Investigation of surface structural modifications caused by the influence of the ablative layer in Inconel718 Ni-base superalloy through laser shock peening

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Abstract

The use of a protective sacrificial layer adds to the overall processing cost of laser shock peening which is an effective surface modification technique for Ni-superalloys. This study investigates the effect of this layer on the nature of residual stress, subsurface stressed depth, and microstructural differences such as grain boundary fractions and defect densities. The subsurface largely shows a compressive stress without the ablative layer except the formation of a thin surface layer with tensile residual stress coupled with large textured grains.

Keywords: Laser shock peening, Inconel718, residual stress, Synchrotron diffraction, EBSD

1. Introduction

Laser-shock peening (LSP) is a potent surface hardening method that involves the application of intense laser pulses on the material’s surface, resulting in the generation of shock waves that propagate through the subsurface layers. The high-pressure shock waves cause plastic deformation, grain refinement, and create compressive residual stresses in the specimen surface [1,2]. LSP involves absorption of laser pulses through rapid vaporization (of the surrounding medium) which results in the formation of plasma. Confining layer (distilled water) helps in the entrapment of the plasma, the trapped plasma then creates a rearward shock wave against the workpiece generating a plastically deformed subsurface skin. The sacrificial ablative layer (generally black vinyl tape) when applied, protects the surface of the sample from thermal effect of the laser [3]. The ablative layer, however, adds to the processing cost so it becomes important to understand its role in the LSP set-up. In this study, we therefore investigate the role of this layer in influencing the residual stress pattern and the microstructural response of the surface/sub-surface peened layer using the case of a Ni-base superalloy, Inconel718 (IN718).

2. Experimental Procedure

Two commercial IN718 (for composition see Supplementary File) disc-like samples of 40mm diameter and 8.5mm thickness with mirror-finished surfaces were used as workpieces for the study. The samples underwent LSP on one of the surfaces using a Tyrida LSP machine having
Nd:YAG laser device (YS120-R200A) and following working parameters: wavelength 1064nm, pulse energy of 10J, spot diameter 2.2mm, overlapping rate 50%, and pulse duration of 18ns. Out of the two LSP’d samples, one sample was peened with an ablative layer and the other sample had no such ablative tape. Throughout the paper, the sample with and without tape are denoted as S10T and S10 respectively.

Synchrotron X-ray diffraction (SXRD) experiments were conducted at Beamline-02 of the Indus-2 synchrotron facility, RRCAT [4]. The data were recorded using a Si (111) double crystal monochromator at a specific energy of ~ 15keV (wavelength = 0.826Å). Beam size for SXRD data was 8 mm (H) × 250 μm (V) beam size, while for residual stress beam size of 300 μm × 250 μm was used. For residual stress measurement of a given peak, seven ψ values were taken from 0-60° at an interval of 10° [5]. For obtaining the depth of the residual stress-affected surface, the surface of the samples was electropolished (leading to a removal of ~1000 μm) and the RS measurements were performed as usual. This small area electropolishing did not relieve the locked-in residual stress by more than 5% of the measured value [6].

To evaluate the residual strains, average crystallite size, microstrain, and defect densities caused by laser shock peening, the full profiles of the LSP samples were acquired. The peaks were fitted (Lorentzian function) and further treatment was performed by the modified Williamson-Hall method. Electron backscatter diffraction (EBSD) was carried out using JEOL7800F on the samples over (i) a 400μm x 400μm area on the laser shock peened surface and (ii) a 160μm x 160μm area scan along the depth of each sample with four measurements done along the depth. To study the sub-surface residual stress depth, nanoindentation studies were performed with an ASMEC Universal Nanomechanical Tester across the cross-section of the peened samples. A test load of 100mN, a dwell time of 10 seconds, and a loading rate of 5mN/s were selected using a Berkovich indenter.

3. Results and discussion

Figure 1(a) shows the full SXRD profiles obtained from the LSP’d surfaces of the two samples. The RS values of LSP surfaces, using the sin²ψ method, for S10T and S10 were obtained to be opposite in nature -694 MPa (compressive) and +496MPa (tensile) [7,8]. Figure 1(b) compares the depth profiles of residual stresses in S10T and S10 samples. The S10T sample exhibited very high compressive stress on the surface, gradually mitigating to a nominal compressive stress (~200MPa) up to a depth of 400μm; but importantly the RS remains compressive until the tested depth of 1000μm. In contrast, S10 displayed very high tensile surface stress (~496MPa) this could be potentially due to the heat generated on the sample surface without the presence of the tape. However, the RS stress plummets and becomes compressive from a depth of 100μm subsurface to counterbalance the surface tensile stresses. Following that, again the nature of RS remains compressive to a depth of 1000μm. This basically demonstrates that the surface heating effect is overcome rather promptly by the mechanical impact of the peening process which induces deep compressive residual stresses. Importantly, in both the samples, the residual depths go beyond 1mm, beyond which the experimental recording of data was difficult owing to the shadowing effect upon tilting, caused by ridges due to material removal by electropolishing. For detailed microstructural characterization of the specimens from two conditions the peaks from the full profiles (Figure 1(a)) were fitted and ΔK vs K 1/2 plotted (Figure 1(c)). The detailed methodology following the modified Williamson-Hall technique is given in the Supplementary file. The calculated average crystallite size, microstrain present in the microstructure, and the dislocation density responsible for the microstrain were estimated.
It is evident from the table in Figure 1(d) that there is not much difference in the crystallite sizes, microstrain, and defect densities in the two samples. This indicates that the effect of LSP in imparting sub-surface residual stress with high dislocation densities for both cases is similar and the formation of the recast layer for the S10 sample entirely occurs on the surface with a depth shallower than the interaction depth (18μm at 15keV) of the incident x-ray from where the signal is obtained.

Figure 2(a-a’) and Figure 2(b-c’) show the microstructure and surface topography of the LSP’d surfaces of the two samples respectively. The outline of the laser spots with the surrounding surface contours is clearly visible. The surface topography/roughness shows a variation of ~40μm for S10T, while S10 showed a much larger (about three times) variation in the local topography (refer to the scale of Figure 2(b-b’)). The EBSD inverse pole figure maps of the same surfaces are shown in Figure 2(d-d’). It was observed that S10T had a much more refined and similar grain size on the surface, ~8.7μm[9]. However, S10 had a notably larger grain size of ~66μm. This significantly large grain size clearly indicates the substantial surface heating of the laser without the protective covering of the ablative tape. This result also correlates with the tensile surface RS in S10 below which lies a compressive stress-hardened layer. As a result of the formation of a tensile-compressive region near the surface, two variants of the Ni-phase lattice parameter are noted in the SXRD pattern of S10, manifested by an asymmetric peak pattern (inset of Figure 1(a)). Interestingly, a closer inspection shows the surface grains in S10 are highly textured (predominantly along <001> direction as shown in 2(d’) (inset)) in the absence of adhesive tape. This texturing effect can possibly indicate melting of the LSP surface in this case, causing the grains to reorientate with a certain texture upon resolidification. Nevertheless, no preferred texture was observed in S10T that displayed random fine grains; these are also the ones where the surface RS was highly compressive.

Figure 3 represents the variation in the fraction of low angle (LAGB) and high angle grain boundaries (HAGB) along the depth of S10T and S10 corresponding to EBSD maps obtained from the depths of the sample (Supplementary file). It is evident that in both cases, the HAGB fraction is highest close to the surface which is indicative of the highest degree of grain refinement near the surface as a result of peening irrespective of the presence of the ablative layer. The negligible LAGB fraction suggests the lack of subgrains inside the fine grains despite the presence of high dislocation densities in the LSP samples; this could be attributed to the short time of the laser pulses (few ns) preventing the dislocation rearrangement to occur.

The subsurface profile of S10T shows a peak hardness on the surface (~350HV) with a gradual taper into the substrate as shown in Figure 4. This gives an estimation of the residual stress hardened depth of ~2.2mm. In the absence of the ablative layer (S10), the surface showed a lower hardness which corroborated with the large grain sizes observed on the surface of this sample due to laser heating. A peak hardness is reached at ~0.5mm from the surface and the compressive RS depth was observed up to 1.5mm depth.

4. Conclusions

This study investigated the microstructural differences that occur in LSP treated surface of Ni-superalloy IN718 with and without the ablative layer. In the case of regular protection of the layer, compressive stress is generated on the surface and extends up to a depth of ~2.2mm. In the absence of the layer, a recast layer of ~0.5 mm depth is formed over the surface characterised by coarse grain and tensile residual stress. But a compressively stressed layer is observed underneath the layer. A high dislocation density coupled with refined grain structure is observed in both cases.
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References

Figure 1: (a) SXRD profiles for IN718 laser shock peened at different conditions, (b) depth profiling of residual stress of LSP samples. (c) modified W-H plots for crystallite size, microstrain, and dislocation density calculation tabulated in (d).
**Figure 2:** (a-c) and (a-c') show the surface morphology of the laser shock peened samples. (d-d') shows the IPF maps and the average grain size of the samples, inset in (d') shows the pole figure of S10 at the surface.
Figure 3: Plot depicting the fraction of LAGB, HAGB in cross-section of the sample across the specified depth

Figure 4: Hardness profile along the depth of samples