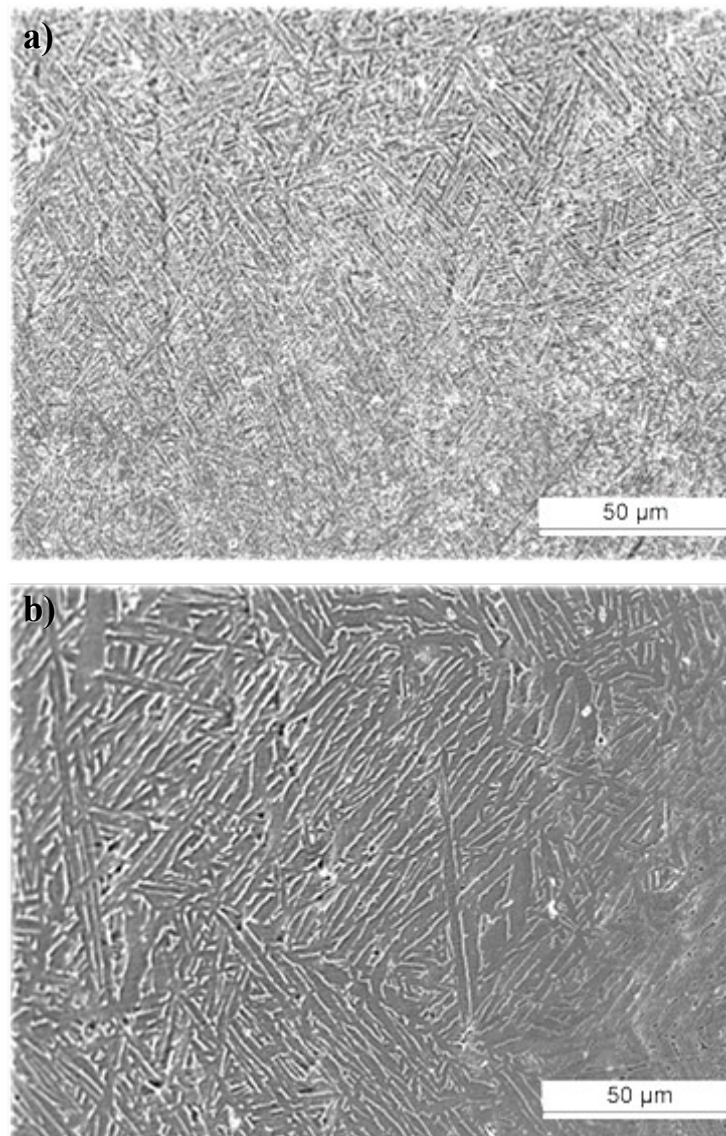


Figure 2. Optical metallographics of the specimen's. a) stress relieved and b) HIPed.

3.4. Fatigue tests results

The fatigue tests will be presented in order to understand the effect of both heat treatments on the cyclic properties of the Ti6Al4V specimen's built by SLM.

3.4.1. Hysteresis loops

Fig. 3a) and 3b) show exemplary hysteresis loops obtained in low cycle fatigue tests with a total strain amplitude of $\epsilon_a = 1.75\%$ for the stress relieved and HIPed treatments, respectively.

In both treatments, the stabilization of hysteresis loops occurred after 17% of the total life, with a more pronounced cyclic softening in the case of the specimen's subjected to the stress relief treatment, that is, the stress amplitude shows a greater reduction since the first cycle in comparison to stabilized cycles.

Fig. 3 shows that the Bauschinger effect occurred for both treatments, although it is more pronounced in the HIPed specimen's. In fact, there is an asymmetry in the hysteresis loops with the maximum tensile stress being lower than the maximum compressive stress. This asymmetry is greater in the case of the HIPed specimen's.

Can be concluded through hysteresis loops analysis that the HIPed samples showed greater ductility and greater strain energy corresponding to the greater area inside the hysteresis loops. These results agree with the microstructure alteration that occurs during the HIP process and its consequences in the preponderance of phase β with the properties already mentioned.

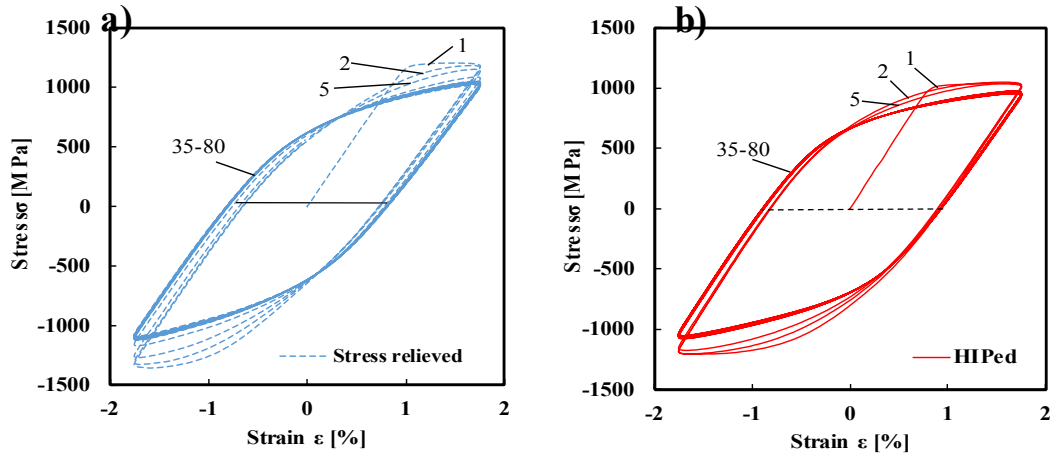


Figure 3. Hysteresis loops in elasto-plastic regime at $\epsilon_a = 1.75\%$ for both heat treatments: a) Stress relief; b) HIP.

3.4.2. Stress variation with fatigue life

Fig. 4a) and 4b) illustrate the variation of the stress ratio, $\sigma_{\max}/-\sigma_{\min}$, with the normalised fatigue life at various total strain ranges for stress relieved and HIPed specimen's, respectively. For values of total strain ranges below 1.5% (predominantly elastic regime), the Bauschinger effect did not occur because the values of tensile stress and compression stress remained equal throughout fatigue life (hysteresis loops symmetry), whereby the stress ratio remained equal to 1. Fig. 4a) and 4b) also show that the

Bauschinger effect occurred in both series (stress ratio below 1), being more pronounced in the early fatigue cycles. This phenomenon was greater in HIPed specimen's, where a higher difference is observed between the peak tension stresses and the peak compression stresses (Fig. 4b)).

For total strain ranges above 1.5% (elasto-plastic regime), it was found that the peak tensile and compression stresses undergo a gradual reduction in the first cycles of the fatigue tests in both heat treatments. Therefore, there is an initial cyclic softening followed by a maintenance period with practically constant stress values up to 90% of the total life. In the remaining 10% of the fatigue cycles there is a gradual drop in the stress peaks until rupture. Cyclic softening is more pronounced for specimen's subjected to the stress relief treatment. For values of total strain ranges below 1.5% (predominantly elastic regime), there is no cyclic softening for both treated series, since the peak stresses remain practically constant during the fatigue tests.

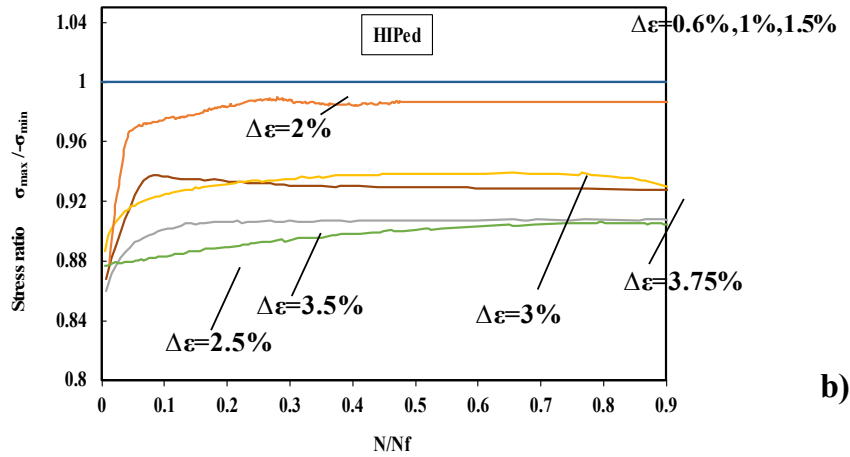
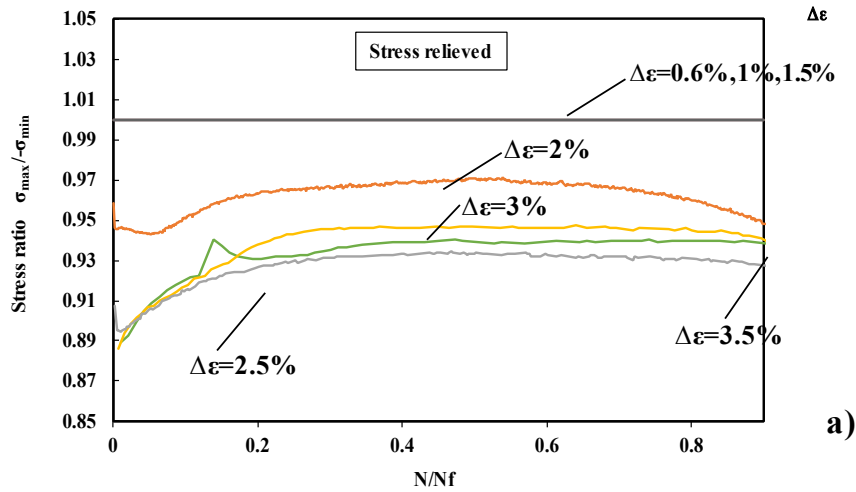


Figure 4. Variation of stress ratio with the normalised fatigue life at various total strain ranges. a) Stress relieved specimen's b) HIPed specimen's.

3.4.3. Cyclic stress-strain behaviour

The cyclic stress vs strain curve can be written in the form (Ramberg-Osgood cyclic stress strain relation³⁰):

$$\varepsilon_a = \frac{\sigma_a}{E} + \left(\frac{\sigma_a}{k'} \right)^{\frac{1}{n'}} \quad (1)$$

where k' and n' are, respectively, the cyclic strength coefficient and the cyclic strain hardening exponent, as summarized in Table 2. Fig. 5 shows the cyclic curves for both heat treatments.

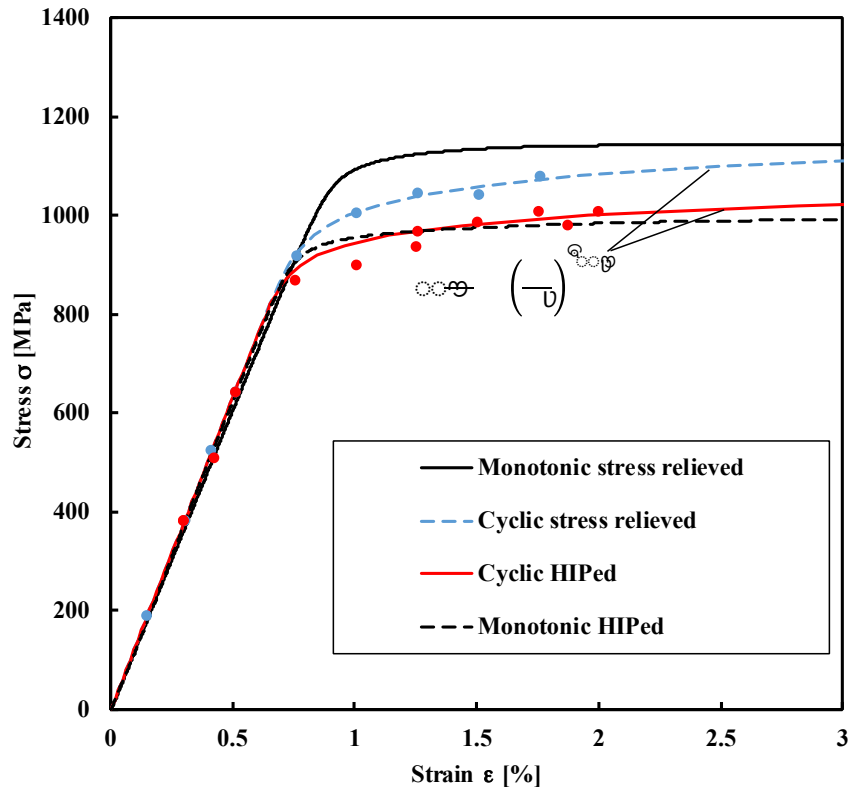


Figure 5. Cyclic stress vs strain curves.

In Fig. 5 it can be observed that the specimen's submitted to the stress relieved treatment undergo a cyclic softening once the experimental points of the cyclic curve are below the monotonic curve, while, the HIPed specimen's present an approximately stable behaviour once most of the cyclic stress-strain curve points are almost coincident with the monotonic curve. The hardened state of the stress relieved specimen's leads to

a cyclic softening when subjected to cyclic straining. On the other hand, the HIPed specimen's that are in an initial state of lower hardness, when subjected to cyclic straining did not undergo cyclic hardening or softening.

Table 2. Cyclic stress-strain curve properties.

k' [MPa]	n'	Stress relieved
1314	0.043	
k' [MPa]	n'	HIPed
1176	0.036	
k^* [MPa]	n^*	Stress relieved
2275	0.155	
k^* [MPa]	n^*	HIPed
1695	0.111	

3.4.4. Masing material behaviour

In Fig. 6 are plotted the cyclic Ramberg-Osgood equation and the mid-life hysteresis loops at different total strain ranges for both treatments. In both cases the cyclic curve and the superior branches at different strain amplitudes do not coincide in a single curve by this reason the material did not showed a Masing behaviour so having a non-Masing behaviour. The non-Masing effect is associated with the alteration in the linear region of the stable circuits³¹.

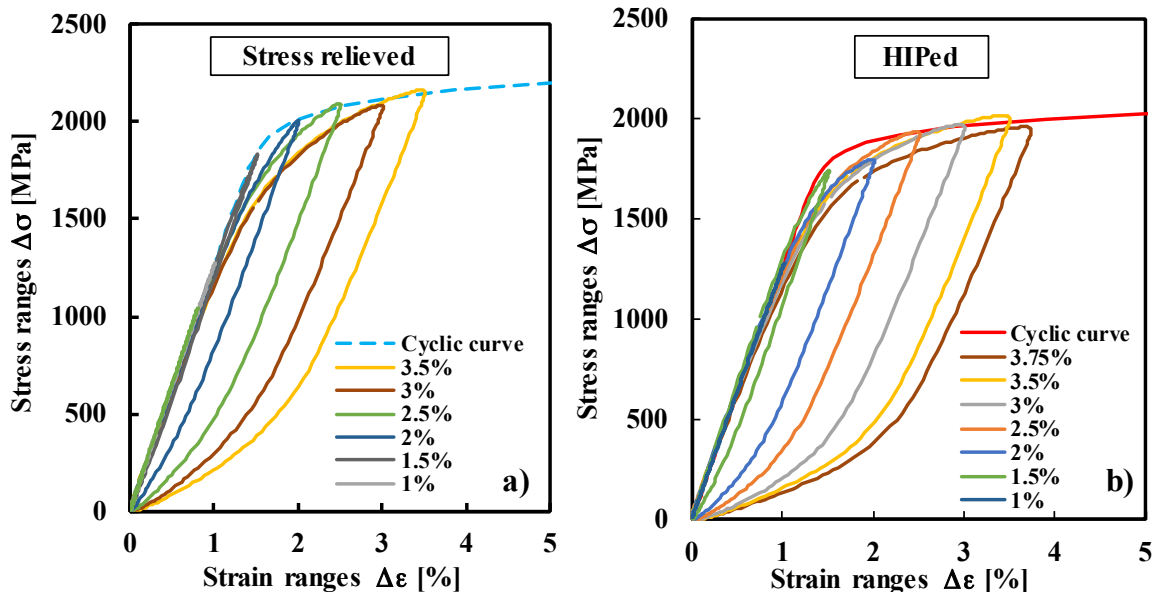


Figure 6. Cyclic Ramberg-Osgood model and mid-life hysteresis loops at different total strain ranges.

The divergence from Masing behaviour can be studied by the creation of a master curve, the master curves were found by matching the loading branch of the constant hysteresis loops with the corresponding minimum strain range.

The equation of the master curve can be expressed as Eq. 2.

$$\varepsilon_a^* = \frac{\sigma_a^*}{E} + \left(\frac{\sigma_a^*}{k^*} \right)^{\frac{1}{n^*}} \quad (2)$$

where asterisk (*) in superscript designates that the parameters correspond to the master curve equation. Table 2 summarizes the parameters k^* and n^* for each specimen.

3.4.5. Strain-life curves

The elastic component of the total strain amplitude, ε_{a_e} , can be related with the number of reversals to failure, N_f , through the Basquin formulation³² given by Eq. 3, where σ_f' is the fatigue strength coefficient, b is the fatigue strength exponent and E is the Young's modulus.

$$\varepsilon_{a_e} = \frac{\sigma_f'}{E} (2 N_f)^b \quad (3)$$

Fig. 7 shows the variation of the number of reversals to failure with the elastic component of the total strain amplitude (in log-log scale) for a) stress relieved specimen's and b) HIPed specimen's. For both treatments a bilinear behaviour was observed (two linear regions with different slopes) in the elastic strain component depending of the total strain amplitude. This phenomenon is due to the dissimilar behaviour of the elastic strain component in the predominantly elastic regime ($\varepsilon_a < 0.75\%$ for stress relieved samples and $\varepsilon_a < 0.68\%$ for HIPed samples) and in the elasto-plastic regime ($\varepsilon_a > 0.75\%$ for stress relieved specimen's and $\varepsilon_a > 0.68\%$ for HIPed specimen's). The bilinearity in the elastic component of the total strain amplitude can be attributed to the variation of the Young's modulus during the fatigue tests in the elastic-plastic regime. This material exhibited a progressive reduction of the Young's modulus when subjected to a total strain amplitude in the elasto-plastic regime ($\varepsilon_a = 1.5\%$), while, in the predominantly elastic regime ($\varepsilon_a = 0.5\%$), the Young's modulus remained constant along the fatigue life at the average value of 127502 ± 362 MPa.

Yang et al³³ found that for some ductile materials the elastic modulus decreases according to the increase of plastic strain is due to the movable dislocation accompanying the pile-up of dislocations near to the grain boundary. This behaviour was also observed by Jyoti et al³⁴ in low cycle fatigue tests in an annealed titanium alloy Ti6Al4V obtained by conventional processes. These authors observed the formation of slip band, hexagonal network and sub-grain boundary using a transmission electron microscope (TEM) and transmission Kikuchi diffraction (TKD). At lower strain amplitudes evidences of the formation of slip bands and dislocation rearrangement was observed, while, in the case of higher strain amplitude, the development of defects in the structure and annihilation/rearrangement of dislocations occurs simultaneously.

As previously mentioned, the phase β is more easily deformed, as it is less hard and more ductile than the α phase. Consequently, the deformation in phase β causes progressive interactions of dislocations (dislocation pile-ups) with the α phase. This leads to the appearance of slip bands that gradually become new grain boundaries, leading to progressive divisions of β phase grains, giving rise to the formation of β phase sub-grains. This modification of microstructure contributes to the progressive increase in back stress and the decrease of the friction stress. As this variation increases with the accumulation of plastic deformation, a gradual decreasing of Young's modulus occurred with the fatigue life. Moreover, this effect in the reduction of Young's modulus will increase with the total strain amplitude. The progressive reduction of Young's modulus with the plastic strain amplitude causes a very significant reduction in the value of the fatigue strength exponent, b , of the Basquin relationship (lowest slope, see Table 3), in the elasto-plastic regime when compared to the predominantly elastic regime.

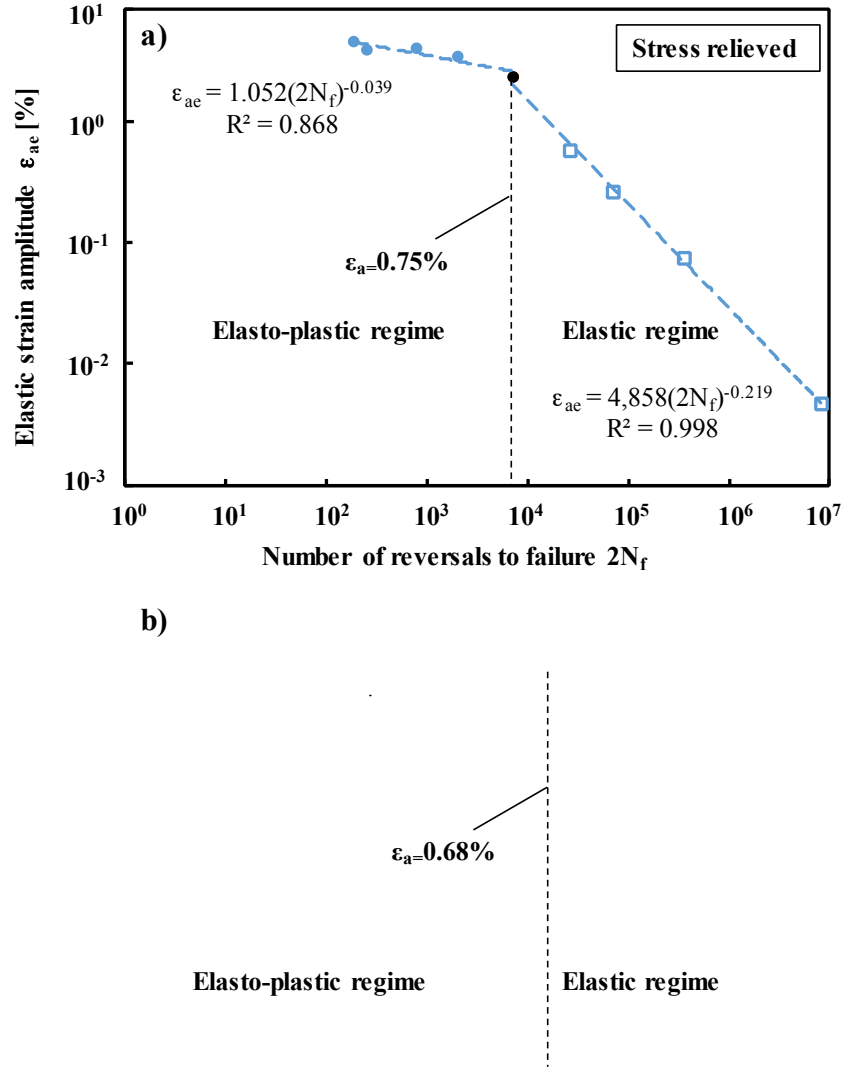
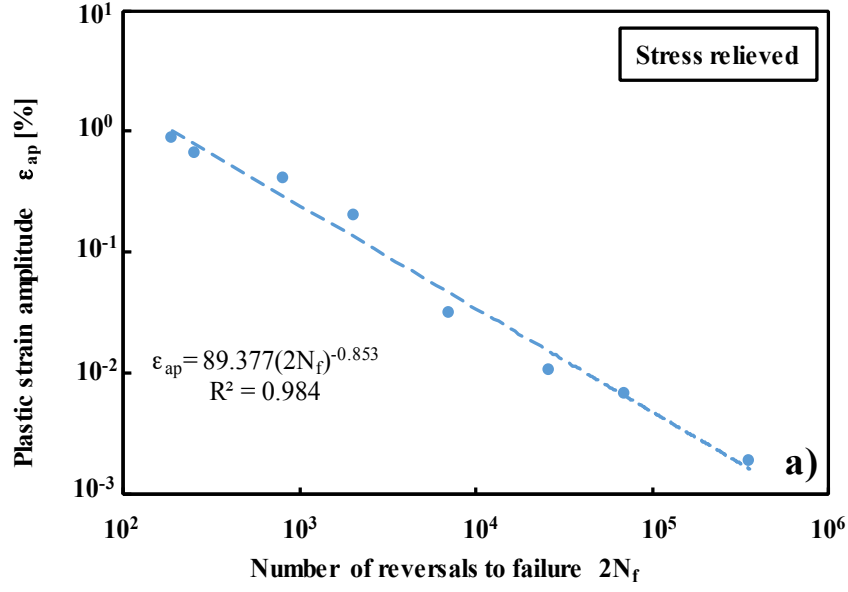


Figure 7. Baquin relationship. a) Stress relieved samples and b) HIPed samples.

The Coffin-Manson relationship³⁵⁻³⁶, Eq. 4, relates the plastic component of the total strain amplitude, ϵ_{ap} , with the number of reversals to failure, where ϵ_f' is the fatigue ductility coefficient, and c is the fatigue ductility exponent.

$$\epsilon_{ap} = \epsilon_f' (2N_f)^c \quad (4)$$

Fig. 8 presents the Coffin-Manson relationship for both treatments. As can be observed, a linear relationship was achieved on a log-log scale, for both stress relieved and HIPed samples, fitting Eq. 4 with a high correlation coefficient in both cases.



b)

Figure 8. Coffin-Manson relationship. a) Stress relieved samples and b) HIPed samples.

Morrow³⁷ proposed that the relationship between the metal's fatigue life and the total strain amplitude can be obtained by superimposing the elastic and plastic strain components. Therefore, adding Eq. 3 and 4, Eq. 5 is obtained.

$$\epsilon_a = \epsilon_f' (2N_f)^c + \frac{\sigma_f'}{E} (2N_f)^b \quad (5)$$

The variation of elastic, plastic, and total strain amplitudes with the number of reversals to failure is shown in Fig. 9a) and 9b) for the stress relief treatment and the HIP treatment, respectively.

a)

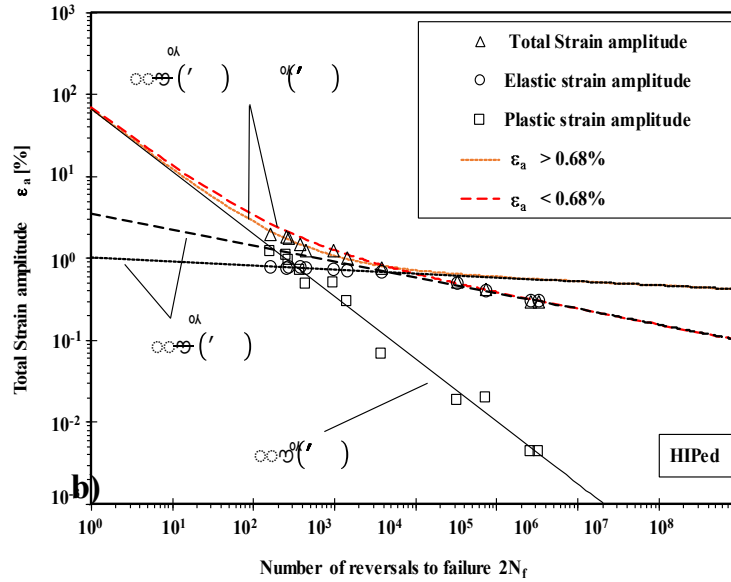


Figure 9. Total, plastic and elastic strain amplitudes versus number of reversals failure.

a) Stress relieved samples and b) HIPed samples.

In each of Fig. 9a) and 9b), two different curves were plotted to fit Eq. 5 in order to take into account the bilinear behaviour of this material. Table 3 summarizes the compilation of fatigue strength and fatigue ductility parameters for both heat treatments in the predominantly elastic tests ($\epsilon_a < 0.75\%$ for the stress relieved sample and $\epsilon_a < 0.68\%$ for the HIPed sample) and tests in elasto-plastic regime ($\epsilon_a > 0.75\%$ for the stress relieved sample and $\epsilon_a > 0.68\%$ for the HIPed sample).

Table 3. Fatigue strength, fatigue ductility and energy-based parameters for stress relieved and HIPed samples.

Stress relieved				
b	c	σ' _f [MPa]	ε' _f [%]	ε_a > 0.75 [%]
-0.050	-0.853	1328	89.4	
b	c	σ' _f [MPa]	ε' _f [%]	ε_a < 0.75 [%]
-0.219	-0.853	6129	89.4	
HIPed				
b	c	σ' _f [MPa]	ε' _f [%]	ε_a > 0.68 [%]
-0.039	-0.764	1311	67.2	
b	c	σ' _f [MPa]	ε' _f [%]	ε_a < 0.68 [%]
-0.193	-0.764	4416	67.2	
Stress relieved				
kp [MJ/m³]		αp	ΔW^{e+} [MJ/m³]	
1524		-0.695	0.345	
HIPed				
kp [MJ/m³]		αp	ΔW^{e+} [MJ/m³]	
599		-0.573	0.215	

In order to compare the difference in fatigue performance between the two treatments Fig. 10a) and 10b) show the strain-life and stress-life curves, respectively. Fig. 10a) shows that in the case of the ϵ -N curves, the difference in fatigue strength is minimal between the two treatments. In contrast, the S-N curves (Fig. 10b)) show higher fatigue strength for specimen's with stress relieved treatment than for the HIPed specimen's in the elasto-plastic regime. Despite the potential of the HIP treatment to decrease typically defects of SLM parts (porosity, lack of fusion, among others), the induced transformation of the microstructure and the consequent loss of hardness and mechanical resistance, leads to a decrease in fatigue resistance for LCF. On the other hand, in the high cycle fatigue regime (predominantly elastic tests), the fatigue performance for the two treatments was quite similar. The S-N curves also exhibited a

bilinear behaviour (two dissimilar slopes in the two regimes) by the reasons already explained.

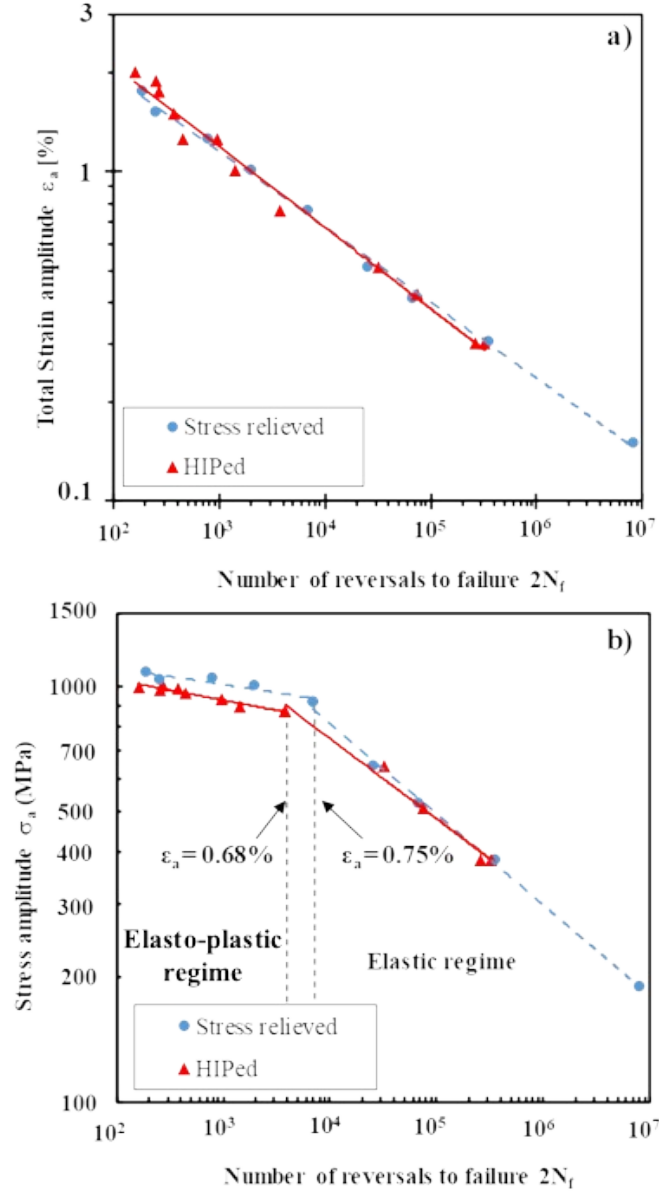


Figure 10. Fatigue performance. a) ϵ -N curves and b) S-N curves.

3.4.6. Strain energy vs fatigue life

The total strain energy density range was calculated as the sum of the plastic and the positive elastic strain energy density ranges of the mid-life hysteresis loops. It can be written by Eq. 6^{31,38}.

$$\Delta W_T = kp(2Nf)^{ap} + \Delta W^{e+el} \quad (6)$$

where kp and αp are material constants, and ΔW^{e+} is the tensile elastic energy range at the material fatigue limit estimated here at $2N_f = 10^7$. Fig. 11a) and 11b) present the total strain energy density range of the mid-life hysteresis loops against the number of reversals to failure, for the stress relieved treatment and HIP treatment, respectively. In general, the fitted Eq. 6 shows a good concordance with the tests results and serving both for short and long fatigue lives. Table 3 summarizes the energy-based parameters for both stress relieved and HIPed samples.

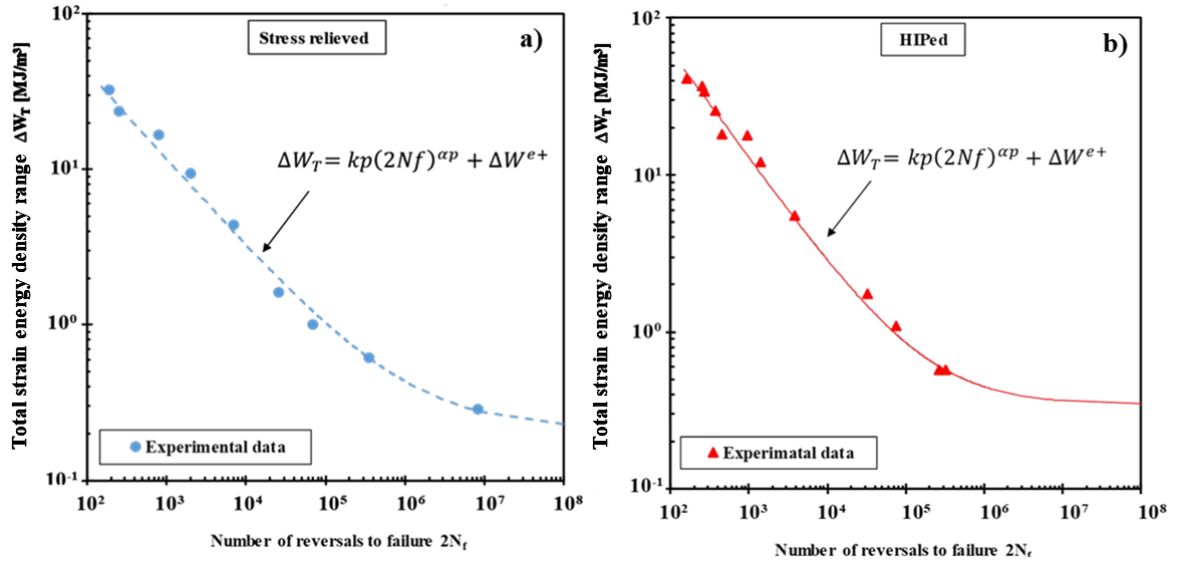


Figure 11. Total strain energy density range of the mid-life hysteresis loops vs number of reversals to failure: a) Stress relieved specimen's; b) HIPed specimen's.

3.5. Fatigue life predictions

Fig. 12 compares the S-N curves obtained for smooth and notched samples subjected to the stress relieved treatment. It is possible to conclude that the un-notched samples presented higher fatigue strength than the notched sample due to the presence of the notch that generates stress concentration. Once the tests were performed in deformation control and all the tests were performed in the predominantly elastic regime, the stress values were obtained by Hooke's law. The maximum local stresses plotted in Fig. 12 for the notched samples were calculated using Eq. 7 (taking into account that $R=-1, \sigma_{max}=\sigma_a$).

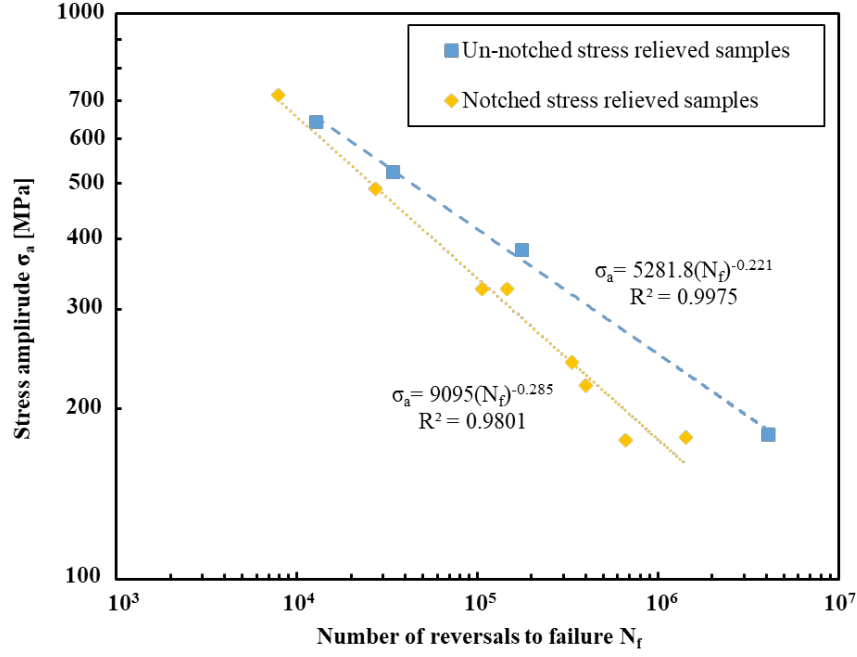


Figure 12. S-N curves for un-notched and notched stress relieved samples. $R=-1$.

Fatigue life predictions for the stress relieved notched specimen's were performed using two different models: the SWT³⁹ (Smith-Watson-Topper) model and a model based on the total strain energy density (Eq. 6). Both approaches need to know the value of the static stress concentration factor. The static concentration factor, k_t , can be estimated by Eq. 7, where, σ_{max} , is the maximum stress at the notch root and σ_n the nominal stress.

$$k_t = \frac{\sigma_{max}}{\sigma_n} \quad (7)$$

The maximum stress, σ_{max} , was obtained by the finite element method (FEM) using the CAD software Autodesk Inventor®, assuming an isotropic linear elastic material. Through this software it was possible to create a precise model of the geometry of the notched specimen's. The nominal stress, σ_n , was calculated by dividing the maximum applied load (10 kN) by the cross-section area of the notch, being estimated that σ_{max} =1326 MPa by FEM. Then, applying Eq. 7 a value of $k_t=1.73$ was obtained.

The fatigue stress concentration factor, k_f , was obtained by Eq. 8, where: σ_{fu} is the fatigue strength for smooth specimen's and σ_{fe} the fatigue strength for notched specimen's. These two parameters were calculated using the S-N relationships presented in Fig. 12 assuming $N_f=10^7$. The estimated value of k_f was 1.63.

$$k_f = \frac{\sigma_{fu}}{\sigma_{fe}} \quad (8)$$

Using the values of the static and fatigue stress concentration factors it is possible to calculate the notch sensibility, q , using Eq. 9⁴⁰, being equal to 0.86 at the fatigue life of $N_f=10^7$.

$$q = \frac{k_f - 1}{k_t - 1} \quad (9)$$

This notch sensibility value of 0.86, being close to unity, indicates that this titanium alloy produced by SLM have a high sensibility to the presence of geometric discontinuities such as notches or surface defects resulting from the manufacturing process. Defining the fatigue strength as the stress amplitude for a fatigue life of $N_f=10^7$ a value of $\sigma_{fu}=150$ MPa is obtained. This value suffered a drastic reduction when compared to the fatigue strength of the wrought material, which is $\sigma_{fu}=410$ MPa⁴⁰, ($N_f=10^7$, light polished samples and $R=-1$) due to a poor surface finish, porosity and detrimental residual stresses as reported by Wycisk et al.⁴, Edwards et al.⁵, Leuders et al.⁶ and Cao et al.⁷.

All maximum stresses in Fig. 12 are within the predominantly elastic regime, whereby the parameters used from Table 3 were those corresponding to $\epsilon_a < 0.75$ %. Using the equation proposed by Smith-Watson-Topper (SWT)³⁹ based on strain-life testing data and the Eq. 7, the fatigue life can be predicted for the notched specimen's by Eq. 10.

$$\dot{\epsilon} \quad (10)$$

For the model based on total strain energy density (Eq. 6), the total energy density ΔW_T was calculated by Eq.11 once all tests were in the predominantly elastic regime,

$$\Delta W_T = \frac{\Delta \sigma_{loc} \Delta \epsilon_{loc}}{2} \quad (11)$$

where $\Delta \sigma_{loc}$ is the maximum local stress ranges and $\Delta \epsilon_{loc}$ is the maximum local strain ranges estimated through Hooke's law. Predictions of the fatigue life were performed by Eq. 6 using the parameters presented in Table 3 for the stress relieved sample. Fig. 13a) and 13b) plot the fatigue life predictions versus experimental fatigue lives for SWT and the energy model, respectively. There is a good correlation between predicted and tests results once 100% of the points are within scatter bands with a factor of 2. The SWT

model gives non-conservative fatigue life predictions while the model based on the total strain energy density showed to be conservative.

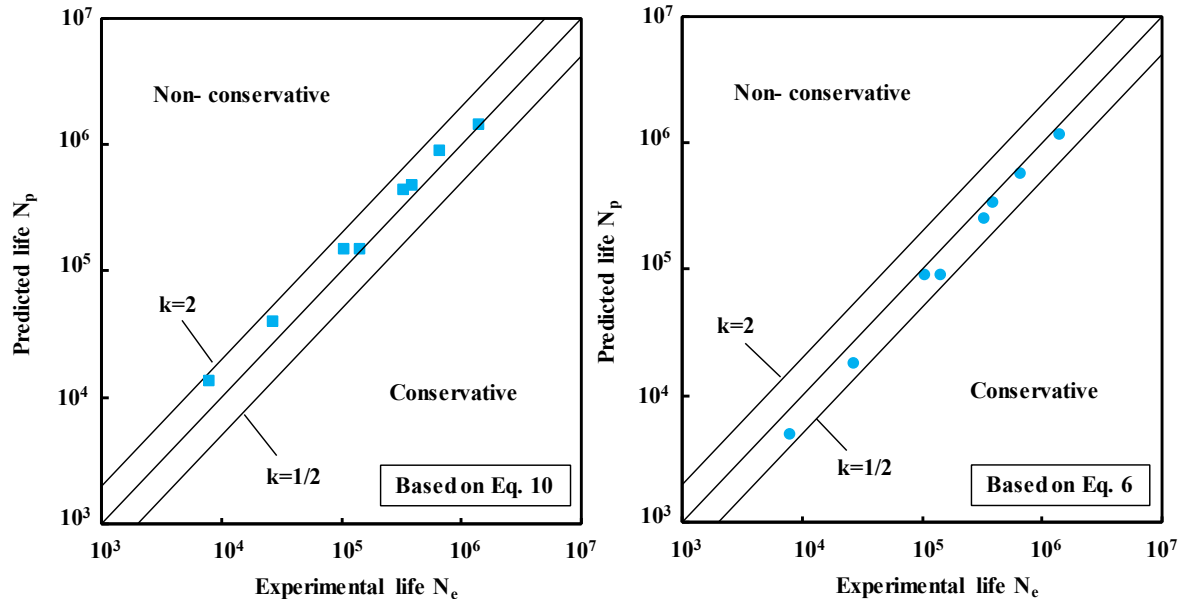


Figure 13. Fatigue life predictions versus experimental fatigue lives. a) SWT model
b) total strain energy density based model.

4. CONCLUSIONS

In this work, fatigue tests were performed ($R_e=-1$) in samples produced by SLM in Ti6Al4V alloy in order to understand the influence of two heat treatments (stress relief and the HIP treatment) on the fatigue behaviour. Fatigue life predictions were also performed using two different prediction models for notched specimen's. From this study the main conclusions drawn are the following:

- The application of HIP treatment caused a microstructural transformation leading to a loss of hardness and consequently a decrease of the monotonic mechanical properties;
- The stress relieved sample showed a cyclic softening while the HIPed sample did not alter its behaviour under cyclic deformation. In both cases a non-Masing behaviour was obtained.
- The elastic strain amplitude vs fatigue life curve presented a bilinear behaviour, explained by the Young's modulus decreasing during the cyclic elasto-plastic tests as a consequence of the formation of a sub-grained microstructure;
- The HIP treatment reduces the fatigue strength in comparison with the stress relief treatment;

- Two fatigue life prediction models were successfully applied in notched stress relieved samples using the fatigue properties obtained in this work. The SWT model tended to be non-conservative while the model based on the total strain energy density showed to be conservative.

ACKNOWLEDGMENTS

The authors thank to the project no. 028789, financed by the European Regional Development Fund (FEDER), through the Portugal-2020 program (PT2020), under the Regional Operational Program of the Center (CENTRO-01-0145-FEDER-028789), and the project POCI-01-0247-FEDER-042536, financed by European Funds, through program COMPETE2020, under the Eureka smart label S0129-AddDies. Furthermore, the authors also thank to FEDER funds through the program COMPETE–Programa Operacional Factores de Competitividade – and by national funds through FCT–Fundação para a Ciência e a Tecnologia , under the project UIDB/00285/2020.

CONFLICT OF INTEREST

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

AUTHORS CONTRIBUTIONS

J.S. Jesus: Experimental tests, results analysis and writing original manuscript. L.P. Borrego: writing review, editing and results analysis; J.A.M. Ferreira: supervise, J.D. Costa: writing review, editing and results analysis; C. Capela: specimens acquisition and transportation. All authors read and agreed to this version of the manuscript.

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