

Fatigue Life Prediction of Additively Manufactured Metallic Materials Using a Fracture Mechanics Approach

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Abstract

The present study aims to model the fatigue strength of additively manufactured metallic materials employing a fracture mechanics approach. Specimens with different build orientations were subjected to strain controlled fatigue testing. Upon failure, the defect(s) responsible for crack initiation were identified by fractographic analysis. From these defects an equivalent internal defect size is calculated using the \sqrt{area} method based on Murakami model. Using this parameter, the elastic-plastic energy release rate (ΔJ_{eff}) was determined, and the relationship between ΔJ_{eff} and fatigue life was investigated. The results showed that this method improves the predictability of the fatigue strength of additively manufactured materials when the defects size and location is known. The $\Delta J_{eff} - N_f$ relationship appeared to better fit the fatigue data of the experimental materials as compared to the $\varepsilon_a - N_f$ relationship and contributed to a reduction in data scatter.

Keywords: Additive manufacturing, Defects, Fracture mechanics, Fatigue modeling

Introduction

Additive manufacturing (AM) is an attractive fabrication method with many benefits. AM allows for the fabrication of parts with complex geometries unobtainable through traditional machining processes [1-3], functionally graded materials [4-9], and internal porosity [4, 6, 9]. Additionally, AM can be used to fabricate parts customized for specific applications. However, despite these advantages, AM cannot attain full industry adoption until the effects of structural inconsistencies on the tensile and fatigue properties of additively manufactured components can be accounted for.

From a fatigue perspective, some of the most impactful structural inconsistencies found in additively manufactured components are the defects, such as pores and lack of fusion (LOF) voids, inherent to the AM process. These internal defects can act as stress risers and cause premature fatigue failure of the specimens [10, 11]. Pores and LOF have also been shown to contribute to the scatter found in the fatigue data of additively manufactured materials [12, 13]. While post processing techniques such as hot isostatic pressing (HIP) can help shrink these pores and improve fatigue performance [14-16], its application might not be desirable due to the changes in microstructure stemming from the applied heat and pressure. Additionally, HIP has been shown to

be less effective at improving the fatigue properties of specimens with surfaces in the as-built condition, as the detrimental effects of surface roughness may override those of internal defects [16-19]. At the current state of AM technology, it is nearly impossible to fabricate defect free additively manufactured specimens. Therefore, a defect sensitive fatigue analysis is required to further improve the predictability of the fatigue response of AM parts.

Multiple models have been used to account for the fatigue behavior of additively manufactured materials, each accounting for different sources of uncertainty [10, 20-24]. One model that has exhibited promising results in accounting for the effects of process inherent defects is the \sqrt{area} model by Murakami et al. [25, 28]. This model uses a stress intensity factor calculated via an equivalent internal defect size derived from the defect area projected on the loading plane, while also accounting for the location of the defect.

In this study, the fatigue behavior of 316L stainless steel (SS) and Ti-6Al-4V are investigated. Fractography analysis was used to identify the defects responsible for crack initiation and failure of the specimens. From these defects, the equivalent internal flaw sizes were determined using the \sqrt{area} method developed by Murakami et al [25]. Using the equivalent flaw size, the elastic-plastic energy release rate ΔJ_{eff} was calculated, and the relationship between defect size and fatigue life was then investigated.

Experimental Procedure

The strain-life data utilized in this study was obtained from strain-controlled fully reversed ($R_\epsilon = -1$) fatigue tests conducted on both additively manufactured 316L SS and Ti-6Al-4V, reported in detail in [29] and [30], respectively. The specimens were fabricated in three different build orientations (vertical, diagonal (45°), and horizontal) via laser based powder bed fusion (L-PBF) process in an argon environment. The 316L SS specimens were fabricated via a Renishaw AM250 machine [29], while the Ti-6Al-4V specimens were fabricated utilizing an EOS M290 [28].

After fatigue testing, the fracture surfaces were carefully examined using a scanning electron microscope (SEM) in order to determine the factors responsible for crack initiation and the failure mechanism in each of the material. Special care was taken to determine if multiple crack initiation sites were present on each fracture surface. If multiple initiation sites were found, the largest one was used in the ΔJ_{eff} calculations.

Fracture Mechanics

In defect free materials, the fatigue life can be divided into three stages: crack initiation, crack growth, and final fracture. First, cyclic loading and plastic deformation causes the movement of dislocations within the material, called “slip.” A crack then forms due to formation of persistent slip bands or intrusions and extrusions on the specimen surface [31], or due to slip between grain boundaries [30]. This crack grows due to plastic deformation at the crack tip stemming from cyclic loading and unloading, and, once the crack reaches critical size, final fracture occurs [31]. However, additively manufactured materials can behave differently than typically processed materials. During AM, pores and LOF voids can form due to sub-optimal process parameters [33, 34], high speed flow [35] or vapor recoil [36] in the melt-pool, or from porosity contained in the feedstock used during fabrication [12, 33]. These process inherent defects can be considered as pre-existing cracks, thereby shortening or bypassing the crack initiation phase [12, 15, 20, 22, 37, 38]. Fractography analysis of the specimens used in this study revealed pores from entrapped

gases, as well as LOF and defects resulting from partially melted powder particles. These defects were present at random locations and varied in both shape and size, as can be seen in Fig. 1(a-b) and 1(c-d) in 316L SS and Ti-6Al-4V specimens, respectively.

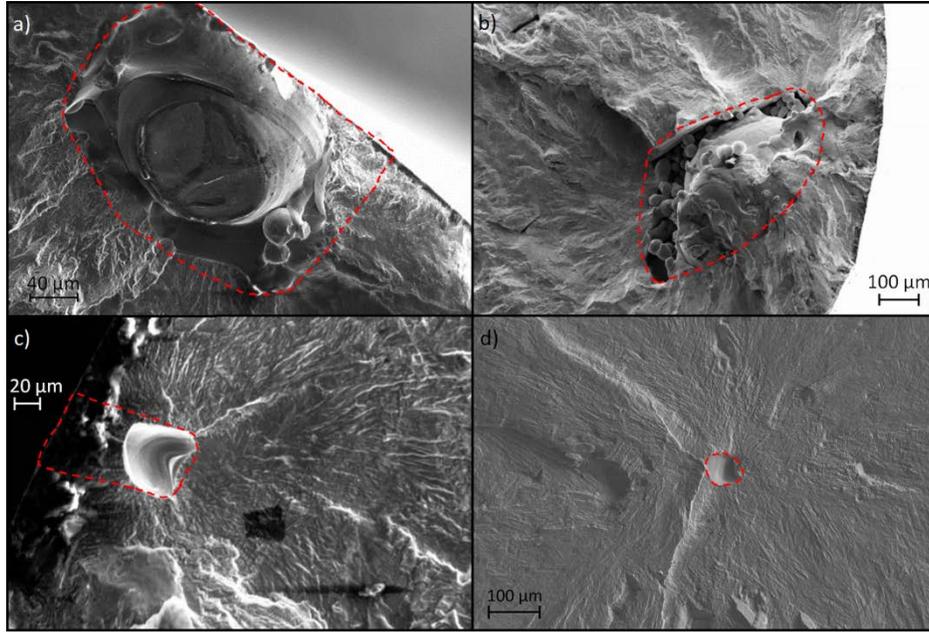


Fig. 1. Representative fracture surfaces for (a-b) 316L SS [29] and (c-d) Ti-6Al-4V [30] specimens, with area used in equivalent internal flaw size calculations outlined in red.

Therefore, considering defects as pre-existing cracks, a fracture mechanics-based concept was used in this study to relate the defect geometry to the elastic-plastic energy release rate, ΔJ_{eff} , using the following [24]:

$$\Delta J_{eff} = Y^2 \pi \sqrt{area_{eff}} \left[\frac{\Delta \sigma_{eff}^2}{E'} + \frac{3(\Delta \sigma_{eff})(\Delta \varepsilon_{pl,eff})}{4\sqrt{n'}} \right] \quad (1)$$

where $\sqrt{area_{eff}}$ is the effective size of the defect determined using Murakami's approach [25], Y is a shape factor, taken as 0.65 for surface defects and 0.5 for sub-surface defects [25], n' is the cyclic strain hardening exponent from the Ramberg-Osgood relationship, and E' is the cyclic modulus of elasticity. $\Delta \sigma_{eff}$ is the stress range, and $\Delta \varepsilon_{pl,eff}$ is the plastic strain range further calculated using [26].

$$\Delta \sigma_{eff} = \sigma_{max} - \sigma_{cl} \quad (2)$$

$$\Delta \varepsilon_{pl,eff} = \Delta \varepsilon_{eff} - \Delta \varepsilon_{e,eff} \quad (3)$$

where σ_{max} is the maximum stress, σ_{cl} is the crack closing stress that is related to the plasticity induced crack closure theory presented in [26, 27]. Similarly, $\Delta \varepsilon_{eff}$ is the effective strain range calculated by subtracting crack tip opening strain from the maximum strain, and $\Delta \varepsilon_{e,eff}$ is the

elastic portion of the effective strain range defined as the ratio of cyclic modulus and effective stress range [26]:

$$\Delta\varepsilon_{eff} = \varepsilon_{max} - \varepsilon_{op} \quad (4)$$

$$\Delta\varepsilon_{e,eff} = \frac{E'}{\Delta\sigma_{eff}} \quad (5)$$

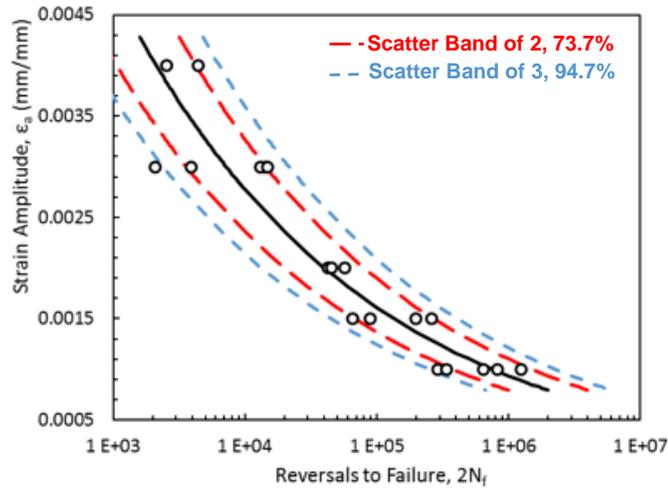
It should be noted that ΔJ_{eff} in Eq. 1 accounts for both the plastic and elastic energy released during the life of a component. This is important, as at high strain amplitudes, the fatigue deformation is governed by both elastic and plastic strains. At lower strain amplitudes, the fatigue deformation was almost entirely governed by elastic strain ($\Delta\varepsilon_{pl,eff} \approx 0$). In such cases, ΔJ_{eff} can be related to stress intensity factor, K_{max} , as [31]:

$$K_{max} = \sqrt{\Delta J_{eff} E'} \quad (2)$$

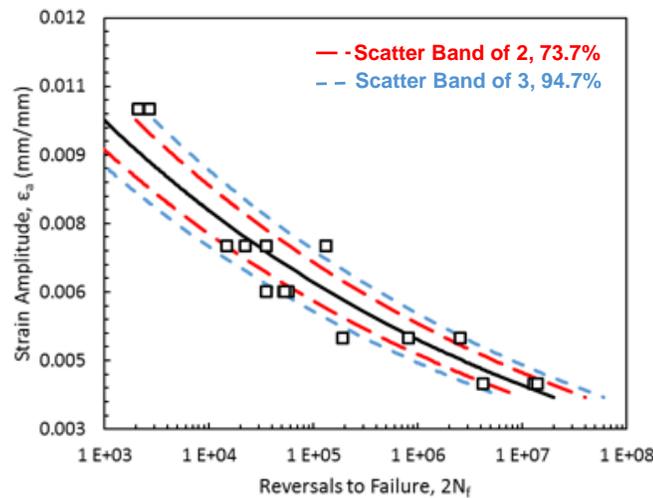
Note that the $\sqrt{area_{eff}}$ used in these calculations can be greater than the actual size of the defect, as shown in Fig. 1.

Results and Discussion

The strain life correlations for L-PBF 316L SS and Ti-6Al-4V are shown in Fig. 2(a) and 2(b), respectively, along with factor of two and factor of three scatter bands. Relatively larger amounts of scatter in the fatigue life data of both 316L SS and Ti-6Al-4V were primarily observed in the high cycle fatigue (HCF) regime, where the influence of porosity and other defects on fatigue life is more pronounced [12, 34]. For 316L SS, the overall larger scatter was attributed to the inconsistent variations in size and distribution of LOF defects serving as crack initiation factors in the tested specimens [29]. On the other hand, for the Ti-6Al-4V specimens, the scatter in fatigue life data was attributed to more uniformly shaped pores serving as crack initiation sites rather than LOF as seen in the 316L SS [30]. Additionally, as was noted in the initial study of the 316L SS specimens, due to the influence of LOF defect orientation with respect to the loading direction there were some build orientation effects observed in the 316L SS specimens [29]. Results showed that horizontally fabricated specimens exhibited higher fatigue life as compared to the vertical specimens, which in turn performed better than the diagonal specimens [29]. Furthermore, from Fig. 2, the inability of strain amplitude to correlate the fatigue life data was also evident as only 73.7% and 37.5% of the fatigue life data were within the factor of two scatter band for 316L SS and Ti-6Al-4V, respectively. Therefore, because the presence of defects such as LOF and gas entrapped pores were considered as the factor responsible for this scatter in the data, a damage parameter that can incorporate the effects of defects might be able to accurately correlate the fatigue life data for both AM materials.



(a)

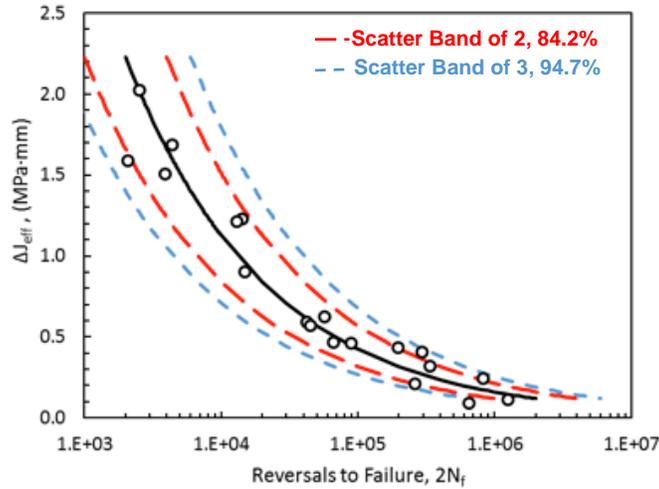


(b)

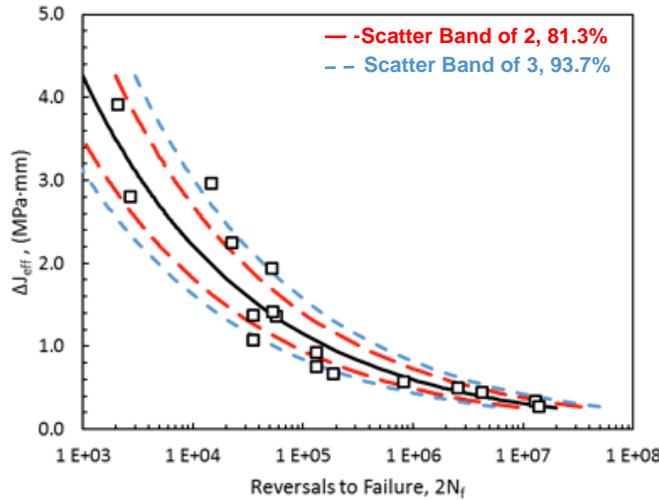
Fig. 2. Strain versus reversals to failure for fully-reversed, strain-controlled fatigue tests of machined L-PBF (a) 316L SS [29] and (b) Ti-6Al-4V [30] specimens.

After the fractographic investigation and $\sqrt{area_{eff}}$ calculations were completed, ΔJ_{eff} was calculated using Eq. 1. The ΔJ_{eff} versus fatigue life correlation, plotted in Fig. 3, was shown to satisfactorily model the fatigue behavior of both the L-PBF Ti-6Al-4V and 316L SS. The scatter observed in the strain-life relationship appears to have been reduced, as 84% and 81.3% of the fatigue data were within the factor of 2 scatter bands as compared to 73.8% and 37.5% for 316L SS and Ti-6Al-4V, respectively. It is interesting to note that the amount of data points falling within the factor of 3 scatter band did not change for the 316L SS, but did increase for the Ti-6Al-4V, increasing from 75.0% to 93.7%. The ΔJ_{eff} relationship was able to fit the curve well due to a number of factors. ΔJ_{eff} has been used to assess fatigue life in low cycle fatigue (LCF) scenarios in which cracks begin propagating during the first loading cycle [39]. As mentioned previously,

the defects such as voids and pores present in additively manufactured materials can be treated as preexisting cracks [12, 15, 20, 22, 37, 38]. Therefore, crack propagation can be considered to begin in the first loading cycle in additively manufactured materials as well. Furthermore, calculation of ΔJ_{eff} also considers the local plastic deformation in crack tip by incorporating effective stress, which is based on the crack closer concept. Through the use of the $\sqrt{area_{eff}}$ parameter and shape factor Y , ΔJ_{eff} can also account for defect size and location. Defects have been shown to be a source of scatter in strain-controlled fatigue data for additively manufactured components; therefore, accounting for the effects of size and location of such internal defects is vital to accurately correlate the fatigue life data.



(a)



(b)

Fig. 3. ΔJ_{eff} -life relationships for L-PBF (a) 316L SS [29] and (b) Ti-6Al-4V [30] specimens.

Conclusions

In this study, the ΔJ_{eff} was calculated for L-PBF 316L stainless steel and Ti-6Al-4V specimens fabricated in multiple build orientations, and its relationship to fatigue life was investigated based on the type, size, shape, and location of the defects responsible for the failure in the specimens. From these results, the following conclusions have been drawn:

1. A ΔJ_{eff} -life relationship satisfactorily fit the fatigue behavior of both the 316L SS and Ti-6Al-4V L-PBF specimens.
2. The ΔJ_{eff} -life relationship for both materials exhibited reduced scatter as compared to the corresponding strain-life data.
3. The better correlation of fatigue life data obtained using ΔJ_{eff} as a damage parameter can be attributed to its ability to incorporate shape, size, and location of the defects along with localized plastic deformation at the tip of these defects, which acts as a small crack.
4. Therefore, with the crack growth properties known, ΔJ_{eff} may be utilized to predict fatigue life in AM materials using crack growth behavior

From the results of this study, it has been shown that ΔJ_{eff} can account for the internal defects inherent to the additive process and reduce the scatter in the fatigue response of additively manufactured materials. Defect shape, however, may be a factor in the effectiveness of this approach. Further study is needed to investigate the effects of the defect geometry serving as the crack initiation site, as well as the applicability to other material systems and specimen conditions.

Acknowledgement

This material is based upon work partially supported by the National Science Foundation (NSF) under Grant No. 1657195 and the National Aeronautics and Space Administration (NASA) Award No. NNX17AJ93A and 80MSFC17M0023.

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