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Mechanical Properties of Ti-6Al-4V Additively Manufactured by Electron Beam Melting

A. Kirchner[1], B. Klöden[1], T. Weißgärber[1], B. Kieback[1], A. Schoberth[2], S. Bagehorn[2], D. Greitemeier[2]

[1] Fraunhofer Institute for Manufacturing Technology and Advanced Materials IFAM, Winterbergstrasse 28, 01277 Dresden, Germany
[2] Airbus Group Innovations, 81663 Munich, Germany

Abstract

The application of additively manufactured titanium components is attractive due to a number of potential benefits. The high freedom of design enables the fabrication of structurally optimized, lightweight parts. Complex geometries may serve additional functions. The widespread use of additive manufacturing could revolutionize logistics by dramatically reducing lead time and allowing customized production. Manufacturing near net shape parts reduces scrap of expensive material. Together with the economy of scale this is bound to reduce part costs.

Especially for the application in safety relevant parts certainty about static and fatigue strength is critical. A challenge arises from complex influences of built parameters, heat treatments and surface quality. Ti-6Al-4V specimen built by electron beam melting (EBM) were subjected to heat treatments adapted to various employment scenarios. The results of tensile and fatigue testing as well as crack propagation will be compared to conventionally manufactured titanium. The mechanical behavior will be correlated to the microstructural evolution caused by the heat treatments.

1. Introduction

Electron beam melting (EBM) is a powder bed based process for additive manufacturing [1, 2]. Optimal process parameters for the fabrication of Ti-6Al-4V by EBM have been investigated previously [3, 4]. The aim was to manufacture parts with full density and understand changes in alloy composition. The scope of the present work is to elucidate the influence of various subsequent heat treatments on the tensile, high-cycle fatigue and crack propagation properties.

2. Materials and methods

The Ti-6Al-4V powder with a nominal particle size distribution of 45 - 105 µm was supplied by TLS Technik Bitterfeld. Chemical analysis revealed a chemical composition of 6.4 % Al, 4.2 % V, 0.2 % Fe and 1200 ppm O (all by mass).

EBM was performed on an Arcam A2x machine operating at a vacuum below 2×10^{-2} mbar and an acceleration voltage of 60 kV. The remaining gas consisted predominantly of He, which was released into the chamber to reduce electrostatic charging. At the beginning of a build process, a steel start plate was heated to 730 °C as measured by thermocouple underneath. The addition of each layer started by raking a powder layer of 50 µm nominal thickness. The fresh powder was preheated by fast scanning with a defocused electron beam. After solidifying the part contours, the specimen interiors were molten using a focused electron beam (focus offset 3 mA) with 15 mA current and a speed function of 98. This relates to a beam speed of 3.2 m/s and a line energy of 280 J/m. The distance between parallel scan lines was 100 µm. Between each layer the principal scan direction was alternated from x-axis to y-axis. The parts were oriented in the build space such that the direction of subsequent mechanical testing coincided with the build direction (z-axis).

The sample blanks were subjected to heat treatments between 650 °C and 1050 °C, marked with HT 1 to HT 6 in ascending order. HT 2 to HT 6 were hot isostatic pressing (HIP) treatments with pressures up to 200 MPa and durations ranging from 0.2 h to 2 h aiming to remove residual porosity. Test samples were machined from the blanks. Tensile specimen were 91 mm long and 2 mm thick with a 6 mm wide gage section according to EN 2002 type 2. Specimen for high-cycle fatigue test were 90 mm long and 2.5 mm thick with a minimum width of 7 mm as specified in EN 6072 type 1. Samples to determine crack growth rates according to ASTM E647 were 50 x 48 x 12 mm³.

Tensile tests were performed on a Zwick 1466 with 3 specimen per test at a strain rate of 0.5 %/min (to stress values >YTS) and 2 %/min (beyond). High-cycle fatigue measurements were carried out on Mikrotron and Fractronic testing machines with a stress ratio of 0.1 at a frequency of 88 to 146 Hz. One specimen per heat treatment and stress level was used. In case the fatigue life of a sample exceeded $2 \cdot 10^7$ cycles the test was stopped (run-out). Within one test series, the test was stopped after $1 \cdot 10^7$ cycles. Fatigue crack growth tests were performed on a Schenck PSA with a stress ratio of 0.1 at a frequency of 20 Hz on one specimen per test parameter set and 3-8 specimens per heat treatment. All mechanical tests were performed at room temperature.

3. Results and discussion

The tensile experiments yielded Young's moduli comparable to wrought Ti-6Al-4V. The highest value of 122.8±2.4 GPa was obtained for HT 1, while HT 6 / HIP exhibited the lowest value of 113.6±0.3 GPa. This is consistent with fully dense material showing little or no texture. The principal results of the quasi-static mechanical tests are presented in Figure 1. The ultimate tensile strength (UTS) of 1022.7±5.6 MPa and total elongation of 14.7±0.6 % for HT 1 is practically identical to the optimum as-built condition [4]. Up to a certain point, heat treatments at higher temperatures and the application of pressure manage to retain UTS and slightly increase ductility (strain to failure A). HT 3 / HIP exhibits an UTS of 1028.6±7.6 MPa and a strain to failure of 16.3±0.6%. The treatment at the highest temperature (HT 6 / HIP) results in a UTS drop to 910.4±2.4 MPa. The yield tensile strengths (YTS), determined using 0.2 % offset strain, follow the trend of the UTS. The measured mechanical properties of all heat treated EBM specimen exceed the requirements for Ti-6Al-4V set by AMS4928 [5]. In comparison to untreated Ti-6Al-4V specimen fabricated by selective laser melting (SLM) YTS and UTS are lower but ductility is much higher [6, 7].

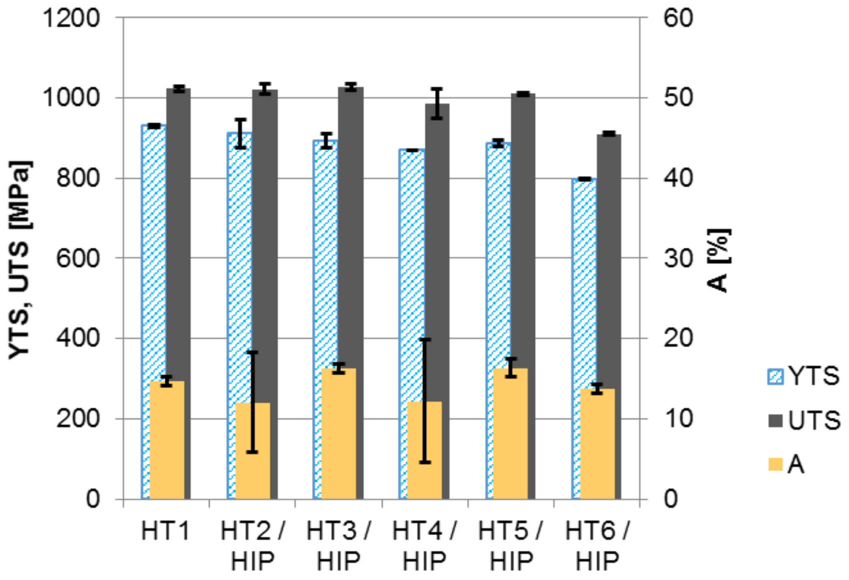

Figure 1: Yield tensile strength YTS, ultimate tensile strength UTS, and elongation at rupture A of heat treated specimen built by EBM.

S-N-curves representing the fatigue behavior of heat treated EBM-fabricated Ti-6Al-4V are shown in Figure 2. The specimen HT 1 heat treated at the lowest temperatures reached the fatigue limit of more than 10^7 cycles at 350 MPa. This agrees well with the published value of 340 MPa for as-built samples [7]. Reducing any remaining porosity by HIP first lifts the "knee" above 10^6 cycles as visible in sample HT 2 / HIP. HIP treatments at higher temperatures (HT 3 to HT 5) lift the S-N-curve to higher stress levels with a fatigue limit at approximately 600 MPa. In case of HT 3 / HIP a run-out at a stress level of even 650 MPa was observed. Shorter lifetimes at higher stress levels were not investigated due to the risk of plastic deformation and strain hardening at stress levels close to the YTS. A further temperature increase leads to a reduction in HCF stability as visible in HT 6 / HIP. In comparison the fatigue limit of chemically polished double vacuum-arc remelting forging stock is given as 550 MPa [8]. Very similar values of 500 to 550 MPa are stated for machined SLM-fabricated Ti-6Al-4V [7, 9, 10].

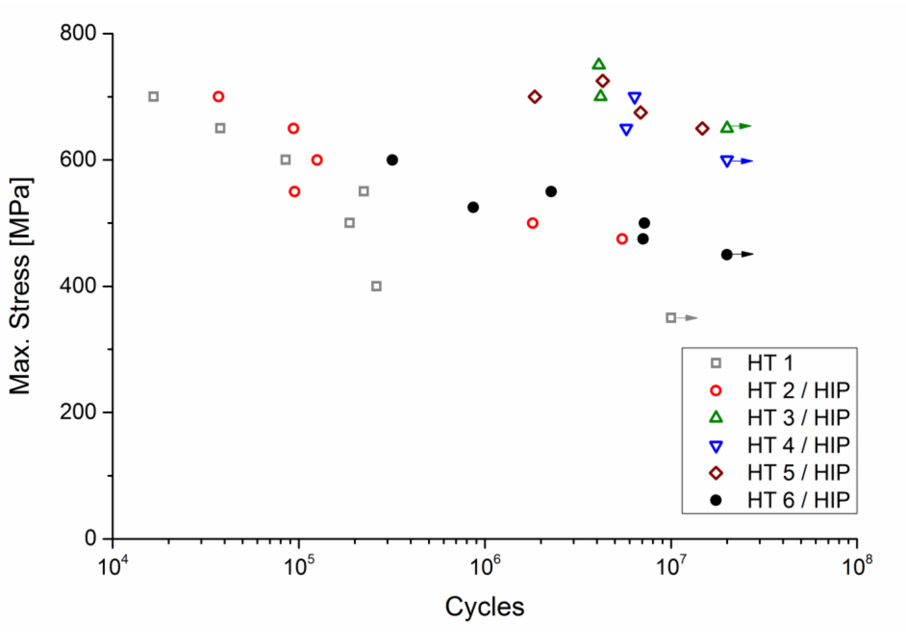

Figure 2: High-cycle fatigue behavior of heat treated specimen built by EBM. Data points with arrows mark specimen without rupture (run-out).

The crack propagation rates of the EBM-built samples are plotted in Figure 3. The results coincide with each other except for HT 6 / HIP. This implies that the temperature increase from HT 1 to HT 5 / HIP does not influence crack propagation noticeably. At a stress intensity range ΔK of 20 MPa√m a crack growth rate of $2.5 \cdot 10^{-4}$ mm/cycle was observed. This value is in good agreement with the data for solution treated and overaged Ti-6Al-4V exhibiting a bimodal microstructure [11]. The data presently available shows that the plane strain fracture toughness K_{IC} exceeds 50 MPa√m for all heat treatments.

HT 6 causes a distinct retardation of crack growth at ΔK below 20 MPa√m. From data points at very low rates of crack growth a threshold of 10 MPa√m can be estimated. This is higher than the threshold value of 3.5 MPa√m cited for SLM-fabricated Ti-6Al-4V in the as-built state [10].

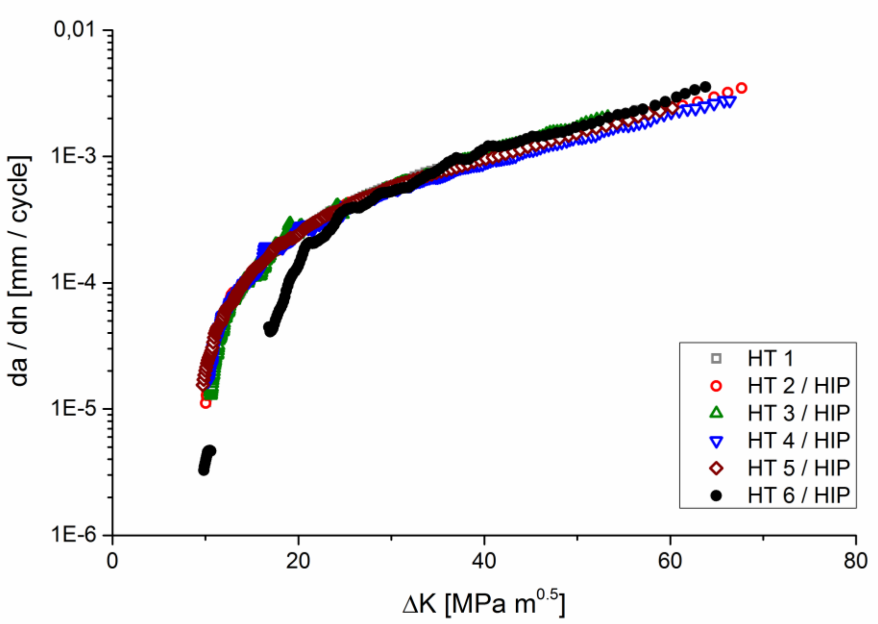

Figure 3: Crack growth of heat treated specimen built by EBM.

The microstructure of three selected specimen after polishing and etching is depicted in Figure 4. The structure of HT 1 tempered at the lowest temperature is a very fine basket-weave pattern. The maximum thickness of the α-platelets is about 3 µm. With respect to the as-built state, no changes are discernible [4]. Upon increasing temperature the α-phase thickness grows slightly, especially along the prior β grain boundaries. Sample HT 3 / HIP contains a few α-platelets up to 8 µm thickness. This is in accordance with very similar tensile properties presented above. Specimen HT 6 / HIP tempered at the highest temperature exhibits a distinct coarsening of the α-phase with plate thickness up to 15 µm and starting globularization. This explains the decrease in tensile strength since the mechanism of plastic deformation in Ti-6Al-4V at low rates was found to be planar slip in the α-grains [12].

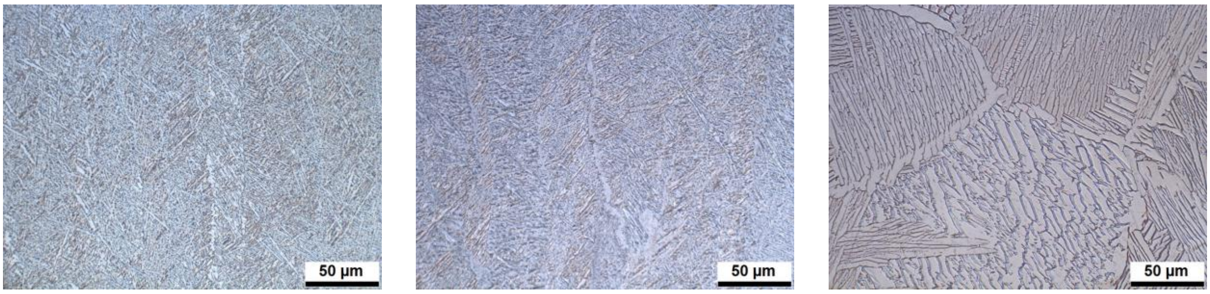

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Figure 4: Optical micrographs of heat treated samples built by EBM. (a) HT 1. (b) HT 3 / HIP. (c) HT 6 / HIP.

4. Conclusion

The presented results highlight the potential for the optimization of the mechanical properties of EBM-fabricated Ti-6Al-4V by heat treatments. Since the powder bed and the parts therein are kept at around 700 °C during the EBM build process and cooled down slowly, dense specimen combine UTS of more than 1000 MPa with good ductility. Therefore heat treatments at temperatures above 700 °C offer only small improvements in static tensile properties. In contrast, significant increases in fatigue resistance can be achieved by combining HIP and thermal treatments. In that way Ti-6Al-4V material additively manufactured by EBM can match fatigue performance of forging stock.

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