Mechanical properties of titanium-based Ti-6Al-4V alloys manufactured by powder bed additive manufacture
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Abstract
Additive manufacturing is currently a topic of considerable interest at both academic and industrial levels. While a significant amount of data exists on the mechanical properties and structure-property relationships of traditional wrought alloys, less information is available on alloys manufactured by additive manufacture. This review examines current state-of-the art of the manufacture of titanium-based Ti-6Al-4V alloys by powder bed additive manufacture. Published mechanical properties to date are collected which include tensile strength, yield strength, hardness, wear, fracture toughness and fatigue. Differences in microstructure and properties are compared to conventional wrought alloys of the same composition are described.

Introduction

Additive manufacturing is a term widely used to refer to a range of manufacturing processes where, unlike in traditional ‘subtractive manufacturing’ processes, raw material is selectively laid down or deposited as required. This allows for a greater design freedom, although there are design constraints depending on the precise method used. These methods cover a wide range of materials (polymers, metals and alloys, ceramics, composites), feedstocks (powder, wire, foil, liquid), and consolidation mechanisms (sintering, melting, ultrasonic consolidation, photopolymerisation, etc).

Among the potential additive manufacturing methods, powder bed methods are some of the most well developed, although powder fed and wire fed systems including blown powder and Wire+Arc are available [1-4]. In the powder bed process the feedstock is a fine powder, which is laid down in incremental layers; generally measured in layers of thickness between 10 and 100 µm. Each layer is selectively consolidated to create a final 3D object, generally by melting or sintering (with a heat source such as a laser or electron beam), although other methods such as selective deposition of a binder have been developed.
Fig. 1 is a schematic of the powder bed laser additive manufacturing (LAM) process which includes the deposition of powder layers, scanning by heat source and final creation of the product. After removal of the product any unconsolidated powder remains in the machine during the build, but can be subsequently removed, and recycled. The ability to recycle the powder depends, to a certain extent, on the precise AM method used since some aging of powder does occur, causing coarsening of the powder which may affect part properties.

A range of materials can be produced by powder bed methods, among them metals and alloys. Due to the high power requirements to produce fully melted material, early work often used multiple phase systems with a low melting point component or sintering of the metal powder rather than fully melting the material. However, modern machines are capable of producing full melting even in high temperature alloys, including tungsten.

The challenges in the production of metals and alloy systems by powder bed additive manufacturing include:

i. thermal residual stresses developed produced during the build,

ii. the high surface roughness of the manufactured part due to partially melted or un-melted material from the surrounding powder bed sticking to the surface,

iii. the lack of self-supporting structures, limiting production of overhangs unless support structures are added to the design,

iv. the generation of microstructures that are different compared to equivalent wrought material due to rapid heating and cooling of the material and the absence of cold work in
v. the fabrication of components free of defects such as porosity, delamination, lack of fusion of layers and balling is also a challenge and requires precise control of a number of process variables such as alloy chemical composition, powder size distribution, powder flow rate, laser power, power density distribution and laser scanning speed [8-10].

These and other factors mean that the material properties of metals and alloys produced by powder bed additive manufacturing are often very different from those expected of wrought or cast material,

**Scope of review**

This review paper examines the evidence in the literature as to the effect on materials properties of processing a common titanium alloy, Ti-6Al-4V, by additive manufacturing. This review covers all additive powder bed methods, where the raw material is in the form of a Ti-6Al-4V powder which is laid down layer-by-layer and selectively melted with either an electron or laser beam. Although there are a variety of titanium alloys which can be processed using these methods, this review focuses on Ti-6Al-4V since it is the most commonly used titanium alloy in additive manufacturing. This is a result of its high strength-to-weight ratio and resistance to corrosion of Ti-6Al-4V make it suitable for a wide range of applications, particularly in industries, such as aerospace and transport, where component weight is important.

Medical applications also exist for Ti-6Al-4V but due to concerns with the possible impact of vanadium, Ti-6Al-7Nb was developed as an alternative for applications such as hip implants. This material is also processable by powder bed methods [11, 12]. Commercially Pure (CP) titanium [13-16] can also be produced. However, comparatively limited information on the properties of CP titanium and Ti-6Al-7Nb as produced by laser additive manufacturing (LAM) or electron beam melting (EBM) is available and therefore we concentrate on Ti-6Al-4V to indicate the range of important issues when producing important alloys via additive manufacture.

ASTM standards for powder bed manufacturing are being currently developed under subcommittee F-42 and standards exist for terminology ([17]) and for the manufacture of Ti-6Al-4V [18]; vocabulary and definitions are under ASTM F2792-12a. The main mechanical properties of interest for appropriate materials selection for the types of application described above are:
i. tensile strength (and its dependency on heat treatment),
ii. hardness (and relation to tensile strength),
iii. wear,
iv. fracture toughness and impact toughness,
v. fatigue and the appropriate S-N curve.

**Ti-6Al-4V Phases and Properties**

There are excellent sources that overview the microstructure and properties of titanium alloys [19,20]. Pure titanium at room temperature has an hexagonal close packed (hcp) structure, called the alpha (α)-phase which transforms to the body centred cubic structure (bcc) at elevated temperature (888°C) which is called the beta (β)-phase [21].

Commercially available titanium alloys are fall into a number of categories, which reflect the contribution of the α and β phases; this includes (i) α alloys, (ii) near- α, (iii) α-β (including Ti-6Al-4V), (iv) near β and (v) β alloys. A schematic quasi-vertical section of a ternary titanium system containing both α and β stabilising species is shown in Fig. 2a; the composition ranges corresponding to the five classes of alloy are also indicated in the upper part of the diagram. The dashed line marked M_S/M_F represents the martensitic start and finish lines, which are typically very close together [19]. The Ti-6Al-4V alloy is an α-β alloy, containing 6wt% aluminium, which acts as an α stabiliser and 4wt% vanadium as a β stabiliser. The β-transus for this alloys system is approximately 980°C under equilibrium conditions and above this temperature Ti-6Al-4V consists of 100% β phase. This temperature is of interest since heat treatments are often undertaken at a temperature above or below the β transus. As the alloy is cooled from the β transus the α phase forms in the microstructure and the characteristics and arrangement of the α phase depends on the particular heat treatment applied.
Figure 2. (a) Schematic quasi-vertical section for ternary titanium alloys containing both $\alpha$ and $\beta$ and stabilising solute; the composition ranges corresponding to the five classes of alloy are indicated in the upper part of the diagram and $M_s$ and $M_f$ are martensite start and finish temperatures respectively [19] (b) schematic of continuous cooling transformation curve (adapted from [22]).
The final structure formed in the alloy depends on the speed with which the material is cooled from above the β transus temperature [23]. Cooling slowly from the β phase field will result in α-phase which forms as Widmanstätten laths within a matrix of β; see ‘slow cool’ in Figure 2b. If the cooling rate is sufficiently fast, then β undergoes a diffusionless transformation to the martensitic phase, α’ (hcp), see ‘intermediate cooling’ curve in Figure 2b, although it has been reported that α’ (orthorhombic) martensitic type phases can form [24]. The α’ phase is of interest since it can enhance the strength of the alloy, but at the expense of ductility [25]. The α’ phase is not stable and if held at elevated temperatures below the β transus it can decompose to α and β. As an example, a temperature of approximately 800°C is required for complete breakdown of the martensite phase [26].

Annealing of wrought material is performed in the α+ β phase field of Figure 2a. A “mill-anneal” of Ti-6Al-4V is generally performed below the β-transus at approximately 700°C [27]. Mill annealing is undertaken on deformed material where the break up of α plates leads to recrystallisation of the α phase [28] leading to an equi-axed structure. The final microstructure ultimately depends on cooling rate from the anneal temperature and also at what temperatures the material is worked.

In power-bed additive manufacturing, there is a lack of cold work in the material and the heating and cooling rates are generally high, with literature values varying between $10^3$ and $10^5 \, ^\circ C s^{-1}$, which can form a fully α’ microstructure (see ‘AM cooling’ in Figure 2b). In the electron beam melting (EBM) process, the entire build chamber is at a raised temperature (typically ~700°C, although a range of temperatures are possible[30]) and the build chamber is under vacuum leading to lower rates of heat transfer. This results in a slower cooling rate, and therefore different microstructures, compared to laser additive manufacturing where the build chamber is at ambient temperature in argon, and any pre-heating is provided by heating the build plate (see Figure 1); the use of bed preheating can help to reduce residual thermal stresses and anisotropy of the properties [31].

While ‘as-built’ LAM components are therefore likely to contain the α’ phase due to the speed of cooling, compared to as-built EBM which contains more α [32] the deposition of new layers on previous layers results in a complex, time dependent temperature profiles within the part being fabricated [1]. During LAM, the repeated heating and cooling cycles result in dissolution and growth of the different alpha morphologies leading to visible banding observed macroscopically across the width [33]. Thijs et al. [34] examined in detail the development of microstructure in Ti-6Al-4V due to changes in local heat transfer conditions which are determined by the scanning
strategy and part geometry.

Thijs et al [34] and Martina et al [35,36] have examined the microstructural evolution during manufacture of Ti-6Al-4V. An example from Vrancken et al. of a Ti-6Al-4V microstructure formed by AM is shown in Figure 3a showing a fully acicular martensitic ($\alpha'$) microstructure [28] and growth bands. Martina et al [35,36] have examined in detail the formation of growth bands in the Wire+Arc process where there is remelting of part of the previous layer as each layer is deposited and annealing of the layers below. Since the heat flow is primarily through the base-plate on which the material is being built, a columnar microstructure in the build direction tends to develop where large columnar grains span multiple layers of the build which are identified as prior $\beta$ grains which grow epitaxially during the process [34].

Due to this non-linear cooling, and the potential for differences between intra and inter-layer bonding of the powder, Ti-6Al-4V formed by AM often shows anisotropy in its mechanical properties. Carrol et al. [37] examined the anisotropic tensile properties in Ti-6Al-4V produced by directed energy deposition additive manufacture. While the tensile strength was approximately 1060MPa in both directions, the elongations were 11% and 14% along the longitudinal and transverse directions respectively. The anisotropy in ductility was thought to be due to the columnar prior $\beta$ grain morphology (Fig 3a) and presence of grain boundary $\alpha$ phase which acts as a path for damage accumulation and fracture. Qui et al. [38] also reported materials having poorer in the horizontal direction than the vertical direction.

Vrancken et al. [28] examined the influence of heat treatment time, residence time and temperature on microstructure. The original $\alpha'$ was converted to a lamellar $\alpha$ and $\beta$ mixture on heating below the $\beta$ transus and features of the original microstructure are retained. When heat treated above the $\beta$ transus, large grain growth was observed to occur which transforms to lamellar $\alpha$ and $\beta$. For example, the microstructure of the alloy after 2h at 850°C with air cooling showing a $\alpha$-Widmanstätten microstructure or basket weave structure, as shown in Fig 3b. Such as structure is different to the wrought Ti-6Al-4V, which is a dual phase structure with isolated lamellar $\alpha$ and $\beta$ between almost equi-axed connected grains of $\alpha$ (Fig 4a). The microstructure of the additive manufactured Ti-6Al-4V reported by Mower et al. [39] also consists of martensitic $\alpha'$ phase (Fig. 4b,c) and the microstructure after subjecting the alloy to hot isostatic pressing 900°C/2hrs at 102MPa in shown in Fig. 4d which is a similar, but coarser, structure to the heat treated material in Fig 3b.
Figure 3. (a) columnar martensitic microstructure in as-built Ti-6Al-4V, from [28]. [Reproduction approval in process]. (b) 2h at 850 °C air cooling. The α phase is light and β is dark. Reprinted from [28] with permission from Elsevier.
Fig. 4 Microstructure of Ti-6Al-4V (a) wrought (b) horizontal powder bed fusion (c) vertical powder bed fusion (d) horizontal after HIP (900°C/102MPa/2hrs). [39, Creative Commons]

**Tensile properties of Ti-6Al-4 alloys**

The tensile strength of Ti-6Al-4V alloys are controlled by microstructure and thus are dependent on the thermal history during the method of manufacturing and any heat treatment. The standard annealing for wrought material as defined in MMPDS [27] is known as “mill annealed” is an annealing treatment at 705°C (1300°F) and leads to an equiaxed microstructure (Figure 4a). As discussed, due to the rapid cooling rates of LAM the as-deposited laser beam melted Ti-6Al-4V generally has a martensitic (α’) structure with banding and a columnar structure (see Fig. 3a), which compared to the typical annealed wrought material produces high tensile strength, but low values for elongation and strain to failure. For example, Hollander [40] reported that tensile strength for laser beam melted Ti-6Al-4V was well above ASTM F136[41] requirements, with an ultimate tensile strength of 1211MPa (ASTM minimum: 860MPa) and yield strength of 1100 (ASTM minimum 795MPa) but the elongation to failure (6.5%) was below the same ASTM specification (10%). Similarly Facchini et al [25] found that as-deposited laser beam melted Ti-6Al-4V had an ultimate strength of 1095MPa and yield strengths of 990MPa, both well above requirements of ISO 5832-3[42] (860 and 780MPa respectively). Again, the elongation to failure (8.1%) was below the standard specification (10%). Wauthle [49] examined the influence of build orientation on selective
laser melted Ti-6Al-4V lattice structures, indicated a strong dependency of strength with build orientation and this is also likely to be the case for dense materials.

A variety of heat treatments for additively manufactured Ti-6Al-4V have been described in the literature which generally increase the ductility but reduce the tensile strength by converting the martensite phase to the more ductile α and β phase (see Figure 3b for example). These heat treatments are summarised in Table 1. While the wrought microstructures are typically equiaxed (Fig. 3a), the heat treated LAM microstructures are often a basketweave microstructure. In addition to heat treatments that alter the phase structure of the alloy AM builds, a low-temperature heat treatment is often performed to remove residual thermal stresses (known as a 'stress relief' cycle), typically at about 650°C for two to four hours for Ti-6Al-4V[44]. In contrast, electron beam melting is performed at a raised temperature (usually 700 °C or above), which limits the formation of martensite by preventing fast cooling to room temperature. This results in an as-deposited microstructure which typically contains no martensite and has mechanical properties closer to that of annealed wrought material.

Heat treatments within the literature generally take three possible forms:

i. a single annealing temperature, above the usual stress relief cycle but below the β transus (Figure 2a) that is comparable to annealed wrought material treatments, such as the “mill anneal”;

ii. a two-stage heat treatment, first above the β transus, followed by a quench to form the martensitic phase, and a second tempering stage below the β transus

iii. hot isostatic pressing (HIP) which combine heat-treatment with externally applied pressure.

Hot Isostatic Pressing (HIP), where used, is generally performed at temperatures over 900°C on Ti-6Al-4V, the purpose of the pressure is to reduce the amount of porosity in the as-built material. HIP is commonly used for the same purpose in titanium castings [45]. Examples in the AM literature include treatments at 920°C for two hours under a pressure of 100MPa [30] 925 at 4 hours [46]. As the closure of closed porosity through HIP is thought to have relatively little effect on tensile strength, in the following sections, “heat-treated” refers to samples which have been subjected to post-processing at a higher treatment than a “stress-relieved” material (650°C). Where no heat treatment is explicitly stated or where only “stress-relieved” is stated without specification, it is assumed that treatment did not exceed 650°C and the material is treated as “as-built”.
A list of references used for tensile test of laser additive manufactured (LAM) or electron beam melted (EBM) Ti-6Al-4V is given in Table (1) and are summarised in Fig. 4 and 5. Although some references exist for the tensile testing of scaffolds or similar structures with designed porosity (e.g. [47]), these have not been included in the comparison. The generally low ductility of the as-built LAM is likely due to the present of large amounts of martensite in the structure, which is not present in EBM or heat-treated samples.

**Table 1: Summary of heat treatment cycles and tensile testing of LAM or EBM Ti-6Al-4V alloys**

<table>
<thead>
<tr>
<th>Reference</th>
<th>LAM or EBM</th>
<th>Processing/Post-Processing</th>
</tr>
</thead>
<tbody>
<tr>
<td>Hollander et al (2006) [40]</td>
<td>LAM</td>
<td>As-built and annealed at 950°C</td>
</tr>
<tr>
<td>Facchini et al (2009) [48]</td>
<td>EBM</td>
<td>As-built and HIP at 915°C/1000bar</td>
</tr>
<tr>
<td>Facchini et al (2010) [25]</td>
<td>LAM</td>
<td>As-built (two process variants) and heat-treated (two variants – temperatures unspecified)</td>
</tr>
<tr>
<td>Koike et al (2011) [49]</td>
<td>LAM/EBM</td>
<td>As-built - both at build temperature of 700°C</td>
</tr>
<tr>
<td>Thöne et al (2012) [50]</td>
<td>LAM</td>
<td>As-built and heat-treated (five variants between 750°C and 1050°C)(*)</td>
</tr>
<tr>
<td>Vrancken et al (2012) [28]</td>
<td>LAM</td>
<td>As-built and heat-treated (eight variants, including two-stage treatments)(**)</td>
</tr>
<tr>
<td>Qiu et al. (2013)</td>
<td>LAM</td>
<td>As-built and HIP</td>
</tr>
<tr>
<td>Xu et al. (2015)</td>
<td>LAM</td>
<td>In-situ decomposition of α’ martensite</td>
</tr>
<tr>
<td>Al-Bermani et al (2010) [30]</td>
<td>EBM</td>
<td>As-built and HIP (four build temperatures, as built and HIP, eight variants total)</td>
</tr>
</tbody>
</table>

*Yield strength not reported, used for tensile strength only

**One 'heat treatment' below stress-relief temperatures - counted under "as built". One heat treatment has no elongation data reported.
Figure 4. Elongation to failure and tensile strength.

Figure 5. Elongation to failure and yield strength.
It is clear from Figures 4 and 5, that there is generally a balance between the required elongation and tensile strength. It can be seen that the ‘as-built’ LAM material is typically of higher strength but is less ductile than wrought. The ‘heat-treated’ LAM or ‘as-built’ EBM are more ductile than ‘as-built’ LAM. Generally, annealing at higher temperatures leads to higher elongation and lower tensile strength. However, annealing at temperatures close to or above the β transus in Figure 2a can result in a decrease in both tensile strength and elongation, most likely due to excessive grain growth while in the β field. These results are broadly similar to what is seen in post-weld heat-treatment in Ti-6Al-4V [52]. For Ti-6Al-4V with a lamellar α and β structure the average α lath thickness has been used as a characteristic of its mechanical properties by Xu et al. [53] since yield strength versus the inverse square root of lath thickness is linear and this is analogous to the Hall Petch relationship, indicating that a forming an fine scale lamellar α and β structure leads to improved strength.

**Hardness and wear properties of Ti-6Al-4V**

A limited number of references [32, 48, 49, 51] measure both tensile strength and hardness for either LAM or EBM Ti-6Al-4V. In many alloys there is a strong correlation between tensile strength and hardness [54] but the correlation is less clear for the limited data available in the literature for AM Ti-6Al-4V.
Wear is related to, but cannot be necessarily directly predicted, the hardness of a material since the type of wear and method of testing also impact the wear behaviour. Two references were found which discussed the wear of titanium produced by additive manufacturing. Gu et al [16] found that the wear properties of commercially pure (CP) titanium produced by additive manufacturing were better than the equivalent material produced by powder metallurgy. The wear behaviour was directly related to the hardness and densification, with optimally prepared samples of AM titanium having the highest hardness and wear resistance, and samples with cracking or internal porosity having lower hardness and wear resistance. Fretting wear of laser additive manufactured Ti-6Al-4V was tested and compared to additively manufactured metals by Kumar and Kruth [55], under the conditions of amplitude 200µm, frequency of 10Hz and 10,000 cycles, with loading between 2 and 6N. Ti-6Al-4V was found to have the lowest wear resistance of the AM materials tested, which included two stainless steels, a tool steel, and a cobalt-chrome alloy, but a comparison with Ti-6Al-4V produced by other manufacturing methods was not provided.

**Surface Roughness**
Surface roughness in AM is highly variable and depends on the location in the part (with overhanging areas generally having the highest roughness), the process parameters[56] and the raw
material characteristics[1]. This is particularly true for powder size which can change over multiple builds as material is recycled, a feature of all materials manufactured by AM and not just Ti-6Al-4V[57]. Table 2 provides some example values in the AM of Ti-6Al-4V where both upward facing and downward facing (unsupported overhang) values are shown; generally the downward facing roughness measurements are larger as the molten material sits on top of a bed of powder in these cases and may sink into it.

Pyka et al [38] tested a combination of chemical etching and electrochemical polishing as a method of surface finishing. This was reported to be particularly effective on reducing roughness on the downwards facing surfaces. Furumoto et al [59] used free abrasive grains to finish internal channels. For both these references, Table 2 gives two sets of surface roughnesses; those for the material ‘as built’, and those for material which had undergone a degree of surface finishing.

Table 2. Roughness data for AM materials before and after surface finishing. Data for upward facing and downward facing (unsupported overhang) are shown.

<table>
<thead>
<tr>
<th>Reference</th>
<th>Surface roughness ($R_a$) [µm]</th>
<th>Surface roughness ($R_z$) [µm]</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Upward</td>
<td>Downward</td>
</tr>
<tr>
<td>Pyka et al [58]</td>
<td>7</td>
<td>12</td>
</tr>
<tr>
<td>After surface finishing</td>
<td>5.5</td>
<td>6.5</td>
</tr>
<tr>
<td>Furumoto et al [59]</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>After surface finishing</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Cooper et al [60]</td>
<td>3.96</td>
<td>17.5</td>
</tr>
</tbody>
</table>

Fracture toughness and impact testing

Van Hooreweder [61] showed that LAM-built Ti-6Al-4V has a lower fracture toughness than a sample produced by vacuum arc remelting (VAR), with $K_{IC}$ for the additive manufactured material being $52.4 \pm 3.48$ MPa m$^{1/2}$, approximately 75% of that for the VAR material ($69.98 \pm 3.53$ MPa m$^{1/2}$). The presence of the martensitic microstructure, due to the high cooling rates, was suggested to be the primary cause for the decrease.

Similarly, Charpy impact testing [62] on as-built LAM Ti-6Al-4V showed that the toughness was lower than reported values for cast material, with an impact energy of 11.5 J for as-built LAM material compared to 15 J for cast Ti-6Al-4V. The heat-treatments examined did not improve the
toughness; for example an annealing cycle at 735°C for 2 hours reduced the impact energy to around 10J, and stress-relief at 595°C for 3 hours decreased it further to around 7.5J. Again, the larger amount of martensitic, brittle microstructure present in the stress relieved cycles was suggested as a cause of the lower toughness.

**Fatigue properties**

For aerospace and medical application the fatigue behaviour of the material is a vital part of material selection and component design. While tensile and toughness data are relatively easy to characterise the measurement of fatigue data can be a lengthy and onerous task. While there is a wealth of historical data of fatigue properties of wrought alloys, there is significantly less on those manufactured by AM. It is difficult to directly compare fatigue data from different sources, as the choice of method, stress ratio (R-ratio) and specimen preparation will affect the results greatly; however a summary of relevant papers on Ti-6Al-4V is made below.

**Stress Ratio (R-ratio)**

In fatigue testing, the stress ratio is defined as minimum stress/maximum stress. A stress ratio of zero indicates a cycle that varies from zero stress to a maximum tensile stress; a stress ratio between 0 and 1 indicates a cycle where the specimen is always under tensile load, e.g. 0.1 indicates that the minimum stress is set to 10% of the maximum stress. Where the stress is perfectly reversed (compressive stress equal to tensile stress), the stress ratio is equal to -1.

**Types of fatigue loading**

Fatigue loading methods found in the literature for additive manufacturing of titanium alloys include axial, torsional, rotating bending, three-point bend [63] and plane bending [48]. Axial fatigue testing uses a sample geometry similar to that used for standard tensile tests. A range of stress ratios (common values include -1, 0, 0.1 and 0.5) can be achieved. In rotating bending fatigue testing a bending moment is placed on a sample being rotated at high speeds (3000rpm or more). This reduces the time required for testing, particularly when high cycles tests with up to $10^7$ cycles or more are required, but the only possible stress ratio is -1. Due to the nature of the loading, the stress is highest at the surface of the gauge length.

**Specimen preparation**

Specimen preparation plays a large part in determining the fatigue life of a component. Fatigue specimens may be notched to examine the effect of stress concentrations on the fatigue life, which varies depending on the material [44]. Surface finish also has a large effect on fatigue life, as a
rough surface acts as a source of defects to initiate fatigue. For this reason, a machined, polished finish is required by many fatigue testing standards, for example ISO 1143 [64] recommends for rotating bending fatigue testing a maximum $R_z$ of 0.2µm. As shown in Table 2 additively manufactured specimens have a rough surface finish in the as-built condition. In addition it is not necessarily possible to obtain a uniform level of surface finish when parts built have complex shapes. Mower et al. [39] recently examined a variety of polishing methods on fatigue properties.

In order to improve surface roughness and fatigue properties alternate finishing methods (e.g. [59]) for additive manufacturing materials have been described in the literature but not in combination with fatigue testing. Some surface finishing methods, such as shot peening, are also known to increase fatigue strength by inducing compressive stresses at the surface of the sample to prevent the initiation of fatigue failure, and these have been tested by Edwards et al [45] for electron beam melted Ti-6Al-4V.

Residual stresses are often present in as-built AM materials [66], particularly those produced with laser melting. These could potentially lower the fatigue life; however as ‘stress-relief’ is typically carried out before parts are removed from the baseplate so there is little data on the impact of residual stresses.

Hot Isostatic Pressing (HIP) has the potential to remove or reduce closed porosity and is often described as improving fatigue behaviour in additive manufactured metals [65] which is consistent with its effects on castings [45]. The HIP process can also influence the microstructure, for example recently Mower et al. [39] subjected Ti-6Al-4V to HIP at 900°C at 102MPa for two hours and the $\alpha$ lath thickness after HIP was greater than that of the non-HIPed alloy (see Fig 4d). Facchin et al [48] also observed that HIP slightly coarsens the microstructure, resulting in a small decrease of strength and an increase in elongation at fracture; fatigue strength was more significantly improved. Qui et al. [38] reported that HIP (900°C at 103MPa/4h and furnace cooling) closed porosity and transformed the martensitic as-built structure into $\alpha$ and $\beta$ with a subsequent improved ductility, with a reduction in strength.

Due to these various factors, it is difficult to directly quantify differences in the literature where they have used different methods of sample preparation, fatigue testing, or both. For a comparison between additively manufactured and wrought material, $R = -1$ data from Hollander et al [40], Mower et al [39] and Leuders et al [67] was plotted along with representative data for annealed wrought material taken from the MMPDS [27]. No heat treatment other than stress relief was used
in Hollander, and points from Leuders where heat treatment was applied are labelled. Points from Leuders et al. [67] represent mean fatigue life rather than the results of individual tests; in this testing none of the samples that underwent HIP treatment failed under this stress after 2 x 10^6 cycles at which point the tests were stopped. Further staircase testing with 18 HIP samples showed the fatigue limit to be 615MPa. All data present in Fig. 7 is from test pieces with a machined finish, and without notches \((K_t = 1)\).

However, these general factors can be observed:

i. surface finish is an important issue, with significantly increased fatigue strength measured where the samples are built as bars or cylinders and machined down into shape compared to samples produced directly with the as-built surface.

ii. HIP generally improves the fatigue properties, possibly by closing sub-surface porosity which might otherwise act as a fatigue initiator or by providing a heat-treatment which converts the microstructure to one which is more resistant to crack growth, or a combination of both factors. Porosity and lack of fusion create defects for initiation of fatigue and influences both the mean fatigue life and the variability of fatigue data. For example, Tammas-Williams et al. [68] examined Ti–6Al–4V samples built with Arcam Selective Electron Beam Melting (SEBM) that typically exhibit tensile properties comparable to those of wrought material but without post-manufacture treatments, such as HIP, the high cycle fatigue was observed to exhibit a high degree of large scatter. Such defects generally result from gas pores a lack of fusion defects at the interfaces between each laser pass and can be more damaging because of their larger size (e.g. 200 \(\mu\)m). Work is needed in sufficient detail to allow reliable statistical relationships to be developed with process variables for AM applications.

\textbf{Table 3. Summary of published fatigue data on Ti-6Al-4V alloys by AM.}

<table>
<thead>
<tr>
<th>Reference</th>
<th>Year</th>
<th>Material</th>
<th>Method</th>
<th>Loading Type</th>
<th>Stress Ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>Abe \textit{et al} [13]</td>
<td>2003</td>
<td>CP Ti</td>
<td>LAM</td>
<td>Torsional</td>
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<td>Santos \textit{et al} [14]</td>
<td>2004</td>
<td>CP Ti</td>
<td>LAM</td>
<td>Torsional</td>
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<tr>
<td>Hollander \textit{et al} [40]</td>
<td>2006</td>
<td>Ti-6Al-4V</td>
<td>LAM</td>
<td>Rotating Bending</td>
<td>-1</td>
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<tr>
<td>Facchini \textit{et al} [48]</td>
<td>2009</td>
<td>Ti-6Al-4V</td>
<td>EBM</td>
<td>Plane Bending</td>
<td>Not specified</td>
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<tr>
<td></td>
<td>Year</td>
<td>Material</td>
<td>Process</td>
<td>Test Method</td>
<td>Stress, S_{max} (MPa)</td>
</tr>
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<td>------</td>
<td>----------------</td>
<td>---------</td>
<td>--------------------</td>
<td>------------------</td>
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<tr>
<td>Chan et al [63]</td>
<td>2012</td>
<td>Ti-6Al-4V</td>
<td>LAM /EBMizin</td>
<td>3-point bending</td>
<td>0.1</td>
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<tr>
<td>Gong et al [44]</td>
<td>2012</td>
<td>Ti-6Al-4V</td>
<td>LAM</td>
<td>High Cycle Vibration</td>
<td>0.1</td>
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<tr>
<td>Leuders et al [67]</td>
<td>2012</td>
<td>Ti-6Al-4V</td>
<td>LAM</td>
<td>Axial</td>
<td>-1</td>
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<td>Thöne et al [50]</td>
<td>2012</td>
<td>Ti-6Al-4V</td>
<td>LAM</td>
<td>Axial</td>
<td>-1</td>
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<tr>
<td>Edwards et al [65]</td>
<td>2013</td>
<td>Ti-6Al-4V</td>
<td>EBM</td>
<td>Axial/other</td>
<td>-0.2 or -0.39</td>
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<td>Brandl et al [69]</td>
<td>2011</td>
<td>Ti-6Al-4V</td>
<td>EBM</td>
<td>High Cycle Vibration</td>
<td>0.1</td>
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<td>Mower et al. [39]</td>
<td>2016</td>
<td>Ti-6Al-4V</td>
<td>LAM</td>
<td>Rotating beam</td>
<td>-1</td>
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**Fig 7: Fatigue life for additively manufactured Ti-6Al-4V versus annealed wrought**

**Conclusions**

This review has described powder bed additive manufacturing methods for Ti-6Al-4V. The following points can be made.
i. Material properties of components manufactured by AM cannot be assumed to be equivalent to wrought and there are distinct differences in the microstructures.

ii. Tensile mechanical properties, such as yield or ultimate tensile strength, are comparable and sometimes better than wrought. The properties depend on build method, build direction and heat treatment and tend towards high strength and low ductility behaviour.

iii. There is limited information on fracture toughness and wear of Ti-6Al-4V manufactured by AM.

iv. There is little published information on the fatigue properties considering its importance in design for critical components. There is a considerable influence of factors such as surface roughness, the way in which the AM part is built, thermal history during the build process and even orientation.

v. The design freedom offered by AM may mean that the final design cannot be post-machined due to internal surface structures. Nevertheless, surface factors must be taken into account and this can be a significant barrier to industrial and commercial acceptance.

vi. Future research certainly required more effort of mechanical properties of AM parts consistency and fatigue testing which takes into account the wide variation in surface finish/specimen preparation, thermal profile, microstructure, defects and providing accurate measure of surface factors and the effects of heat treatments and HIP on tensile properties and fatigue.

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42. *Implants for surgery. Metallic materials. Wrought titanium 6-aluminium 4-vanadium alloy.*


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