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Effects of laser shock processing on high cycle fatigue crack growth rate and fracture toughness of aluminium alloy 6082-T651

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37 **Abstract**
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39 The effects of laser shock processing without protective coating on high-cycle fatigue crack
40 growth and fracture toughness were investigated. Laser shock peening treatment was
41 performed on compact tension specimens from both sides perpendicular to the crack growth
42 direction, followed by subsequent grinding. Fatigue crack growth tests were performed at
43 frequencies between 116 and 146 Hz, at $R=0.1$ and a constant stress intensity range during the
44 fatigue crack initiation phase and K -decreasing test. A lower number of cycles was required to
45 initiate a fatigue precrack, and faster fatigue crack growth was found in tensile residual stress
46 field of LSP-treated specimens. The crack growth threshold decreased by 60 % after LSP
47 treatment. The fracture toughness decreased by 28-33 % after LSP treatment. The fatigue-to-
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ductile transition boundary on fractographic surfaces show linear fatigue crack fronts in non-treated specimens and curves after LSP treatment.

Keywords: laser shock processing, fatigue crack growth, aluminium alloy, residual stresses, fracture toughness.

1. Introduction

Laser shock processing (LSP) is a relatively new surface treatment technology to extend the lifetime of dynamically loaded components. Several studies [1-3] have already demonstrated that LSP treatment decreases the fatigue crack growth (FCG) rate through the generation of large amplitude compressive stresses as a result of material local plastic deformation induced by the pressure shock wave. Since the fatigue cracks normally initiate from the surface, ‘locked’ compressive stresses will, under external load, be superimposed on the applied stress, leading to smaller stress intensity factors at the crack tip and possible crack closure to incur the reduction of effective driving force for the FCG [1-2].

LSP can also improve fatigue strength and fatigue crack initiation life due to a considerable densification of dislocations and microstructural grain refinement. Huang et al. [3] confirmed a reduced FCG rate due to the highly tangled and dense dislocation arrangements in the LSP-processed surface of an Al-Mg-Si alloy (6061-T6). In another study Lu et al. [4] reported obvious microstructure refinement in Al-Cu-Mg alloy (LY2) after LSP. The results confirmed LSP to be a promising method to obtain a high density of dislocations to improve fatigue resistance, whereas the minimum grain size in the top surface after multiple LSP impacts was about 100–200 nm. Furthermore, residual stress resulting from laser processing can be significantly higher with much deeper effective penetration in the material compared to conventional shot peening [5].

Huang et al. [1,3] reported beneficial effects of LSP under different process setups on the fatigue crack growth properties of 6061 T6 aluminium alloy. Their results indicate that the FCG rate decreases with increasing laser energies [1] and LSP coverage areas [3], especially in the initial fatigue crack growth stage. However, the strengthening effects are weak in the final period of fatigue crack growth since the residual stresses release as the crack grows. In another study, Hatamleh et al. [5] compared laser and shot peening effects on FCG in friction-stir-welded Al-Zn alloy (7075-T7351) joints, reporting reductions of the FCG rate in comparison with the base material for the LSP-treated specimen, whereas shot peening revealed negligible improvement of the FCG rate.

However, despite the fact that the LSP process in which a protective coating is used has already been proven as an advanced, competitive surface treatment for improving the fatigue life, corrosion resistance, hardness and wear resistance of various metals and alloys [6-9], several negative points exist that limit the widespread application of this process. Open discussion on the last, i.e. 5th international conference on laser peening and related phenomena in Cincinnati, 2015 [10] pointed out that new/combined strategies are needed in the industrial field applications. Basically, two main approaches laser shock treatment to achieve considerable merging of dislocations and generation of compressive residual stresses are used; (i) first regime, uses high-energy laser pulses (<100 J), longer pulse duration up to 100 ns combined with the protective coating, (ii) the second approach uses lower laser energies in the order of few joules or fewer, increased overlap, smaller spots and no coating (also known as LSPwC). A detailed description of both principles was reported previously [11].

Although, the protective coating on the processed part prevents surface ablation, while maintaining high surface quality, it is a time-consuming affair since the overlay is damaged severely during the LSP process and requires frequent replacements, making it slow and expensive in industrial applications [12]. Furthermore, since some parts being peened might

not be accessible for the application of protective overlay, much attention was focused on LSP process without a need for a protective ablative overlay. However, after this process typical surface craters are obtained with increased surface roughness and waviness due to the local surface ablation, which in turn creates a high-pressure plasma that, contained by a thin layer of water flowing on the specimen surface, generates a pressure wave propagating into the specimen [13]. Based on the open literature review, very few comprehensive studies have been conducted on laser shock processing without protective coating on fatigue crack growth, and none of them have evaluated fatigue characteristics after subsequent grinding procedure to eliminate the downward effect on uncoated LSP-induced surface topography.

Rubio-Gonzalez et al. [14-15] investigated the effects of a two-sided uncoated LSP process on compact tension (CT) specimens of 2205 duplex stainless steel and 6061-T6 Al alloy. The results confirmed that the FCG rate decreased while fracture toughness increased with the increased pulse density. Although there are numerous investigations of laser processing effect on material fatigue crack growth properties [1,5,14], there is limited literature on the possible detrimental effect of laser shock processing. Further, all the above studies have been performed after pre-existing fatigue pre-crack, under low sine wave frequencies (<30 Hz) and high maximum load in a low cycle fatigue (LCF) regime.

With this in mind and since early recognition of premature failures plays a crucial role in engineering components, a detailed investigation of the uncoated LSP effects on AA 6082 on fatigue crack growth rates will be assessed. Special attention was focused on the determination of fatigue crack growth behaviour under dynamic loading at high resonant frequencies, starting at approx. 150 Hz, measuring crack growth rate using a small load and high cycle fatigue (HCF) regime ($N > 10^6$) [16]. To obtain a quantitative comparison between the fracture toughness of untreated and LSP specimens in the final rupture stage under static load will be determined as well. In addition, the possible softening effect due to the local

surface melting, ionization, and re-solidification during laser process will be carefully investigated by means of dislocation arrangements, residual stress, and microhardness in depth distribution. All LSP specimens under investigation were analyzed after subsequent grinding procedure in order to obtain a proper surface finish.

2. Experimental methods

2.1. Material and specimen preparation

Plates of commercial wrought Al-Mg-Si aluminium alloy (EN AW 6082-T651) were machined to obtain the specimens. The overall heat treatment T-651 procedure (homogenization, solution treatment, aging, etc.) is detailed in ref. [17]. Its chemical composition in wt% was 0.87 Si, 0.72 Mg, 0.42 Mn, 0.35 Fe, 0.15 minor elements (Cu, Cr, Ni, Zn, Ti) and balance Al. In order to obtain mechanical properties for the given material in the specific rolling direction, three cylindrical tensile specimens were machined in longitudinal and three in the transversal direction of rolling with a test diameter of 5 mm and a measuring length of 20 mm (Fig. 1). The average measured tensile strength, yield stress, and elongation are given in Table 1. A total of six C(T) specimens were tested from the plate in the L-T direction for fatigue crack growth and fracture toughness tests. Three specimens were tested in not-treated conditions, and three specimens were LSP-treated with different parameter sets: LSP 1, LSP 2 and LSP 3. Notches were made using electrical discharge machining (EDM) using a 0.35 mm wire, which resulted in a notch with a radius of 0.19 mm. The geometry of the C(T) specimen, adopted for testing with Rumul Cracktronic unit and Krak-gages AMF-5, is given in Fig. 2.

The applied stress intensity range ΔK was used, calculated according to the equation [18]:

$$\Delta K = \frac{\Delta P}{B\sqrt{W}} \cdot Y, \quad (1)$$

where $\Delta P = P_{\max} - P_{\min}$ is the alternating force, B is the specimen thickness, W is the length from the load centerline to the edge of specimen, Y is a geometrical factor, which is derived for linear elastic material for compact-tension specimens:

$$Y = \frac{(2 + \alpha)}{(1 - \alpha)^{3/2}} (0.886 + 4.64\alpha - 13.32\alpha^2 + 14.72\alpha^3 - 5.6\alpha^4), \quad (2)$$

where $\alpha = a/W$, a is the crack length.

2.2. Laser shock processing

Laser shock processing was performed using a Spectra-Physics Q-switched Nd:YAG laser with an irradiation wavelength of 1064 nm. The maximum laser beam energy was 2.8 J/pulse, whereas the FWHM (Full Width Half Maximum) of the generated Gaussian intensity pulses was 10 ns. Focused laser beam diameters were varied via modification of convergent lens. Beam diameters, used during LSP 1, LSP 2 and LSP 3 treatments, were set to 1.5, 2.0 and 2.5 mm, respectively. In this study, a predefined overlapping pulse density of 1600 pulses/cm² was chosen, with the laser-advancing direction parallel to the plate-rolling direction (L).

Specimens were irradiated using laser shock processing method without any protective coating in a water confinement regime, whereas laser pulses were overlapped and scanned in a zig-zag pattern (Fig. 3). A water jet set up was employed to create a thin water layer and to avoid the formation of water bubbles or the concentration of impurities resulting from the material ablation, thus, constantly ensuring a pure laser–matter interaction.

The treated area on the C(T) specimens was approx. 10×10 mm² on both sides of the specimens. For TEM microstructural analysis, additional LSP specimens were prepared using the same processing conditions as for C(T) specimens. Prior to fatigue crack growth testing, C(T) specimens were subsequently ground to the final thickness of 9.8 mm to remove the rough asperities after LSP treatment (Fig. 4). Prior to laser shock processing, no pre-crack was

initiated. However, specimens for TEM analysis were not ground in any way to obtain the representative microstructures produced by the laser peening treatment.

2.3. Surface roughness

After the surface near pre-machined notch was laser surface treated, the surface profile and the roughness were measured using Surtronic 3+ profilometer (Taylor Hobson) and the x-y micrometer sliding table stroke. TalyProfile Lite v.3 software was used with micro-roughness filtering ratio of 2.5 μm and a Gaussian filter to extract the roughness parameter. The selected parameter for evaluation of surface roughness is R_a , which is the average arithmetic deviation from the mean line on the measuring length. The measuring profile length for the roughness parameter R_a determination of LSP-treated surfaces was 16 mm, while the measuring length used for measuring roughness on crack surfaces was 2.5 mm.

2.4. Transmission electron microscopy

The LSP effect on the material microstructure was characterized using a JEOL 2000-FX transmission electron microscope (TEM) operated at a voltage of 200 kV. In order to obtain proper insight into LSP effects on microstructure and dislocation configurations, we chose not to grind the LSP surfaces in any way, due to a possible loss of information. Instead, we prepared cross-sectional thin foils for TEM analysis in the following steps: (i) bonding two pieces of the LSP-treated specimens face-to-face with GatanTM glue (treated surfaces in the middle), (ii) cutting a cross-sectional specimen into 1.0 mm thick sheets, (iii) grinding it carefully to about 100 μm thick; (iv) punching out to 3 mm diameter discs, (v) ion thinning by Ar⁺ bombardment. On average, 80 microstructural images for each specimen condition were taken from the depth 0 to about 500 μm from the original surface. For the sake of

convenience, images with the highest dislocation densities were chosen as the representative ones for each LSP-processing condition.

2.5. Microhardness

The microhardness was measured with a micro-Vickers test using a small load of 0.071 N, Leitz Wetzlar gage, Leica Microsystems. Indentations were made in the vertical direction through the entire cross section of C(T) specimen. In order to obtain proper information about LSP effects on hardness, the surface was not ground, since the hardened material could have been completely removed. The first and last 10 measurements below the surface were made using a spacing of 32 μm to increase the number of measurements in the subsurface region of interest. Then, the spacing over the rest of the material thickness increased to approx. 214 μm in order to reduce the experimental time.

2.6 Residual stresses

The analysis of residual stresses was performed using the hole-drilling relaxation method in accordance with the Standard Test Method for Determining Residual Stresses by the Hole-Drilling Strain-Gage Method ASTM E 837-08 [19]. Residual stresses were determined with measurement equipment, a product of the Vishay Measurements Group (Vishay Intertechnology Inc., Malvern, PA, USA) with the resistance gage CEA-06-062UM-120. The hole was drilled using a depth feeder with micrometer (resolution 0.01 mm). The hole diameter was measured after drilling and averaged on three locations using an optical Olympus microscope, a ColorView Camera, and AnalysisDocu software. Maximum and minimum principal residual stresses and the orientation angle were calculated using H-drill v3.10 software and the integral method of stress calculation with a strain deviation error estimation of 3.2 μm .

2.7 Fatigue crack growth test

Fatigue testing was performed on the electromagnetically driven testing device, Cracktronic 8204 with Fractomat, by Russenberger Prüfmachinen AG. Testing was performed according to the available testing space on Cracktronic and Standard Test Method for Measurement of Fatigue Crack Growth Rates ASTM E647 [20]. The basic module of Cracktronic has separate static and dynamic drives. The static load is generated by a ball screw spindle driven by a DC-motor and couples over the torsion rod into the oscillating system. The dynamic load is generated by means of electromagnetic driven resonator. The electromagnet is integrated in a closed loop system and excites the oscillating system in its natural resonant frequency, where the operating point is situated at the peak of the resonance curve with typical frequencies between 50 and 200 Hz. The dynamic and static parts are controlled by separate circuits, connected with the Credo control unit and computer software, and allow any combination of stress ratios with high accuracy. The alternating force is measured using Rumul load cells (piezoelectric) with the maximum load of 8000 N and a resolution of 1 N. Furthermore, the specimen with its elasticity is part of the system [21]. The crack lengths were continuously monitored using the indirect electric potential drop method. In this method, the electric potential value is taken off the thin (5 μ m) constantan resistance foil Krak-gages (type RMF-A5), as opposed to the “direct potential drop method”, in which the potential is taken off directly from the specimen. The Krak-gages were attached on both sides of the specimen to monitor and average the crack lengths. The crack in the gage is growing simultaneously with the real crack of the specimen. The measurement of crack length is recorded in a resolution of 0.001 mm and the accuracy is higher than 2 %. The main source of error is gage misalignment when gluing the gage on specimen, which amounts to ± 0.2 mm from the actual notch tip radius. The gluing of Krak-gages was performed using

cyanoacrylate-based glue (Vishay Measurements) under optical macro-scope at 15-20 × magnification with the same technique as strain gage gluing. The error of positioning was measured by spatially calibrated microscope. The error length of positioning from actual EDM notch was taken into account at starting notch length value input in Fractomat. In order to perform $\Delta K = \text{const.}$ tests at a constant R -ratio, the relationship in the equation (1) is integrated in testing machine software algorithm to control the dynamic force of testing. As the crack length is increasing, new ΔP , $P_{\max}$, $P_{\min}$ and P_{mean} are calculated at $R = \text{const.}$ where $P_{\max} = \Delta P / (1 - R)$, $P_{\min} = R \Delta P / (1 - R)$ and $P_{\text{mean}} = 0.5(P_{\max} - P_{\min})$. The crack growth rates were calculated using two methods. First, the point-to-point secant method:

$$da/dN = \frac{\Delta a}{\Delta N} = \frac{a_{i+1} - a_i}{N_{i+1} - N_i} \quad (3)$$

This method involves a direct point-to-point calculation of $\Delta a / \Delta N$, in which small deviations from nominal crack length lead to large scatter. The second method involves fitting an appropriate best-fitting equation using the least squares method, which is described in Chapter 3.5.

The fracture toughness K_{c} was measured on compact-tension specimens after fatigue crack growth tests using universal testing machine BETA 50 Messphysik with a piezoelectric force sensor with measuring capacity of 50.000 N and with a resolution of 1 N. The displacement was measured using an optical laser-speckle extensometer with accuracy within 2 %. Initial fatigue crack length was measured under the optical microscope from the line of load to the crack front, where the measurement error is ± 0.1 mm. Fracture toughness was conducted under thickness-dependent plane stress or transition plane stress/plane strain conditions.

Fractographic analysis was conducted using an Olympus stereo microscope SZX10 and JEOL 5610 scanning electron microscope (SEM). The images were taken on preselected

locations on the fatigued surface and at transition zone from fatigue fracture surface to ductile fracture.

3. Results and discussion

3.1. Roughness analysis of LSP-treated surfaces

As a consequence of the absence of an absorbent coating during laser shock processing, the mean arithmetic roughness parameter R_a of LSP-treated specimens increased significantly in comparison with untreated specimens (Table 2). The average roughness in the as-received state was 1.15 μm , while the average R_a roughness after LSP using 1.0, 1.5, 2.0 mm beam diameters, increased to values of 7.3, 10.4 and 7.92 μm , respectively. All LSP specimens were subsequently ground to R_a values of 0.37-0.38 μm before the FCG tests.

Figure 5 shows measured profiles, starting on a non-treated surface continuing to a rougher laser shock-treated surface.

3.2. Transmission electron microscopy

Cross-sectional microstructures of the LSP-treated specimens are shown in Fig. 6. Evidence of intense, strain-hardening deformation is obvious in the surface layer, with various dislocation arrangements. Dislocations are inhomogeneous in terms of depth from place to place, due to different regimes used with specific LSP treatment and due to the heterogeneous nature of the plastic deformation within and between specific grains.

Comparing Fig. 6b and 6d clearly indicates that the highest and the lowest dislocation density was achieved with the specimens LSP 1 and LSP 3, respectively. Such results are rather logical since larger a laser beam diameter reflects a smaller laser intensity and shock wave pressure, since the intensity is inversely proportional to the laser beam area. Thus, the microstructure of the LSP 1 specimen (Fig. 6a and b) revealed randomly arranged, high-

1 density dislocation structures with dislocation tangles and dislocation walls. Moreover, a grain
2 refinement with the formation of ultra-fine grains was confirmed, which can effectively
3 reduce fatigue crack growth in the polycrystalline metals [3]. It seems that a grain refinement
4 mechanism is primarily achieved via formation and evolution of dislocation structures
5 contributing to the formation of dislocation cells (Fig. 6c). With further hardening,
6 dislocation, cells convert to ultra-fine equiaxed grains in order to minimize the total energy
7 state due to repeated shockwave impacts on the target surface.
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19 *3.3. Analysis of hardness below the LSP-treated sub-surface layer*

22 The LSP-treatment could lead to an increase in hardness in the sub-surface layer compared
23 with the base material, due to strain hardening. For example, the study of Gonzales et al. [14]
24 showed that Vickers microhardness after LSP treatment of Al-Mg-Si alloy is increased in the
25 sub-surface layer from 83 HV to values 96 and 94 HV for pulse densities 900 and
26 1200 pulses/cm², respectively. Figure 7 shows location and the line of micro-hardness
27 indentations. Vickers hardness measurements with loads of 0.071 N on same alloy series
28 revealed no significant hardness changes in the subsurface layer after LSP. The average
29 Vickers indentation size amounts to 11.16 μm ($\pm 0.27 \mu\text{m}$). For a constant function of
30 hardness, the analysis of residuals reveals a normal distribution of hardness and a mean value
31 of 108 HV (± 7 HV); essentially, all tested specimen had similar values. Thus, the effect of
32 LSP on subsurface and through-depth hardness was negligible.
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48 Microhardness results obtained in our study are consistent with those reported by Peyre et
49 al. [22] and Fabbro et al. [23] concerning the application of laser shock peening on aluminium
50 alloys. Their results also revealed little improvement in the hardness properties of LSP-treated
51 specimens, with lower HV values in comparison to the specimens treated by the conventional
52 shot-peening process. According to Fabbro et al. [23] this behaviour may be explained in
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three ways: (i) shock duration is very small, so that hardness cannot initiate and propagate inside the material, (ii) compared to the shot-peening process, no contact deformation occurs, i.e. no Hertzian loading, and (iii) impact pressures are usually lower than those from shot-peening process.

3.4. Residual stresses

The specimen in an as-received state and specimen LSP 3 after grinding was selected for analysis of residual stress with the hole-drilling relaxation method. Fig. 8a shows the initial residual stress state of the base material after cold rolling, stress-relieving and aging at 160 °C for 10 h. The magnitude of residual stress is between 0-50 MPa in tension at the surface up to 0.5 mm of drilling depth. The short transition to compressive stress between 0 and -25 MPa occurs at 0.65 mm of depth, followed by a gradual return to tension at 0.9 mm. Fig. 8b shows the residual stresses as a function of drilling depth after LSP 3 treatment. At lower depths below the surface, relieved stresses are slightly in tension for the maximum stress component and slightly compressive for the minimum principal stress component.

Small surface compressive RS with the LSP 3 specimen most likely originate from the softening effect during laser treatment due to the local surface melting, ionisation, and re-solidification. Moreover, good agreement with TEM microstructures can be seen, where specimen LSP 3 revealed the lowest dislocation density among all LSP-treated specimens. Nevertheless, as can be seen from Fig. 8b the residual stresses became purely compressive at the depth of ~0.05 mm and reached a maximum compressive RS of -230 MPa at 0.25 mm. The residual stresses remain fully in compression up to 0.85 mm of depth, followed by the transition to balancing tensile residual stresses. It should be noted that there were no residual stress measurements made in the region of the plate from 1 to 9 mm deep. However, in order to balance the compressive stresses at the surface, tensile residual stresses are most likely

present in the centre of the specimen. Similar results were recently reported in the investigation of LSP effects without protective coating on Inconel 718 alloy by Gill et al. [24]. Their results also confirmed small surface compressive stresses (–50 MPa), which became tensile in nature, and persisted for approximately 80 µm from the surface that afterwards changed to the compressive state. Through TEM, the same authors proposed that during laser surface ablation, shock waves re-deposit ablated particles onto the matrix, which prevented grain refinement. Their proposal is partially consistent with our results, but only for the LSP 3 specimen, since both LSP 1 and LSP 2 exhibit high dislocation densities, dislocation cells, and refined grains.

3.5. Analysis of fatigue crack growth at low-stress intensity levels

To investigate the fatigue crack growth rate before and after LSP treatment, the standard procedure of the fatigue precracking was performed at $R=0.1$ and at $\Delta K=\text{const.}=4 \text{ MPa}\sqrt{\text{m}}$ to reach a fatigue crack length of 1.5 mm. The length of pre-crack is in agreement with ASTM E640-05, where the precrack length should be at a minimum length of $0.1 \cdot B$ or at least 1 mm. After precracking, the stress intensity range ΔK increased from $4 \text{ MPa}\sqrt{\text{m}}$ to starting value for K -decreasing test. The K -decreasing procedure is well suited for fatigue crack growth rates below 10^{-8} m/cycle [20]. In this test method, K -decreasing is performed by shedding force, either continuously or by a series of incremental steps as the crack grows. The stress intensity range ΔK was set according to the exponential function:

$$\Delta K = \Delta K_0 \cdot e^{[C_p(a-a_0)]} , \quad (4)$$

where ΔK_0 is the starting value of stress intensity and C_p is the gradient value. On a basis of preliminary tests and the research of Shelton et al. [25] who investigated the influence of shed-rate on threshold values, the gradient value C_p was set to -0.8 mm^{-1} . The starting value ΔK_0 was also selected on the basis of findings from preliminary tests of not-treated and LSP-

1 treated specimens. The aim was to obtain the data for the largest possible ΔK interval (to
2 reach ΔK_{th}) for the shedding parameter selected and the available fatigue crack length of 5
3 mm (KraK-gage AMF -5 limitation). For not-treated specimen #1 and not-treated #2
4 specimen, ΔK_{th} was reached from 12 MPa $\sqrt{m}$. In contrast, due to much faster crack growth
5 rate in LSP-treated specimens in our preliminary tests, ΔK_{th} was not established before crack
6 length reached the limiting value of 5 mm. Therefore, the LSP-treated specimens presented in
7 this research were tested from lower starting value of 7 MPa $\sqrt{m}$. However, to establish
8 equality in test and adequate comparison, additional not-treated specimen #3 was tested from
9 7 MPa $\sqrt{m}$.

10 During both stages of fatigue testing, the pre-cracking procedure and K -decreasing test, the
11 testing resonant frequency, and the crack length was monitored with the indirect electric
12 potential drop method on both sides of the specimen. Fig. 9a presents the fatigue crack length
13 a' as a function of a number of cycles during pre-cracking and K -decreasing test, where each
14 curve represent a single specimen. In this plot, each specimen was subjected to the same
15 testing parameter regime. Moreover, the crack direction did not deviate from the plane of
16 symmetry for more than 2°. From Fig. 9a, it is evident that fatigue crack initiated earlier in
17 LSP-treated specimens than in not-treated material. The number of cycles, required to grow a
18 precrack (of the total length of 1.5 mm) is largest for the non-treated specimen and lowest for
19 the LSP 3 specimen. In the linear part of a - N curve that is within interval from 1.2 to 1.5 mm,
20 the crack growth rate was calculated with secant method, Table 3. From this data, it was
21 calculated that the number of cycles required to grow 1.5 of pre-crack for LSP 1, LSP 2 and
22 LSP 3 specimens decreased by 28, 33 and 47%, respectively.

23 The resonant testing frequency reduces with the crack length or increases in case of strain
24 hardening at the tip of the notch or crack. Testing resonant frequency is a function of the
25 elastic spring constant of the specimen. Figure 9b shows testing frequency as a function of

fatigue crack length, $f(a')$. Overall, the highest frequency was recorded during the test of non-treated material. The testing frequency was increasing slightly before the fatigue crack initiated due strain hardening effects in the tip of the notch. After the crack growth stabilized, the frequency was decreasing with the crack length. After reaching the pre-crack length of 1.5 mm, the testing machine was increasing the stress intensity ratio range until $7 \text{ MPa}\sqrt{\text{m}}$ was reached. The frequency curve for not-treated specimen was on the highest level and most convex among tested specimens. The starting frequency standard deviation, as obtained from $n=5$ tests of base material, amounts to $\pm 2.1 \text{ Hz}$. Assuming that similar standard deviation is true for LSP-treated specimens, differences of testing frequencies among LSP treatments are not significant.

A short initial acceleration of crack growth occurs immediately after the increase of stress intensity factor range. The initial fatigue pre-crack lengths increased from 1.5 to 1.64 mm for not-treated specimen, to 1.74 mm for the LSP 1 specimen, to 1.76 mm for the LSP 2 specimen and to 1.81 mm for the LSP 3 specimen.

First, the point-to-point secant method was used to generate $\Delta a/\Delta N(\Delta K)$ graphs (Fig. 10.) from $a(N)$ data. With this method, small error deviations from continuous real $a(N)$ lead to larger “scatter” in fatigue crack growth curves. Variability of measurements can be observed for three tested not-treated specimens, (#1, #2 and #3), where #1 and #2 were tested from $\Delta K_0 = 12 \text{ MPa}\sqrt{\text{m}}$ and #3 from $\Delta K_0 = 7 \text{ MPa}\sqrt{\text{m}}$. The source of variability is the inhomogeneity of material as well as the fatigue crack growth rate measuring error. From the fatigue crack growth rate “raw” data, it can be concluded that for all specimens after being treated with laser shock processing, LSP 1, LSP 2 and LSP 3 show higher crack growth rate, which supports the earlier conclusions. In contrast, only small differences are between curves of individual LSP-treatments.

According to Broek [37], data sets (as shown in Fig. 10) must be interpreted before they can be used in analysis. In order to obtain plots with less scatter without any meaningful change of the actual crack growth length, new da/dN data was calculated from best-fitting equations. The $a(N)$ data was fitted using with least squares method with different functions. For example, best fitting function for not-treated specimen is the Hocket-Sherby 2D function:

$$a(N)_{LSP1} = b - (b - a) \cdot e^{(-c \cdot (x^d))} \quad (5)$$

where $a=1.43026$, $b=3.10645$, $c=0.101526$, $d=0.48909$.

The experimental $a(N)$ data for LSP 1 specimen was fitted using Hoerl function:

$$a(N)_{LSP1} = ab^x N^c, \quad (6)$$

where $a=1.11347187$, $b=1$ and $c=0.125$. The $a(N)$ data for LSP 2 specimen was fitted using Dr. Hill's function:

$$a(N)_{LSP2} = \alpha + \frac{\theta N^\eta}{\kappa^\eta + N^\eta}, \quad (7)$$

where $\alpha=4.286326$, $\theta=-2.82718$, $\eta=-0.481$, $\kappa=1.8E5$. The $a(N)$ data for LSP 3 specimen was fitted using the same function with $\alpha=4.579980126$, $\theta=-3.220466252$, $\eta=-0.459$, $\kappa=1.24E5$.

The data for stable II. region growth was fitted using the Paris equation:

$$\frac{da}{dN} = C \cdot \Delta K^m \quad (8)$$

where C and m are material constants. Table 4 gives C and m parameters with the fitting correlation coefficient R^2 , which are higher than 0.98. The NASGRO model [26] was used to predict a crack growth in other regions. The equation for positive R ratios was used:

$$\frac{da}{dN} = \frac{C \cdot \Delta K^n (1 - (\Delta K_{th} / \Delta K)^p)}{(1 - K_{max} / K_c)^q} \quad (9)$$

where

$$K_{max} = \Delta K / (1 - R), \quad (10)$$

where C , n , p , and q are empirical constants.

Fig. 11. represents the $da/dN(\Delta K)$ purified fatigue crack growth plots with practically no scatter (due to the da/dN calculation from continuous function) and with fitted models of Paris and NASGRO. The model coefficients are presented in Table 4. For LSP-treated specimens, the FCG curve was not sigmoidal where the slope gradually increases at values from 4.6-4.9 MPa $\sqrt{m}$. For the Paris region, the curve was fitted up to change of slope of LPS-treated specimens. The experimental data shows change of slope (knee), where one function from ΔK_{th} to knee and second from knee to K_c was used. The FCG curves for LSP 1 and LSP 2 almost overlap, indicating the equal crack growth rate. A slightly higher crack growth rate was found in the LSP 3 specimen. Therefore, without performing many repetitions, no firm conclusions can be made about the impact of LSP-parameters on fatigue crack growth.

The fatigue crack in the base material was no longer propagating ($da/dN \rightarrow 0$) at a threshold level of 3.36 MPa $\sqrt{m}$, while the crack in the LSP 3 specimen at the same SIF was propagating at approx. $3 \cdot 10^{-9}$ m/cycle. Hence, the highest fatigue crack growth rate was found in the LSP 3 specimen with the lowest threshold value of 1.1 MPa $\sqrt{m}$. Although, LSP process induced high density of dislocations, grain refinement and even compressive residual stresses in the near surface, it is obvious that this alone cannot effectively reduce the FCG rate. It is clear that all LSP specimens exhibited higher crack propagation rates in comparison to the untreated specimen due to the equilibrium tensile stresses in the centre of specimen, which reduced the effect of crack closure. It is obvious that the compressive residual stresses have not effectively decreased the crack growth rate. Chahardehi et al. [2], who studied the effect of residual stresses arising from LSP, stated that the growth rates of the crack observed are more affected by the tensile core than by the compressive surface stresses. In this study, crack length against number of cycles measurements did not show any retardation of fatigue cracks, which is believed to be due to the tensile core in the material that arises as a “by-product” of laser shock peening.

It should also be noted that Fig. 10 and 11 are plotted as a function of applied ΔK data. At low-stress intensities, crack closure occurs, and the influence is increasing near the threshold level. According to Borego et al [28] and Gavras et al [36], plasticity induced and roughness induced closure mechanisms are dominant studied aluminium alloys (among them the 6082-T6 was also the subject of the research). Considering this, corrected da/dN vs. ΔK_{eff} curves would be shifted to the left from da/dN vs. ΔK data.

Another factor to be noted is the sequence of LSP treatment. In contrast to the majority of the investigations [3, 14] in which LSP was performed after pre-existing fatigue precrack, our specimens were precracked after LSP. Secondly, LSP were performed over the edge of the EDM notch, which possibly created tensile hot spots over the notch edge. Similar observations were reported previously by Ivetic et al. [29], investigating the effects of an “uncoated” LSP process on open-holed thin Al 7075-T73 specimens. Their results indicate that the sequence of operations (first LSP and then open-hole or vice versa.) has an enormous influence on fatigue crack propagation behaviour. Moreover, it was suggested that laser beam, being focused by “spilled” plasma by the inner side of the hole, creates a local tensile hot-spot, which drastically reduces fatigue life. The same authors argued that this effect is influencing only the crack initiation phase, since the progressive relaxation of residual stresses occurred once the crack has formed.

3.6. Fracture toughness

The fracture toughness K_c test was performed on C(T) specimens after FCG tests to evaluate the effect of LSP treatments. In order to satisfy plane-strain conditions in fracture toughness testing, the thickness B of the specimen should be greater than $2.5(K_c/\sigma_Y)^2$. The yield stress and fracture toughness of investigated 6082-T651 plate is 295 MPa. Moreover, based on the literature review, the expected value of fracture toughness of 6082-T6, which

depends on age-hardening, is between 35-40 MPa√m [30]. Thus, to perform a plane-strain linear elastic fracture toughness testing according to the $B > 2.5(K_{IC}/\sigma_Y)^2$, the specimen thickness should be 46 mm. Because an aluminium plate with very large thickness is seldom practically useful, transition plane-strain/plane-stress thickness-dependent tests are quite common. In the case of thinner aluminium plates, the Standard Practice for Fracture Toughness of Aluminium Alloys B 646-12 [31] provides guidance for the testing of aluminium alloys with (i) thin products, with thicknesses of less than 6.3 mm, (ii) intermediate plate thicknesses, too thin for valid plane-strain fracture toughness tests but too thick for treatment as a sheet, thickness range from 6.3 to 50 mm, and (iii) relatively thick specimens. The investigated 6082-T651 plate with $B=10$ mm classify into (ii) intermediate thicknesses, where the fracture toughness shall also be determined according with the test method E399 [18] (as supplemented by Standard Practice for Linear-Elastic Plane-Strain Fracture Toughness Testing of Aluminium Alloys B645 [32]).

Three specimens without treatment were used to measure the average fracture toughness, while a single specimen was used for each LSP treatment. Table 5 shows the starting crack lengths, a/W ratio, maximum force P_{max} , reference force P_Q , and the calculated values of fracture toughness K_{IC} . The crack lengths (i.e. the total length of notch and fatigue crack length) were between 0.67 and 0.74 W (recommendation by ASTM E399 is between 0.45 and 0.55 W). Fig. 12 shows $P-v$ plots for specimens with starter crack length of 0.71-0.74 W . The average fracture toughness of non-treated specimen, based on three separate tests, was found to be 37.8 MPa√m. After LSP treatment, the fracture toughness was 28–33% lower compared to the untreated material. Considering that similar value of standard deviation from base material tests (± 0.86 MPa√m) also apply for LSP treatments, it can be concluded there is not a significant effect on fracture toughness values among LSP-treated specimens (LSP 1=29.01, LSP 2 = 29.34 and LSP 3=28.21 MPa√m).

3.7 Fractographic analysis

The macro-fracture surfaces of the untreated and LSP 3 specimens were compared under different magnifications using optical and scanning electron microscopes.

Fig 13a shows a relatively linear shape crack front in representative not-treated specimen. In contrast, Fig.13b shows macro-fractograph of the LSP 3 specimen, where the curvature of the fatigue crack front can be observed. Therefore, the effects of residual stresses influence the crack growth in LSP-treated specimens, where the fatigue crack length increases with the depth; it is the largest in the middle of the specimen. The curvature in crack front can be explained through crack retardation at the sides in the presence of surface compressive residual stresses. In contrast, tensile residual stresses have an adverse effect, because they reduce the effect of crack closure.

Fig. 13b and d show the detail of fatigue crack surface at near-threshold stress intensity and transition to fracture surface, which occurred during fracture toughness test. Fractographic observation of the surface where the static fracture toughness test was conducted showed typical dimple rupture, caused by a process known as microvoid coalescence [33].

At the near-threshold regime of crack propagation, the fatigue fractographic surfaces exhibited no typical fatigue striation, which is common for stage I fatigue fracture surfaces [33]. The important phenomenon, which in turns influences the visual appearance of fatigue crack surface, is the crack closure. The crack surfaces in contact during fatigue define the texture and crack surface roughness. As mentioned earlier, mainly roughness-induced crack closure occurs at near-threshold of stress intensity range in aluminium alloys [28]. Roughness-induced crack closure is generally used to describe contact of faceted surface features which are dimensionally small, e.g. the order of the grain size [34]. Based on the visual appearance of fatigued crack surfaces on SEM micrographs, Fig. 13b and 13d at near threshold, that base material has faceted and rounded cleavage planes, followed by crack

arrest at ΔK_{th} of 3.3 MPa $\sqrt{m}$. In contrast, LSP 3 specimens has finer and smoother asperities with fatigue tear ridges oriented in the direction of crack growth in the region where crack growth stopped at comparatively much lower threshold value of 1.1 MPa $\sqrt{m}$. The smoother surface suggests that lower energy and more brittle cleavage fracture occurred along crystallographic planes in LSP-treated specimens. Moreover, visual macro-scope examination (optical microscope Olympus) confirmed that on LSP 3 specimen the central part of crack surface appears to be smoother, especially in the direction of crack growth. In order to support the qualitative visual examinations, the roughness on crack surface was measured using a contact profilometer (Taylor Hobson) and a micrometer sliding table stroke. The area of measurement of roughness is a rectangular area of 2.5 $\times$ 2.5 mm in the central part of the crack surface, as shown on Fig 14a. The number of measurement lines for each direction on an individual specimen was eight, measured with offset increments of around 350 μm and a measuring line length of 2.5 mm. From these values, the average roughness value R_a and standard deviation were calculated and presented in a histogram, Fig. 14b. From the result it can be depicted that the average roughness in z-direction is similar (average R_a for z-direction is between 16.2 – 17.5 μm and with an average standard deviation of $\pm 2.45 \mu m$). Moreover, overall surface roughness R_a in z-direction is 33–54 % higher than in the x-direction, mainly as the consequence of grain elongation in the rolling direction during plate fabrication.

Histograms reveal that the average of crack surface roughness R_a in x-direction is higher in the non-treated specimen than all in three LSP-treated specimens. However, the standard deviation is high and, therefore, it is not sufficient to conclude before performing a statistical analysis. The double-sampling hypothesis test with Student's t-distribution was used to determine whether the x-direction roughness of the crack surface in the not-treated specimen is statistically different from LSP-treated specimens. The basis for the hypothesis test is roughness measurement data, given in Table 6. To reject hypothesis of sample equality mean,

that is $H_0 : \mu_1 = \mu_2$ the value of t_0 in and $t_{\alpha/2, n_1+n_2-2}$ is compared. If $|t_0| > t_{\alpha/2, n_1+n_2-2}$, the H_0 is rejected and if $|t_0| < t_{\alpha/2, n_1+n_2-2}$ the H_0 hypothesis is accepted. The value of statistic was calculated as:

$$t_0 = \frac{\bar{y}_1 - \bar{y}_2}{S\sqrt{1/n_1 + 1/n_2}}, \quad (11)$$

where $\bar{y}_1$ and $\bar{y}_2$ are average value of the samples, n_1 and n_2 is number of measurements, S is the variance estimator for both samples; it is calculated:

$$S^2 = \frac{(n_1 - 1)S_1^2 + (n_2 - 1)S_2^2}{n_1 + n_2 - 2}, \quad (12)$$

where S_1^2 and S_2^2 are variances of samples. Furthermore, for the significance level of $\alpha=0.05$ and from Student's t-distribution, $t_{\alpha/2, n_1+n_2-2} = t_{0.025, 14} = 2.1448$ is obtained [35]. Thus, from equation (11) statistic for LSP 1 amounts $t_0=2.83$, for LSP 2 $t_0= 3.58$ and, as expected lowest value in LSP 3 specimen, $t_0=2.389$. In all three cases, the H_0 hypothesis is rejected, which means that the crack roughness in x-direction in the not-treated specimen is significantly higher than the roughness in the x-direction of LSP-treated specimens. However, one should be aware that the measured profile of contact profilometer does not perfectly follow the micro-asperities of cracked surface. Nevertheless, the crack surface roughness analysis confirms the visual qualitative SEM and macro-scope examination findings.

All the above results indicate that extreme caution should be made when introducing compressive residual stresses on the surface of the component, which will be subjected to dynamic loading under an HCF regime. Moreover, despite the fact that surface residual stress-measuring techniques are commonly used in practice for the prediction of fatigue life, this cannot serve as any ultimate criteria. To this end, one should note that although LSP has been widely approved as an advanced, very efficient treatment for improving fatigue life of metallic components, process parameters need to be optimized for every application and the

operations should always be performed in the optimum sequence in order to prevent any detrimental effect of the treated material.

4. Conclusions

The LSP process is an innovative new method to increase the lifetime of dynamically loaded components. Although it was expected that compressive residual stresses and refined grains in the near surface would retard the FCG rate after LSP, that was not the case here. Based on the research conducted, the following can be concluded:

- After LSP surface roughness, the parameter Ra was increased by a factor of more than 7 in comparison to the untreated material.
- TEM analysis confirmed tangled dislocations, with dislocation cells and refined grains.
- LSP-induced near-surface compressive stresses was found in LSP-treated C(T) specimens.
- The crack growth threshold value for untreated specimen was higher than after LSP treatment.
- The resonant testing frequencies were higher during testing of non-treated specimen in comparison with LSP-treated specimens at the same testing conditions.
- Crack growth rate significantly increased after LSP in both, i.e. K -decreasing test and in the stable region of pre-cracking phase.
- The fracture toughness obtained with LSP specimens were in average 28–33 % lower compared to the untreated specimen.
- Fractographic analyses in near-threshold regime of testing show linear shape of fatigue crack front and curved front in LSP-treated specimens.

- Crack surface roughness measurements reveal lower roughness in x-direction of crack growth than in z-direction. Furthermore, average crack surface roughness (in x-direction) for not-treated specimen is significantly higher in the central part of the specimen than the average crack surface roughness of LSP 1, LSP 2 and LSP 3 specimens in the selected field of testing.

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Electronic Annex

FIGURE CAPTIONS

Fig. 1. Tensile and C(T) specimen orientation in Al-plate

Fig. 2. Geometry of C(T) specimens and measuring resistance gage AMF-5

Fig. 3. Laser shock processing setup and treated C(T) specimen

Fig. 4. C(T) specimens (a) no treatment, (b) after LSP 1 treatment, (c) LSP 1 after grinding

Fig. 5. Roughness profile (a) LSP-treated C(T) specimen with detail at the vicinity of the notch, (b) roughness profiles at transition.

Fig. 6. TEM bright field images of LSP-treated specimens; (a) LSP 1, (b) LSP 2 and (c) LSP 3.

Fig. 7. Location of Vickers micro-hardness measurements

Fig. 8. Change of residual stress state, (a) as-received, (b) after LSP 3 treatment

Fig. 9. Monitored fatigue crack length and frequency as a function of number of cycles;
(a) fatigue crack length, (b) testing (resonant) frequency.

Fig. 10. Crack growth curves da/dN vs. applied ΔK , obtained with direct secant method

Fig. 11. Crack growth curves da/dN vs. applied ΔK , obtained from fitting equations

Fig. 12. Load-displacement curves

Fig. 13. Fracture surfaces after fatigue and fracture toughness test

Fig. 14. Crack surface roughness measurement (a) location of random measurements in x and z direction in rectangular area of crack surface, (b) histograms of averaged roughness results R_a with standard deviation error plot

TABLE CAPTIONS

Table 1. Average measured tensile properties of 6082-T651 plate

Table 2. Average R_a values for base material and LSP surface

Table 3. Fatigue crack growth values during a pre-cracking procedure at constant $R=0.1$ and constant $\Delta K=4 \text{ MPa}\sqrt{\text{m}}$.

Table 4. Constants for Paris and NASGRO equation for non-treated and LSP 1, LSP 2, LSP 3 treated specimens.

Table 5. Results of plane-stress fracture toughness K_{IC} test before and after LSP treatments

Table 6. Crack surface roughness measurement R_a [μm] in (x-direction), $n=8$ for the test of the equality of samples

Table 1. Average measured tensile properties of 6082-T651 plate

Direction of load	L	T
Tensile strength - R_m [MPa]	311	310
Yield stress - $R_{p0.2}$ [MPa]	296	295
Elongation - A [%]	9.7	8.5

Table 2. Average R_a values for base material and LSP surface

Specimen	State	Roughness as received	Roughness after LSP R_a [μm]	Roughness after grinding R_a [μm]
W/O treatment	As received	1.15	-	-
LSP 1	LSP surface	1.16	7.30	0.37
LSP 2		1.14	10.47	0.37
LSP 3		1.15	7.92	0.38

Table 3. Fatigue crack growth values during pre-cracking procedure at constant $R=0.1$ and constant $\Delta K=4 \text{ MPa}\sqrt{\text{m}}$.

Specimen	da/dN^1 [m/cycle]	Number of cycles ²	Time ³ [min]
No treatment	$4.4 \cdot 10^{-9}$	$3.60 \cdot 10^5$	41
LSP 1	$6.9 \cdot 10^{-9}$	$2.57 \cdot 10^5$	25
LSP 2	$8.8 \cdot 10^{-9}$	$2.40 \cdot 10^5$	24
LSP 3	$12.0 \cdot 10^{-9}$	$1.90 \cdot 10^5$	22

¹Fatigue crack growth rate, estimated between 1.2 -1.5 mm of crack length of linear $a-N$ part

¹Number of cycles required to reach pre-crack length of 1.5 mm

²Time of testing to reach pre-crack length of 1.5 mm

Table 4. Constants for Paris and NASGRO equation for not-treated and LSP 1, LSP 2, LSP 3 treated specimens.

Designation	Paris		Nasgro	From ΔK_{th} to slope change				From slope change to Kc			
	C	m	ΔK_{th} [MPa $\sqrt{m}$]	C	n	p	q	C	n	p	q
Not-treated	$8.37 \cdot 10^{-9}$	3.73	3.36	$9.56 \cdot 10^{-11}$	2.75	0.93	1.2	-	-	-	-
LSP 1	$3.4 \cdot 10^{-11}$	3.72	1.4	$1.05 \cdot 10^{-10}$	2.70	0.45	0.9	$3.5 \cdot 10^{-11}$	4.10	4.5	0.6
LSP 2	$3.4 \cdot 10^{-11}$	3.73	1.4	$1.15 \cdot 10^{-10}$	2.70	0.45	0.9	$1.5 \cdot 10^{-10}$	3.57	6	0.6
LSP 3	$7.4 \cdot 10^{-10}$	3.52	1.1	$1.80 \cdot 10^{-10}$	2.70	0.45	0.9	$9 \cdot 10^{-10}$	2.50	6	0.6

Table 5. Results of plane-stress fracture toughness K_c test before and after LSP treatments

Treatment	Designation	a [mm]	$\frac{a}{W}$	$P_{\max}$ [N]	P_Q [N]	K_C [MPa $\sqrt{m}$]	$\overline{K_C}$ [MPa $\sqrt{m}$]
Not-treated (3 repetitions)	#1	19.88	0.71	2849	2620	38.72	37.8 $\pm$ 0.86
	#2	17.60	0.63	4175	3950	37.68	
	#3	18.65	0.67	3784	3430	37.00	
After laser shock treatment	LSP 1	20.30	0.72	2358	1960	29.01	-
	LSP 2	20.65	0.74	2175	1840	29.34	
	LSP 3	20.19	0.72	2058	1950	28.21	

Table 6. Crack surface roughness measurement R_a [μm] in (x-direction), $n=8$ for the test of the equality of samples

i	1	2	3	4	5	6	7	8
Not-treated	7.82	13.2	7.98	9.37	11.0	11.30	12.7	14.7
LSP 1	5.4	7.35	5.0	10.5	7.81	11.10	5.68	8.2
LSP 2	5.9	5.98	8.47	9.2	4.02	9.04	4.27	7.2
LSP 3	8.85	12.3	8.04	7.89	8.85	6.29	6.42	8.4

Figure 1

Figure 2

Figure 3

(a)

(b)

Figure 4

Figure 6

Figure 7

Figure 8

Figure 9

(a)

(b)

Figure 10

Figure 11

Figure 12

Figure 13

(a) Not-treated specimen

(b) not-treated specimen, $\Delta K_{th} = 3.3 \text{ MPa}\sqrt{\text{m}}$

(c) LSP 3

(d) LSP 3, threshold at $\Delta K_{th} = 1.1 \text{ MPa}\sqrt{\text{m}}$

Figure 14