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29 the aim to analyze the influence of the specific microstructure produced by the
30 DMLS on the mechanism of the propagation of long cracks.

31

32 *Keywords:* Titanium alloys, Ti6Al4V, Fatigue crack growth, Threshold value of
33 stress intensity factor, Direct metal laser sintering

34

35 **1. Introduction**

36 The rapidly developing methods of additive manufacturing (AM) over the past
37 decade brought high expectations in industry. This technology enables the
38 manufacture of metallic components of very complicated shape directly from the
39 computer models [1, 2] without time consuming machining and assembly processes
40 or preparation of casting molds. Direct metal laser sintering (DMLS) is one of the
41 AM methods, which has a very promising potential [3, 4]. The components are built
42 through localized melting of gas atomized powder by a concentrated laser beam. The
43 powder is locally rapidly melted and subsequently rapidly solidified when the laser
44 beam is moved to a neighboring area [5]. The advantage of the DMLS procedure is
45 obvious in the development and optimization process of new components or in the
46 case of production of small series, e.g. for biomedical applications [6, 7].

47 The reliable application of components produced by DMLS technology
48 requires guaranteed mechanical properties. The static performance of many DMLS
49 produced materials was found to be good comparable with conventionally
50 manufactured ones [8, 9]. The problem of porosity and large defects of
51 microstructure, which can deteriorate the tensile properties, has been in the majority
52 of cases solved by optimization of the DMLS process parameters and the post-
53 processing treatment, e.g. HIPping [10, 11].

54 There is an increasing interest of industry in the production of parts, which
55 have to fulfill severe demands on mechanical properties and high reliability also
56 under cyclic loading. The damage tolerant design demands a knowledge of fatigue

57 crack growth behavior. The propagation of long cracks, particularly in the threshold
58 region, is influenced by the material microstructure. The high number of DMLS
59 process parameters like laser power, building layer thickness, scanning strategy etc.
60 influences the resulting microstructure and consequently the fatigue crack growth
61 behavior [12, 13].

62 Ti6Al4V is the most commonly used Ti alloy due to its outstanding
63 engineering properties. The advantage of this alloy consists in good mechanical
64 properties, low density and good corrosion resistance, e.g. [14, 15]. The relatively
65 high cost, however, predetermines this material for applications with specific
66 requirements, in aircraft industry, space technology or for manufacture of implants
67 for human body. There are numerous studies focused on Ti6Al4V alloy
68 manufactured by DMLS, e.g. [7, 16, 17], however, the main attention was paid to the
69 tensile properties. The behavior under cyclic loading and particularly the fatigue
70 crack growth resistance has been investigated in a much lower extent. The threshold
71 stress intensity factor range of the heat treated Ti6Al4V alloy (in the 800 and
72 1050 °C range) for the stress ratio $R = 0.1$ and the crack growth perpendicular to the
73 build direction is of about $4 \text{ MPam}^{1/2}$ [17]. The as-fabricated DMLS material exhibits
74 a much lower value $1.4 \text{ MPam}^{1/2}$ [17]. Moreover, the threshold values were found to
75 be slightly different for the crack growth parallel and perpendicular to the build
76 direction. In the Paris region the crack growth rate is comparable with those
77 determined on conventionally produced material. Moreover, a characteristic feature
78 of as-produced DMLS alloy without heat treatment is a considerable scatter of the
79 experimentally determined crack growth rate data for a given value of the stress
80 intensity factor range [17, 18].

81 The long crack growth is influenced by material microstructure. Its effect is
82 most pronounced in the threshold region. Investigations on the influence of
83 microstructure on the crack growth rate and thresholds in DMLS materials indicate
84 that there is an effect of the direction of material building [17, 19], because the

85 microstructure and the residual stress state are different in different directions with
86 respect to the build direction.

87 This study is aimed at the experimental investigation of the growth of long
88 fatigue cracks in Ti6Al4V alloy in different directions with respect to the DMLS
89 build direction. The heat treatment of the alloy consisted only from the stress
90 relieving at 380 °C/8 h in argon. The growth of long crack was studied in relation to
91 the specific microstructure produced by the DMLS technique.

92

93 **2. Experimental**

94 2.1. Material

95 The specimens for tensile and fatigue crack growth testing were built from
96 EOS titanium Ti64 alloy powder which has been optimized especially for processing
97 on EOSINT M270 systems. Each equipment manufacturer specifies the metal
98 powder to be used with a particular system. Powder particles used here were
99 spherical in shape and fulfill the requirements of ISO 5832-3, ASTM F1472 and
100 ASTM B348 standards as regards the maximum concentration of impurities. The
101 powder particles are shown in Fig. 1a and its size distribution in Fig. 1b. The
102 standard and real chemical compositions of the powder are listed in Table 1. The real
103 chemical composition was determined by SEM equipped with EDX analysis.

104

105 2.2. Direct metal laser sintering process

106 Specimens for testing were produced on an EOSINT M270 system (EOS
107 GmbH) directly from three-dimensional CAD design data. The system was equipped
108 with a 200 W ytterbium fiber laser. A platform for material dispensation and a build
109 platform were located in a build chamber filled with argon to avoid oxidation. For
110 building of specimens the operating parameters of the system were set to laser power
111 200 W and the laser speed to 800 mm^s⁻¹. The layer thickness was 30 μm, the bead
112 hatch distance 100 μm and the multidirectional scanning strategy of rotation between
113 layers of the EOS DMLS system was applied. The build plate was heated to 80 °C.

114

115

116 2.3. Specimens

117 Tensile tests were conducted on flat specimens with the cross-section
118 4 x 6.25 mm and a gauge length 32 mm, Fig. 2. The longitudinal axes of the
119 specimens were in three orientations in respect to the build direction; the particular
120 types of specimens were denoted as *a*, *b* and *c*. The axis of the specimen *c* is parallel
121 to the build direction *Z*. The loading axis of the specimens with orientations *a* and *b*
122 are perpendicular to the build direction *Z*. The surface of the gauge length remained
123 in the as-build state.

124 The specimens for fatigue crack growth determination were manufactured
125 according to ASTM E 647-08 Standard. The characteristic specimen dimension *w*
126 (i.e. distance between pin center and back surface) was 30 mm and the thickness 6
127 mm. The CT specimens were manufactured in such a way that the macroscopic
128 fatigue crack growth direction was parallel to the build direction *Z* in orientation *a*
129 and perpendicular to *Z* for the orientations *b* and *c*, Fig. 3.

130 Because the lateral as-build surfaces were relatively rough and not suitable for
131 optical measurement of the crack growth, they were finely ground and carefully
132 mechanically polished after heat treatment.

133 Stress relieving heat treatment is an important step in the production of parts
134 made by DMLS. A stress relieving heat treatment at 380 °C for 8 hours was carried
135 out in this study. The reason for this type of the heat treatment is substantiated by the
136 idea to obtain crack growth data for the “as build” material. The heat treatment at
137 higher temperatures which changes the material microstructure and related
138 mechanical properties was deliberately not performed. To protect the surface from
139 oxidization, the heat-treatment was conducted in argon atmosphere.

140

141 2.4. Mechanical testing, investigation of microstructure, fractography

142 A servohydraulic testing machine MTS 810 with an extensometer having 25

143 mm gauge length was used for experimental determination of the tensile properties.
144 Test methods fulfilled the ASTM 8/E 8M-08 Standard. The loading rate was 0.01
145 mms^{-1} .

146 The fatigue crack growth testing was conducted under load control on a
147 Roell/Amsler HFP 5100 resonant testing machine. The loading asymmetry
148 characterized by the ratio R (minimum to maximum load in a loading cycle) was 0.1.
149 The loading frequency was in the interval 80 - 50 Hz and was dependent on the crack
150 length, i.e. on the stiffness of the specimen. Both the tensile tests and the fatigue
151 crack growth measurement were performed at ambient laboratory conditions.

152 The length of the fatigue crack was optically measured on both lateral sides of
153 specimens by means of two CCD cameras situated on travelling supports. The crack
154 length was determined with the 0.01 mm accuracy.

155 The load shedding procedure was used after the crack initiated from the
156 starting notch. The load amplitude was reduced at the constant R ratio in such a way
157 that the crack growth rate of the order of 10^{-7} mm/cycle was reached. The
158 determination of the crack growth rates was carried out at the constant load
159 amplitude and fulfilled the requirements of the ASTM E647-08 Standard.

160 The microstructure produced by DMLS on three metallographic sections of
161 each specimen was observed with a light microscope Zeiss Axio Observer Z1M after
162 conventional preparation and etching (i.e. 10% HF). The fracture surfaces were
163 studied in a scanning electron microscope Tescan LYRA 3 XMH FEG/SEM. The
164 determination of the fracture roughness profiles in the threshold and in the Paris
165 regions were optically measured using image analysis software.

166

167 **3. Results**

168 Fig. 4 shows an example of the stress-strain curves obtained by tensile tests on
169 three specimens with different mutual orientations of the build direction and loading
170 direction. The build orientation influences slightly the tensile properties, see Table 2.
171 The ultimate tensile strength σ_{UTS} of the specimens with the orientations *a* and *b*, is

172 1255 ± 21 and 1235 ± 11 MPa respectively and the error scatterbands are
173 overlapping. The σ_{UTS} of specimens with orientation c , i.e. the loading axis is
174 perpendicular to the build layers, is lower, namely 1195 ± 9 MPa. Similarly small is
175 the influence of the orientation on the yield stress $\sigma_{0.2}$ and the elongation to rupture.

176 The experimentally determined fatigue crack growth rate dependence on the
177 stress intensity factor range ΔK is shown in Fig. 5. The data exhibit a large scatter.
178 Nevertheless, there is no detectable influence of crack direction with respect to build
179 orientation on the crack propagation rate and on the crack propagation threshold.
180 With decreasing ΔK the crack rate decreases and tends to reach the threshold value
181 $\Delta K_{th} = 3.7 \text{ MPam}^{1/2}$. The crack growth rate in dependence on the stress intensity
182 factor range ΔK can be generally described by the equation [20]

$$183 \quad da/dN = A(\Delta K^m - \Delta K_{th}^m). \quad (1)$$

184 Taking into account the threshold value $3.7 \text{ MPam}^{1/2}$, the experimental points can be
185 well approximated by the equation

$$186 \quad da/dN = 3 \times 10^{-8} \Delta K^{3.0} - 1.52 \times 10^{-6}, \quad (2)$$

187 which is shown in Fig. 5 by the full line.

188 Observation of microstructure on three mutually perpendicular planes was
189 made both for tensile and CT specimens. The microstructure of tensile and CT
190 specimens is identical and exhibits some degree of directionality. The microstructure
191 on z-x and z-y (lateral) planes that are parallel to the build direction, Fig. 6a, is
192 characterized by primary columnar β grains, which transformed during cooling to
193 fine needles of α' -martensite, Fig. 6b and Fig. 7b. This type of microstructure
194 corresponds to the microstructure typical of as-fabricated products [21, 22]. The
195 temperature of the stress relieving heat treatment (380 °C/8 h) was too low to
196 transform the original microstructure. The elongated β grains having the length of
197 several building layers grew in the build direction and exhibit a texture. Black spots
198 on the optical micrograph, Fig. 6a, correspond to the regions of locally substantially
199 finer microstructure. These regions are elongated in the direction perpendicular to the

200 build direction. The finer microstructure is a consequence of high local internal
201 thermal gradients and high cooling rates caused by the laser beam passes. Moreover,
202 elongated microshrinkages with sharp edges were locally present in the black spots.
203 The microstructure, Fig. 7, observed on the horizontal plane (x-y) perpendicular to
204 the build direction shows different features when compared to the lateral planes.
205 Fig. 7a shows polyhedral grains corresponding to the perpendicular sections of
206 columnar primary β grains. The size of these grains is defined by the width of the
207 columnar grains.

208 The fatigue cracks propagate in DMLS produced material in a transgranular
209 way. Examples of fatigue fracture surface profiles of the specimen *c* together with
210 the microstructure are shown in Fig. 8. The transgranular character of the crack
211 propagation was observed both for the threshold region, Fig. 8a, where the crack rate
212 is low and also for the Paris region, Fig. 8b. The propagating crack does not interact
213 with the grain boundaries of prior β phase. The propagation takes place by damaging
214 of the very fine needle structure of acicular α' -martensite at the crack tip. The
215 fracture profiles are flat with local small sharp asperities. No significant difference in
216 the fatigue fracture surface profiles was found for the three specimen orientations *a*,
217 *b* and *c*.

218 The typical fracture surface appearance does not change significantly with increasing
219 crack propagation rate from the threshold region up to the crack rate of the order $1 \times$
220 10^{-4} mm/cycle. Fig. 9 brings an example of the fracture surface of the specimen *c*.
221 The observed independence of the fracture surface appearance on the fatigue crack
222 rate is in accordance with the independence of the fracture surface roughness *Ra* on
223 the crack growth rate (*Ra* is the arithmetic average of the absolute roughness profile).
224 The *Ra* values for three types of specimens *a*, *b* and *c* in the threshold region and in
225 Paris region are given in Table 3. It can be seen that there is no substantial difference
226 among the *Ra* for orientation *a* and *b* and there is also no substantial difference of
227 fracture surface profile roughness in threshold and Paris region. The roughness of
228 the fracture surface corresponding to the specimen *c* seems to be also independent on

229 the crack growth rate, however it is higher than that of specimens *a* and *b*. The
230 difference in the surface roughness can be explained by the fact that the fatigue crack
231 propagates parallel to the building layers in the specimen *c*, whereas it propagates
232 across the layers in the specimens *a* and *b*.

233

234 **3. Discussion**

235 The tensile properties are only very weakly dependent on the build orientation,
236 i.e. on the directionality of microstructure. Somewhat higher are both the σ_{UTS} and
237 $\sigma_{0.2}$ for the orientation when the loading is applied parallel to the build layers and
238 marginally worse when the loading is perpendicular to the build layers. The
239 explanation may be based on the influence of small and rare microshrinkages
240 elongated in the building plane. The inherent α' -martensite fine needle
241 microstructure, Figs. 6b and 7b, does not exhibit any specific directionality. The
242 tensile properties are good comparable with those of conventionally manufactured
243 alloy heated for 1 h at 1050 °C and quenched in water, as it can be seen from Fig. 4,
244 where comparison with tensile curve published by Fan et al. [23] is shown. The
245 presented microstructures of DMLS after stress relieving treatment and
246 conventionally prepared material after heat treatment with martensitic structure are
247 comparable.

248 The experimental points presented in Fig. 5 correspond to the crack growth
249 measurement performed solely under constant load amplitude, i.e. under increasing
250 ΔK . This is a difference to the data published by the present authors in [24], where
251 the experimental points were collected both under increasing ΔK and also decreasing
252 ΔK (load shedding procedure). The data scatter in Fig. 5 in this work is somewhat
253 lower than the scatter of experimental points in [24]. Note that the data determined
254 under decreasing load conditions were valid according to the ASTM E647-08
255 Standard. The values of *m* and *A* parameters in eq. (1) fitted to the data
256 corresponding to the constant amplitude loading are only marginally different from

257 that published in [24]. Similar conclusion can be made as regards the threshold value
258 ΔK_{th} . The small difference in the crack growth rate scatter for measurement covering
259 both the constant and decreasing loading may indicate high sensitivity of the DMLS
260 material to the changes of loading conditions.

261 The crack growth data for as-fabricated DMLS Ti6Al4V alloy published by
262 Becker et al. [18] are replotted together with our data in Fig. 10. The crack growth
263 measurement was performed on CT specimens with characteristic dimension $w =$
264 25 mm and the thickness of one half of the specimens used in this study. For low ΔK
265 values the crack rates are two orders of magnitude higher than the rates determined
266 in this work for a DMLS Ti6Al4V stress relieved at 380 °C for 8 hours. The scatter
267 of the experimental points in both cases is high. In the case of the as-fabricated alloy
268 investigated in [18] the authors express the idea that the large scatter may be
269 attributed to the residual stresses present in specimens. Occurrence of non-uniform
270 residual stresses approaching the yield stress in as-fabricated DMLS Ti6Al4V alloy
271 was reported in [25]. In this work, however, the heat treatment applied to CT
272 specimens resulted in full stress relieving, which was proved by X-ray measurements
273 prior to testing. Only insignificant residual compressive stresses not exceeding 30
274 MPa were detected. This indicates, that the scatter is more likely related to the non-
275 continuous crack growth observed on the specimen surface. This effect is similar for
276 all three crack growth directions, without any relevance to the build direction. The
277 large data scatter in the case on as-build material is evident also from the paper by
278 Leuders et al. [17] where, however, direct current potential drop was used for the
279 crack monitoring. This method should eliminate the problems with locally different
280 crack growth rate along the crack front, which can manifest itself in the case of
281 optical surface crack increment monitoring.

282 Low temperature stress relieving and annealing heat treatments improve
283 fatigue crack growth resistance relative to the as-build condition and eliminate the
284 influence of anisotropy [26]. Fig. 10 shows comparison of the crack growth curve
285 determined in this work with results on as-build material by Leuders et al. [17] for

286 two orientations a and c . It is obvious from the comparison that (i) the stress
287 relieving used in this study improves the threshold behavior and (ii) removes the
288 effect of directionality. It is worthwhile to note that the stress relieving in our case did
289 not change noticeably the as-build microstructure.

290 The fatigue crack growth rates of the present study and the data replotted from
291 paper by Cain et al. [26] are shown in Fig. 10. Inspection of the figure reveals that all
292 specimens of both studies show very similar behavior and actually overlap when in
293 the same ΔK range. It can be concluded that all data from the crack growth threshold
294 up to the rates of the order 10^{-3} mm/cycle can be described by the eq. (2).

295 The fatigue crack propagates in a transgranular mode through the very fine
296 structure consisting of needles of α' -martensite, which was transformed from the
297 prior columnar β grains during rapid cooling in the DLMS process. The original
298 boundaries of columnar β grains do not obviously influence the crack paths, as it can
299 be concluded from Fig. 8a, which shows the growth of the crack across the build
300 layers. The fracture surface appearance is very similar in the whole measured range
301 of the crack growth rates. From fracture surface observations it follows that the
302 damage of the fine structure consisting of needles of α' -martensite at the tip of the
303 fatigue crack is the decisive process of the crack propagation. The measurement of
304 fracture surface roughness evidenced higher values for the fracture surface created by
305 the crack propagation along the layers when compared to the propagation across the
306 layers but this fact seems to be not decisive for the crack growth. The directionality
307 of the prior columnar β grains does not play a role, i.e. the resistance of the material
308 against the crack growth is the same in the case when the crack intersects the build
309 layers, (specimens a and b) or propagates along the layers, specimen c .

310 The applied parameters of the DMLS process resulted in an alloy which has
311 homogeneous microstructure in the scale of tens of microns, compare Figs. 6b and
312 7b. The directionality of structure, which appears at lower magnification and
313 irregularities related with dark areas and locally finer microstructure, Fig. 6a have no
314 significant effect on the long crack growth. In the frame of the large data scatter,

315 which seems to be inherent to the material behavior in as-build condition, there is no
316 apparent influence of the building orientation on the crack propagation rate and on
317 the crack propagation threshold ΔK_{th} , which is equal to $3.7 \text{ MPam}^{1/2}$.

318 Fig. 11 shows the comparison of the experimentally determined crack growth
319 curve from Fig. 5 (black line, eq. (2)) with literature data on conventionally produced
320 Ti6Al4V alloy. It can be seen that in the Paris region the crack growth resistance of
321 the DMLS alloy is fully comparable with the alloy produced conventionally. The
322 threshold value $3.7 \text{ MPam}^{1/2}$ is lying in the range from 3 to $5 \text{ MPam}^{1/2}$ of
323 conventional material which follows from papers [27-29]. Based on the present study
324 it can be concluded that the long fatigue crack growth resistance of the Ti6Al4V
325 alloy prepared by DMLS technology is fully comparable with the material prepared
326 conventionally.

327

328 **4. Conclusions**

329 The growth of long fatigue cracks in Ti6Al4V alloy manufactured by direct
330 metal laser sintering and subsequent stress relieving at $380 \text{ }^\circ\text{C}/8 \text{ h}$ can be described
331 for loading conditions with stress ratio $R = 0.1$ by the equation

$$332 \quad da/dN [\text{mm/cycle}] = 3 \times 10^{-8} \Delta K^{3.0} - 1.52 \times 10^{-6}.$$

333 The equation covers a broad range of growth rates from threshold $\Delta K_{th} = 3.7$
334 $\text{MPam}^{1/2}$ up to the rates of the order 10^{-3} mm/cycle .

335 The studied Ti6Al4V alloy does not exhibit dependence of the tensile
336 properties and long crack growth on the build direction.

337 The crack growth data have large scatter which is characteristic feature of the
338 as-fabricated material. It was found that the scatter is not related to the residual
339 stresses which are in some studies believed to be the cause of this phenomenon. The
340 observed scatter is more likely related to the non-continuous crack growth observed
341 on the specimen surface.

342 The crack growth rates and the crack propagation threshold are comparable
343 with those determined for conventionally manufactured alloys.

344 Fatigue crack growth takes place in a transgranular mode by cyclic damaging
345 of acicular α' -martensite within the plastic zone. The original β grain boundaries do
346 not play an important role in the crack growth process.

347

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353

354

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	Al	V	O	N	H	Fe	C	Ti
standard value	5.5 - 6.5	3.5 - 4.5	< 0.100	< 0.050	< 0.012	< 0.250	< 0.080	balance
real value	6.06	3.90	0.085	0.006	0.002	0.250	0.007	balance

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Table 1. Chemical composition of Ti6Al4V powder in wt. %.

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orientation	σ_{UTS} [MPa]	$\sigma_{0.2}$ [MPa]	A [%]
<i>a</i>	1255 ± 21	968 ± 14	5.4 ± 0.5
<i>b</i>	1235 ± 11	957 ± 4	6.5 ± 0.6
<i>c</i>	1195 ± 9	952 ± 1	7.4 ± 1.2

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458 Table 2. Tensile properties of DMLS specimens with different orientation to the
459 build direction.

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	specimen <i>a</i> threshold region	specimen <i>a</i> Paris region	specimen <i>b</i> threshold region	specimen <i>b</i> Paris region	specimen <i>c</i> threshold region	specimen <i>a</i> Paris region
Ra [μm]	7.06	5.47	6.11	4.70	10.4	12.4

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466 Table 3. Fatigue fracture surface roughness in threshold and Paris region.

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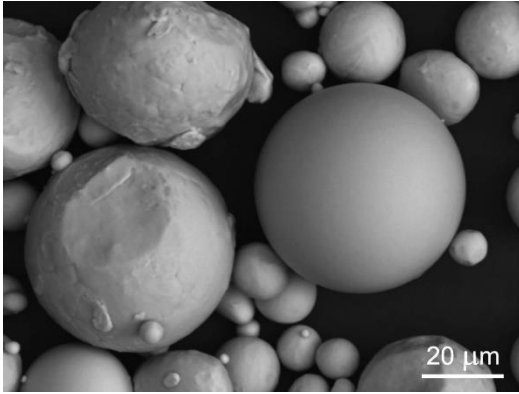
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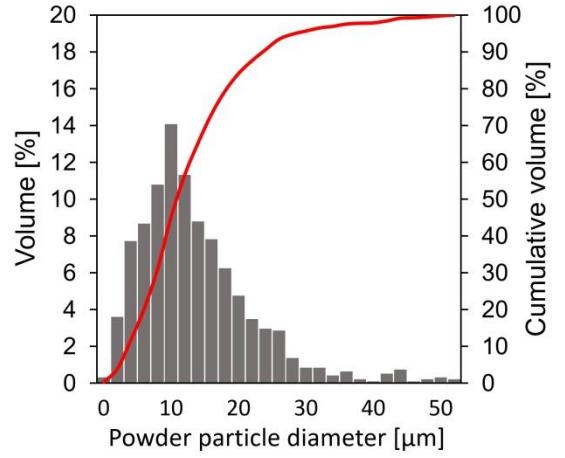
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a)



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479 Fig. 1. a) Ti6Al4V powder particles used for production of specimens. b) Powder
480 particle size distribution.

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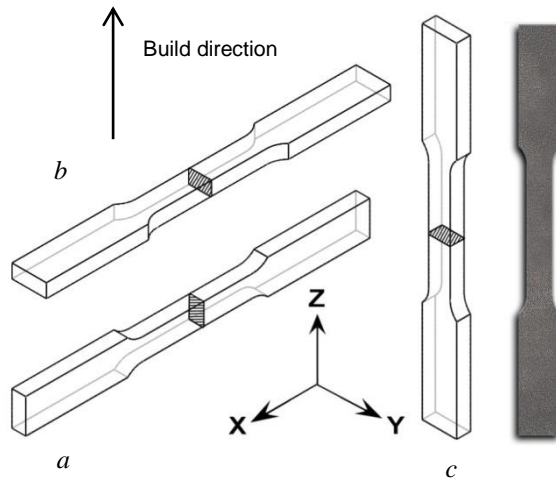
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493 Fig. 2. Three orientations of specimens for tensile test with respect to the build
494 direction. The cross-section plane for the orientations *a* and *b* is the plane z-y and for
495 *c* x-y.

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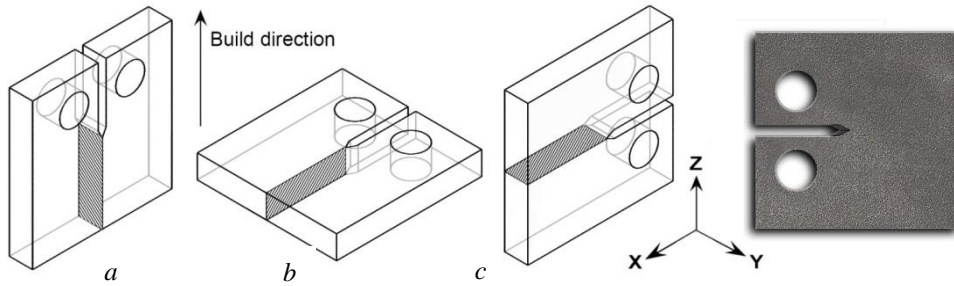
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505 Fig. 3. Three orientations of CT specimens for crack growth tests with respect to the
506 build direction Z, the macroscopic crack plane for the orientation *a* coincides with
507 the plane z-y, *b* with plane z-x and *c* with plane x-y.

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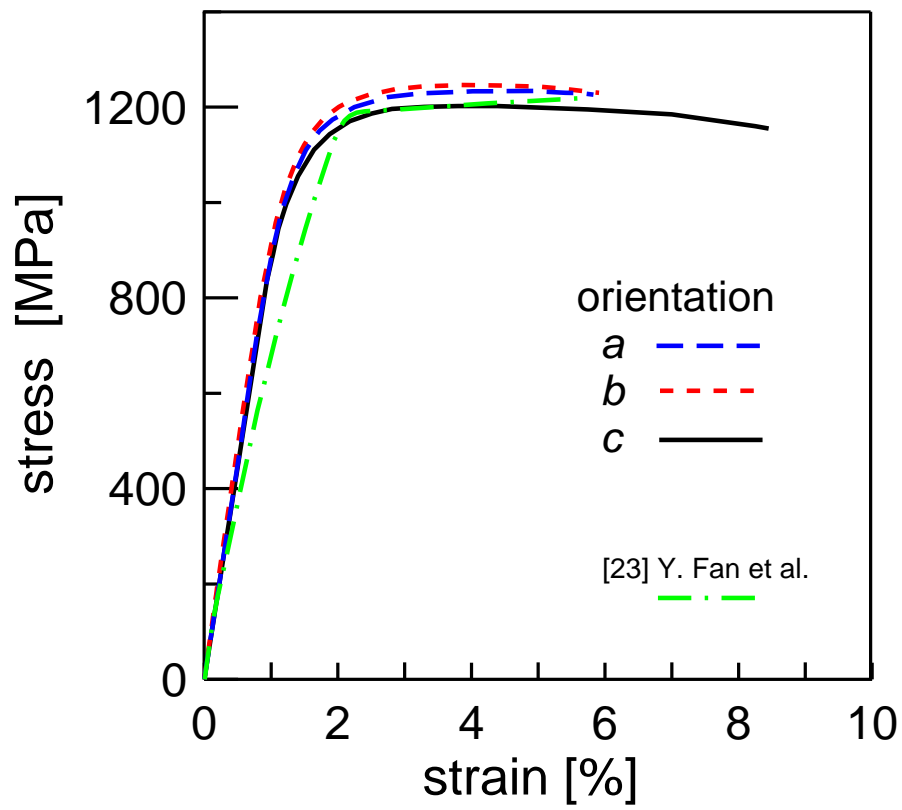


Fig. 4. Stress-strain curves of DMLS and conventional Ti6Al4V alloy.

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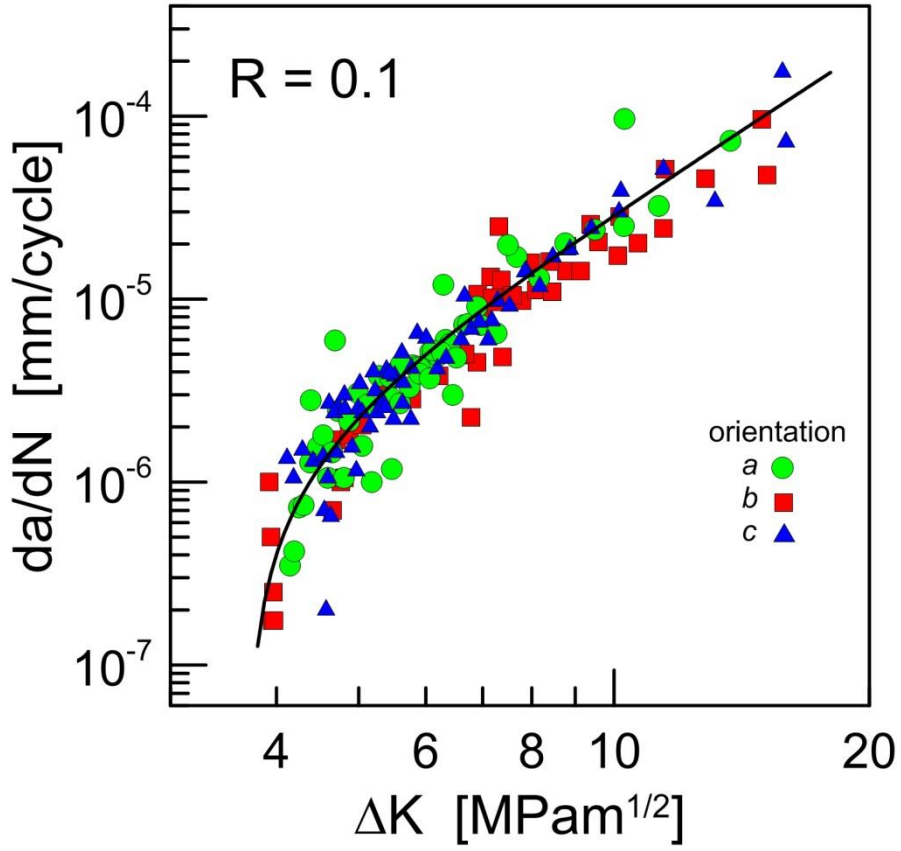


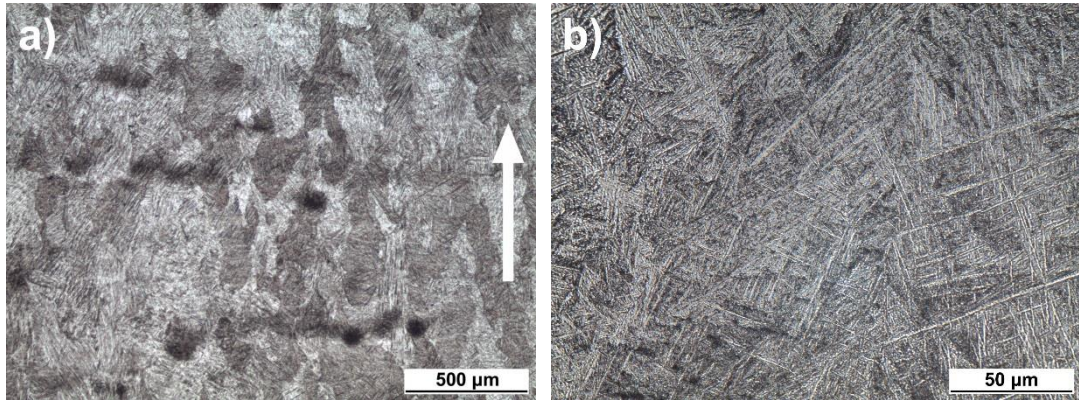
Fig. 5. Fatigue crack growth data for Ti6Al4V alloy manufactured by DMLS.

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557 Fig. 6. Microstructure a) as revealed on the lateral planes, b) details of α' -martensite
558 fine needles, etched by 10% HF. Build direction is indicated by the arrow.

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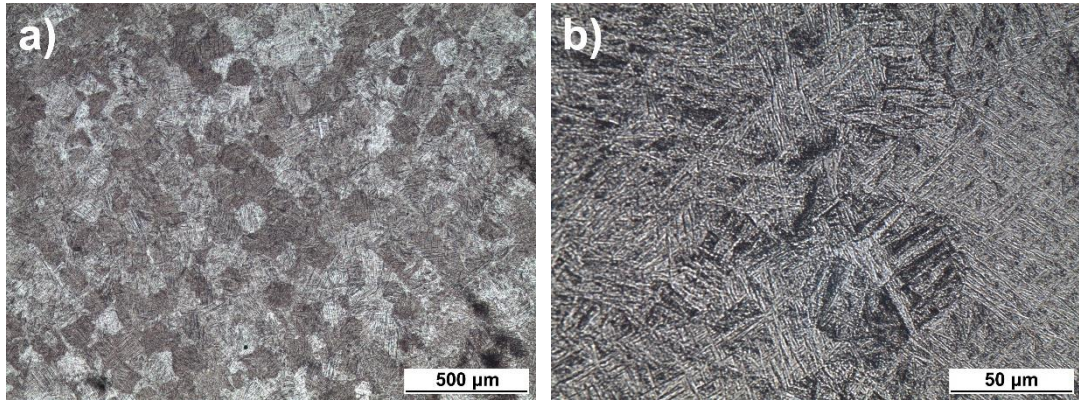
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567 Fig. 7. Microstructure a) as revealed on the horizontal plane and b) details of
568 α' -martensite fine needles, etched by 10% HF. Build direction is perpendicular to the
569 observed surface.

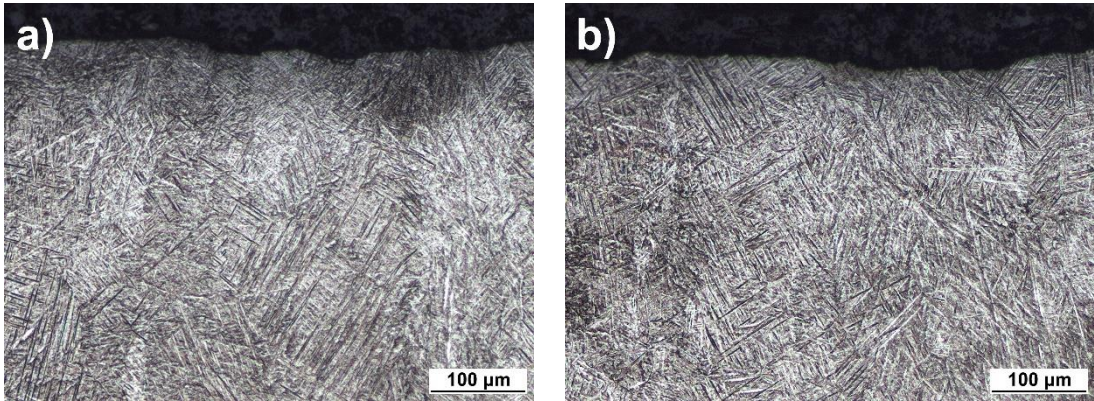
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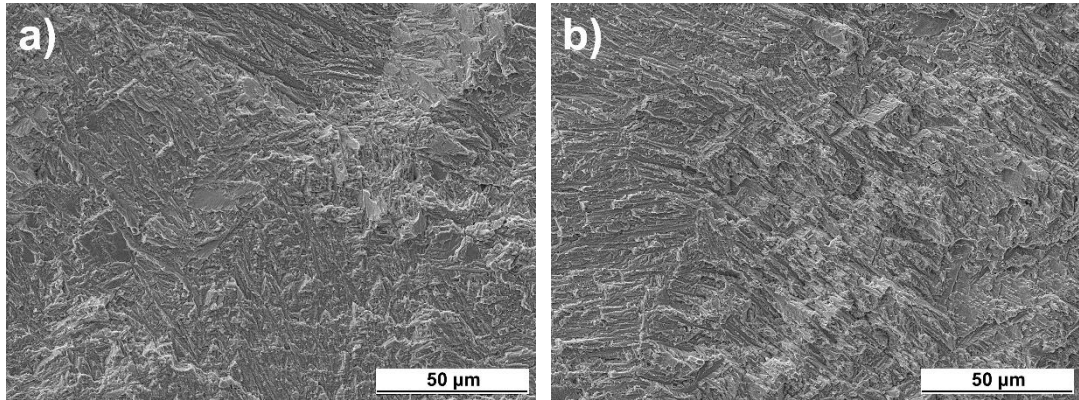
577 Fig. 8. Fatigue fracture surface profiles, a) threshold region, $da/dN \sim 10^{-7}$ mm/cycle,
578 specimen orientation *c*. b) Paris region, $da/dN \sim 10^{-5}$ mm/cycle, specimen orientation
579 *c*, etched by 10% HF. The build orientation is perpendicular to the fracture surface.
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587 Fig. 9. Examples of fatigue fracture surfaces of specimen c, a) threshold region,

588 $da/dN \sim 10^{-7}$ mm/cycle, b) Paris region, $da/dN \sim 10^{-5}$ mm/cycle.

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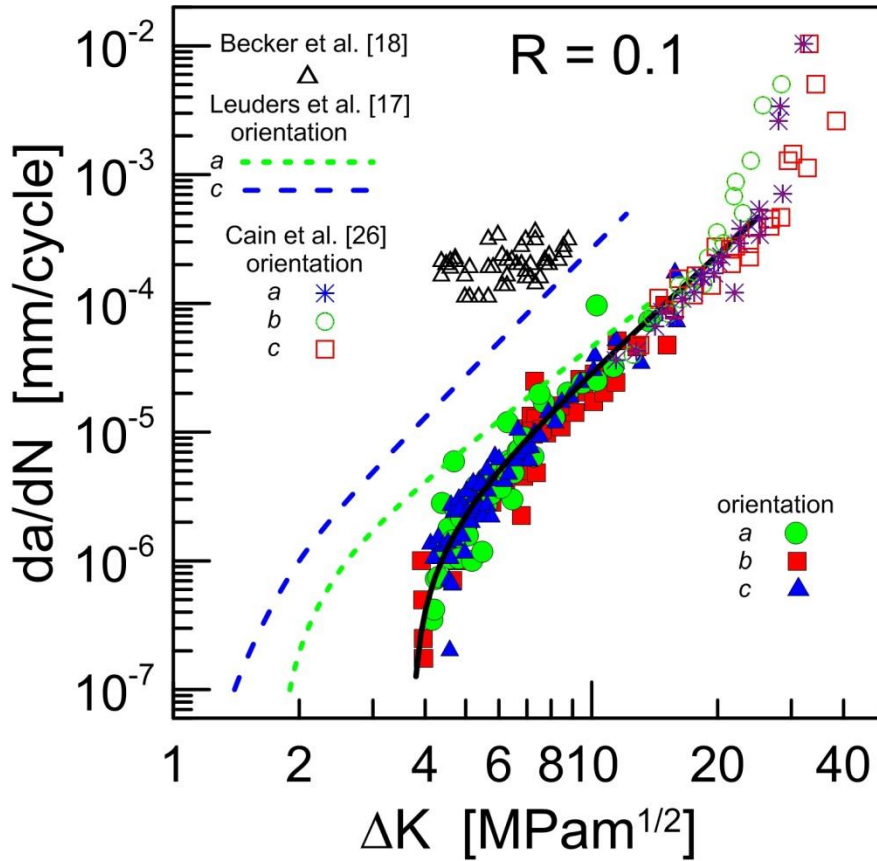
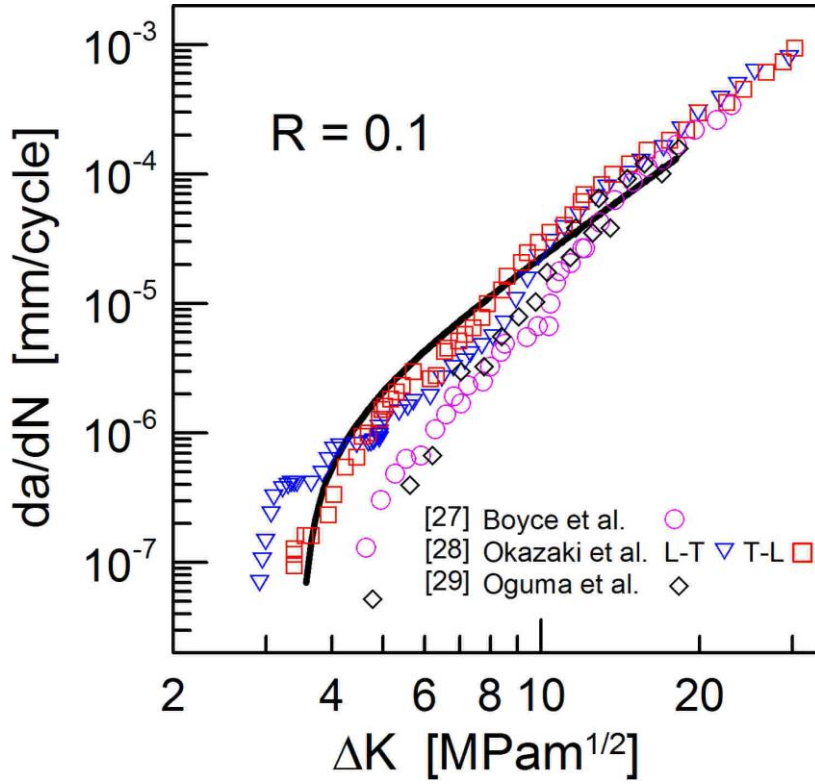


Fig. 10. Fatigue crack growth curves for as build and heat treated Ti6Al4V alloy produced by DMLS.

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635 Fig. 11. Comparison of fatigue crack growth rates in Ti6Al4V prepared by DLMS
636 (black continuous line) and conventionally (different symbols).