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Combined laser shock peening and shot peening for improving fretting wear and low-cycle fatigue performance of Ti-6Al-4V

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ABSTRACT

Ti-6Al-4V components operating under cyclic contact are susceptible to concurrent fretting wear and fatigue damage. In this work, shot peening (SP), laser shock peening (LSP), and sequential LSP + SP treatment were comparatively investigated to clarify their distinct effects on fretting wear (FW) and low-cycle fatigue (LCF) behavior through changes in surface integrity and microstructure. The sequential treatment generated a graded deformation structure, in which the strongest grain subdivision and lattice distortion were confined to the outermost surface region, while elevated misorientation persisted into the subsurface. This gradient microstructure resulted in the highest near surface hardness and compressive residual stress, while preserving the depth of the stress affected layer characteristic of LSP. Under FW conditions, LSP + SP reduced the wear volume by 31.3% relative to the untreated alloy. Wear scar analysis indicated that this improvement was associated with the coupling of a more stable tribolayer and a subsurface with gradient hardening, which together suppressed tribolayer spallation, abrasive wear, and delamination. Under LCF conditions, LSP + SP increased fatigue life by 150.1%, displaced crack initiation into the subsurface, and reduced fatigue striation spacing, indicating retardation of early crack growth. This work demonstrates that the gradient surface structure induced by LSP + SP can simultaneously enhance FW resistance and LCF performance through distinct damage resistance mechanisms.

1. Introduction

Owing to their high specific strength, good thermal stability, and superior corrosion resistance, titanium alloys are extensively applied in critical engineering fields such as aerospace, automotive, and biomedical engineering [1–3]. Despite these advantageous characteristics, the service reliability of titanium alloys in rotating machinery components (e.g., compressor blades) remains compromised by premature fatigue or fretting wear (FW) failures under cyclic loading conditions [4–7]. Recent advances in surface engineering technologies have demonstrated significant potential in addressing this limitation through controlled modification of near surface microstructure. Prominent techniques, including mechanical shot peening (SP), ultrasonic surface rolling process (USRP), and laser shock peening (LSP), have emerged as effective strategies for enhancing fatigue and wear resistance by generating

beneficial compressive residual stress (CRS) and surface work hardening effects [8–12].

As one of the most established surface mechanical treatment techniques, SP has been extensively used to improve the wear and fatigue performance of titanium alloys. Zhang et al. [9] investigated the FW behavior of shot peened Ti-6Al-4V alloy under oil lubrication with varying sliding displacements and frequencies. They showed that surface hardening (433.5 HV), CRS (−303.5 MPa), and a micro dimpled texture synergistically reduced the friction coefficient and enhanced wear resistance by 72.2% at a displacement of 20 μm, mainly through improved lubrication retention and a transition in the wear mechanism from fatigue/delamination to localized abrasive wear. Kumar et al. [13] investigated the effect of ultrasonic SP on the low-cycle fatigue (LCF) behavior of Ti-6Al-4V alloy, showing that surface nanostructuring combined with CRS enhanced fatigue life by up to fourfold at low strain

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amplitudes ($\pm 0.60\%$) through delayed crack initiation and suppressed cyclic softening. However, achieving deeper CRS layers by amplifying treatment intensity or increasing coverage presents technical challenges, as these approaches tend to exacerbate surface imperfections such as roughness elevation and microcrack propagation [14,15]. Consequently, SP demonstrates limited effectiveness in improving component fatigue resistance when applications demand both substantial CRS penetration depth and a high quality surface finish [16,17].

LSP is an advanced surface treatment method that uses high energy laser pulses to generate plasma induced shock waves, resulting in plastic deformation within the material's subsurface region [18,19]. Unlike conventional SP, this technique demonstrates superior capability in producing deeper CRS distributions, typically ranging from 0.5 to 2 mm in penetration depth. Such subsurface modification is widely recognized as a key contributor to enhanced fatigue resistance in metallic components. Dai et al. [20] studied LSP treated 7050 aluminum alloy hole structures and found that LSP introduced dislocation tangles, low angle grain boundaries, and grain boundary precipitates, which enhanced compressive residual stress stability, thereby delaying fatigue crack initiation and propagation. Niu et al. [21] demonstrated that combined nanosecond and femtosecond laser peening enhanced the fretting fatigue life of Ti-6Al-4V dovetail joints at 500 °C by introducing hardened layers, residual compressive stresses, and high density dislocations, reducing surface crack initiation and propagation rates. He et al. [11] examined the effects of LSP on gradient microstructure evolution and fatigue performance of TC21 titanium alloy, revealing that LSP promoted phase transformation-driven twins in α -Ti, formed gradient dislocation structures, and enhanced fatigue life by 161% through residual compressive stress and refined microstructures. However, because LSP involves less intensive near surface mechanical deformation than the multidirectional impacts of high velocity shots during SP, it generally produces lower surface CRS and hardness, limiting its surface strengthening effectiveness [16,18].

Both surface modification techniques present inherent constraints: conventional SP remains confined to shallow CRS penetration depths below 0.5 mm, whereas LSP generates surface CRS with relatively limited magnitude. This technological gap has driven the development of hybrid LSP + SP surface engineering strategies that combine deep CRS penetration with enhanced surface integrity, thereby achieving performance improvements unattainable through individual processes, as reported in previous studies [22–24]. Wang et al. [25] explored the effects of LSP + SP treatments on Ti-6Al-4V alloy, finding that this hybrid treatment enhanced fatigue strength by 35% at 10^7 cycles through synergistic improvements in surface microhardness ($+20\%$), deeper residual compressive stress fields (640 μm depth), and grain refinement, which collectively improved resistance to foreign object damage and crack initiation. Luo et al. [16] explored the joint application of LSP and SP to Ti-6Al-4V alloy and found that the LSP + SP hybrid process extended fatigue life by 9.3 times. This enhancement was attributed to the improved distribution of CRS and the development of a microstructure with a dislocation gradient, both of which restrained crack initiation and growth. Recent studies by Yin et al. [26] and Wang et al. [27] showed that hybrid LSP + SP treatment can markedly improve the fretting fatigue resistance of Ti-6Al-4V dovetail joints by introducing deeper compressive residual stress, enhanced microhardness, and a gradient grain-refined layer, which together reduce crack initiation sources and retard early crack propagation. Taken together, these findings confirm the synergistic effectiveness of LSP + SP treatment in improving the fatigue resistance of titanium alloys.

Despite increasing interest in sequential LSP + SP treatment, its effects on Ti-6Al-4V alloy have not been fully clarified. Existing studies have mainly focused on fatigue improvement, particularly high cycle fatigue, whereas the low cycle fatigue response has received much less attention. Moreover, the role of the gradient surface structure induced by sequential LSP + SP in fretting wear damage evolution has rarely been reported. Accordingly, this study comparatively investigates Ti-

6Al-4V specimens in the untreated, SP, LSP, and LSP + SP conditions, with the aim of separately correlating surface integrity and microstructural evolution with FW behavior and LCF performance. The objective is to clarify how the gradient deformation structure introduced by LSP + SP influences tribolayer evolution during FW and crack initiation and early propagation during LCF.

2. Experimental materials and methods

2.1. Materials and surface treatment

The Ti-6Al-4V titanium alloy investigated in this study exhibits an ultimate tensile strength of 975 MPa, a yield strength of 910 MPa, and a Poisson's ratio of 0.33. Fig. 1 presents the electron backscatter diffraction (EBSD, Oxford NordlysMax3, UK) characterization of the as-received Ti-6Al-4V substrate. The inverse pole figure (IPF) map in Fig. 1(a) reveals a homogeneous fine-grained microstructure consisting predominantly of equiaxed grains with broadly distributed crystallographic orientations, indicative of an overall weak texture. The corresponding grain size distribution is right-skewed, with most grains falling within the below 8 μm regime and a mean grain size of 2.80 μm , confirming the fine-grained nature of the initial substrate. As shown in the kernel average misorientation (KAM) map in Fig. 1(b), the microstructure is dominated by low local misorientation, with an average KAM value of 0.65°, suggesting limited intragranular lattice distortion and a low level of stored deformation. Such a uniform microstructure with limited stored deformation provides a robust reference for delineating the grain refinement, lattice distortion, and strain accumulation induced by subsequent surface treatment.

The Ti-6Al-4V titanium alloy bars were machined into flat plate specimens and tensile bar specimens through the same wire electrical discharge machining and precision grinding procedures for fretting and LCF tests, respectively, to ensure a consistent initial surface condition before surface treatments, with the final surface roughness controlled at $R_a = 0.25 \pm 0.05 \mu\text{m}$. As shown in Fig. 2, both types of specimens were subjected to SP, LSP, and LSP + SP treatments. SP treatment was performed using a customized pneumatic SP system manufactured by Speen Metal Co., Ltd. (Suzhou, China). The SP treatment was conducted at an Almen intensity of 0.25 mma using S110 steel shots with a nominal diameter of 0.3 mm. A shot flow rate of 5 kg/min and a coverage of 150%. LSP treatment was conducted using a laser shock peening system designed by the AVIC Manufacturing Technology Institute. The LSP system mainly consists of a Q-switched Nd:YAG laser, an external optical path system, a control unit, a multi-axis robotic arm, and a water-confinement device. The laser operated at a wavelength of 1064 nm, with a pulse width of 15 ns. The laser beam was configured with a $4 \times 4 \text{ mm}$ square spot pattern and a 15% scanning overlap ratio [28]. The system delivered 30 J per pulse (corresponding to 12.50 GW/cm² power density) while maintaining surface integrity through a protective 100 μm aluminum foil sacrificial layer. To enhance laser induced shock wave pressure, a dynamically flowing deionized water confinement layer with a thickness of 1 to 2 mm was continuously applied during the treatment process. For the LSP + SP condition, LSP treatment was first applied, followed by SP treatment using the same processing parameters described above.

2.2. Surface integrity and microstructure analysis

An optical microscope (OM, VHX-6000, Keyence, Japan) was used to observe the surface morphology. A 3D optical profiler (SuperView W1, Zhongtu, China) was employed to measure surface roughness and characterize surface topography. Changes in phase constitution were characterized by X-ray diffraction (XRD, Bruker D8 Advance, Germany). The depth-dependent microhardness of all samples was evaluated using a Vickers hardness tester (Akashi MVK-H21, Japan) under a load of 100 g and a dwell time of 15 s. At each depth, three separate measurements

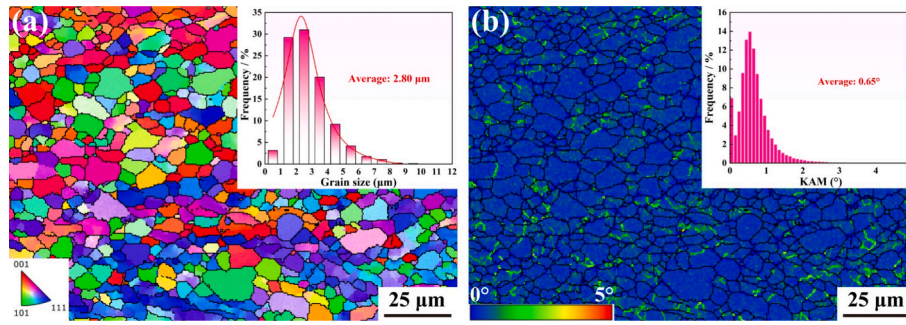


Fig. 1. Ti-6Al-4V EBSD results: (a) IPF and grain size distribution and (b) KAM.

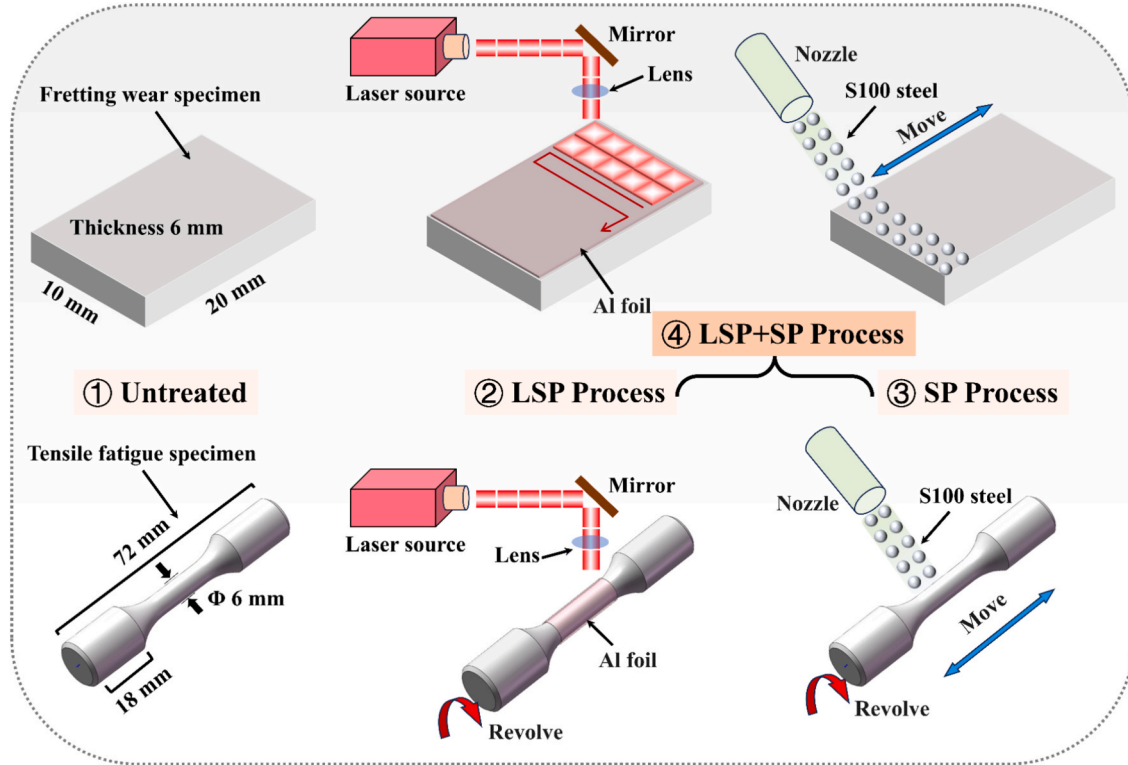


Fig. 2. Illustration of specimen dimensions and surface treatment process.

were performed, and the average value was reported. The depth-dependent residual stress was measured using an X-ray stress analyzer (XL-640, Handan Aist Stress Technology Co., China) based on the $\sin^2\psi$ method. The measurements were conducted using Cu $K\alpha$ radiation at a tube voltage of 25 kV and a tube current of 6 mA, with a 2 mm collimator. The α -Ti (213) reflection was selected for stress evaluation, and the ψ angle was varied from 0° to 45° using eight ψ tilts. The X-ray elastic constant was taken as 86.47 GPa, corresponding to $(1/2)S_2 = 11.6 \times 10^{-6} \text{ MPa}^{-1}$. Depth-dependent residual stress profiles were obtained by sequential chemical polishing. For each condition, three replicate measurements were conducted at each depth, and the average value was reported. Furthermore, EBSD analysis was conducted to investigate the crystallographic texture evolution, grain size distribution, local misorientation, and geometrically necessary dislocation density of the SP, LSP, and LSP + SP specimens.

2.3. FW and LCF tests

In this work, all specimens were subjected to FW tests using an apparatus developed in house, as illustrated in Fig. 3(a). The applied

load was controlled by weights, while a voice coil motor was employed to precisely regulate the tangential displacement amplitude. The tribological counterpart was a Ti-6Al-4V ball with a diameter of 8 mm. The tests were conducted under a normal load of 10 N, a frequency of 10 Hz, a tangential displacement amplitude of 100 μm , and a total number of 3×10^4 cycles at room temperature. Each test was repeated three times. Synchronized acquisition of frictional force and displacement signals enabled the construction of the characteristic friction force–displacement (F-D) diagram. Through mathematical integration of the enclosed area within the F-D loop, the energy dissipation (E_d) per cycle was quantified. The energy equivalent coefficient of friction (COF) and the wear rate of the samples were then calculated according to Eqs. (1) and (2) [29,30]:

$$\text{COF} = \frac{E_d}{4F_N D} \quad (1)$$

$$k = \frac{V}{4F_N D N} \quad (2)$$

The defined parameters follow this convention: F_N corresponds to

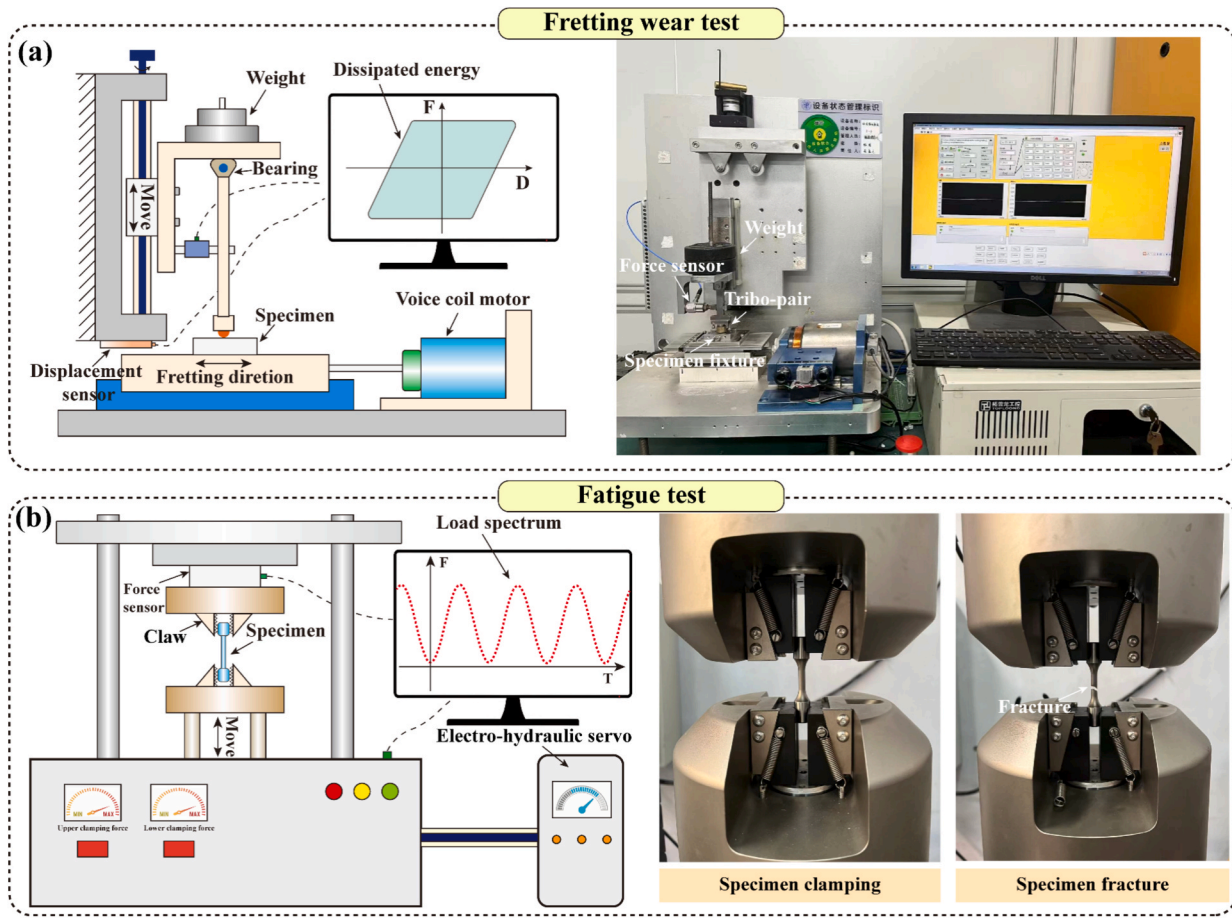


Fig. 3. Schematic diagrams and photographs of the testing equipment: (a) FW test setup and (b) LCF test setup.

applied normal load, D refers to displacement amplitude, E_d characterizes energy dissipated per cycle, V represents wear volume, and k denotes the wear rate.

LCF tests were conducted using an electro-hydraulic servo fatigue machine (SDZ-100, Sinotest Equipment Co., China). As illustrated in Fig. 3(b), the specimens were fixed between the upper and lower grips,

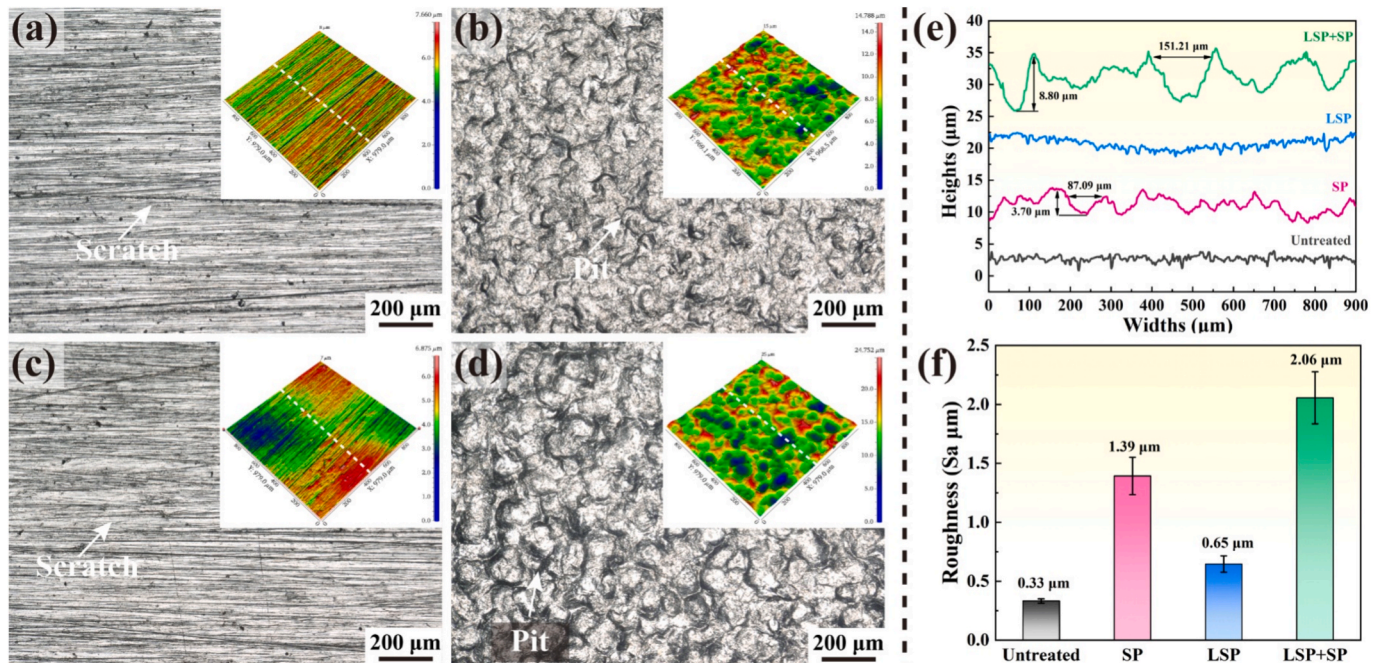


Fig. 4. (a–d) OM of untreated, SP, LSP, and LSP + SP specimens, (e) surface profile curve and (f) surface roughness.

while the tensile load was applied hydraulically from the bottom. The experiments were performed at room temperature with a peak stress of 900 MPa, a stress ratio of 0.1, a frequency of 20 Hz, and a sinusoidal waveform. Each test condition was repeated at least twice, and the mean LCF life was then calculated. After testing, the wear tracks and fatigue fracture surfaces were systematically characterized by OM, a 3D optical surface profiler, SEM, and energy dispersive spectroscopy (EDS, Aztec X-MaxN 80, Oxford, England), to clarify the mechanisms responsible for the improved FW and fatigue performance achieved by different surface treatments.

3. Results

3.1. Surface and cross-section morphology

The untreated specimens exhibit numerous surface scratches (Fig. 4(a)), which are attributed to the grinding process. The SP specimens display a characteristic peak and valley morphology (Fig. 4(b)), associated with substantial plastic deformation induced by steel shot impacts on the surface. For LSP treated specimens (Fig. 4(c)), the surface morphology remains comparable to that of untreated specimens due to the presence of an aluminum foil protective layer during LSP processing. While the LSP + SP specimens show a similar peak and valley morphology to the SP specimens (Fig. 4(d)), the peak and valley features appear notably enlarged, as confirmed by surface profile analysis (Fig. 4(e)). 3D morphology measurements reveal surface roughness S_a values of 0.33 μm , 1.39 μm , 0.65 μm , and 2.06 μm for the untreated, SP treated, LSP treated, and LSP + SP treated specimens (Fig. 4(f)), respectively. This roughness trend is consistent with prior studies [26].

Fig. 5(a) shows that the untreated specimen retains the typical duplex microstructure of Ti-6Al-4V, consisting of equiaxed primary α grains and an interconnected β phase, without apparent distortion induced by deformation. In contrast, the SP, LSP, and LSP + SP treated specimens exhibit clear near surface microstructural modification (Fig. 5(b-d)), including elongation of α grains parallel to the surface and a local reduction in the apparent grain size. These features indicate that plastic deformation induced by peening alters the original α + β microstructure through grain distortion and subdivision assisted by deformation. The

effect is more pronounced in the LSP + SP specimen, implying stronger strain accumulation under the sequential treatment. Fine acicular substructures are also locally observed within some α grains in the SP and LSP + SP specimens, which are reasonably attributed to substructural refinement induced by deformation [31,32]. These observations suggest that the combined treatment produces the strongest near-surface microstructural modification among the four conditions.

3.2. Hardness and residual stress

Fig. 6(a) shows the depth dependent microhardness profiles of the four specimens. Both SP and LSP increase the subsurface hardness, with the maximum values occurring at a depth of ~ 20 μm . Specifically, the peak hardness reaches 456.8 HV for the SP specimen and 412.5 HV for the LSP specimen, corresponding to increases of 34.2% and 21.3%, respectively, relative to the untreated condition. In both cases, the hardness gradually decreases with increasing depth. The hardened layer produced by SP extends to ~ 160 μm , whereas LSP generates a much deeper affected layer of ~ 400 μm . The LSP + SP specimen exhibits the most pronounced hardening response, with a maximum hardness of 524.1 HV, representing a 59.4% increase over the untreated specimen. Notably, within the overlapping influence zone of SP and LSP, the microhardness of the LSP + SP specimen remains consistently higher than that of either individually treated specimen. Beyond ~ 160 μm , however, the hardness profile of LSP + SP nearly coincides with that of LSP, and the total hardening depth remains ~ 400 μm . This result indicates that the composite treatment inherits the strong near surface hardening capability of SP while retaining the deeper strain affected layer imparted by LSP.

Fig. 6(b) presents the residual stress distributions as a function of depth. The untreated specimen exhibits only a shallow CRS layer, with a surface value of about -110.3 MPa and an affected depth of ~ 40 μm , which can be attributed to the prior machining process. In contrast, all surface treated specimens show substantially enhanced CRS. The surface CRS of the SP, LSP, and LSP + SP specimens are -627.7 , -540.4 , and -643.7 MPa, respectively, while the maximum CRS appear in the subsurface region at depths of ~ 10 to 30 μm , reaching -695.1 MPa for SP, -604.8 MPa for LSP, and -746.3 MPa for LSP + SP. In terms of affected

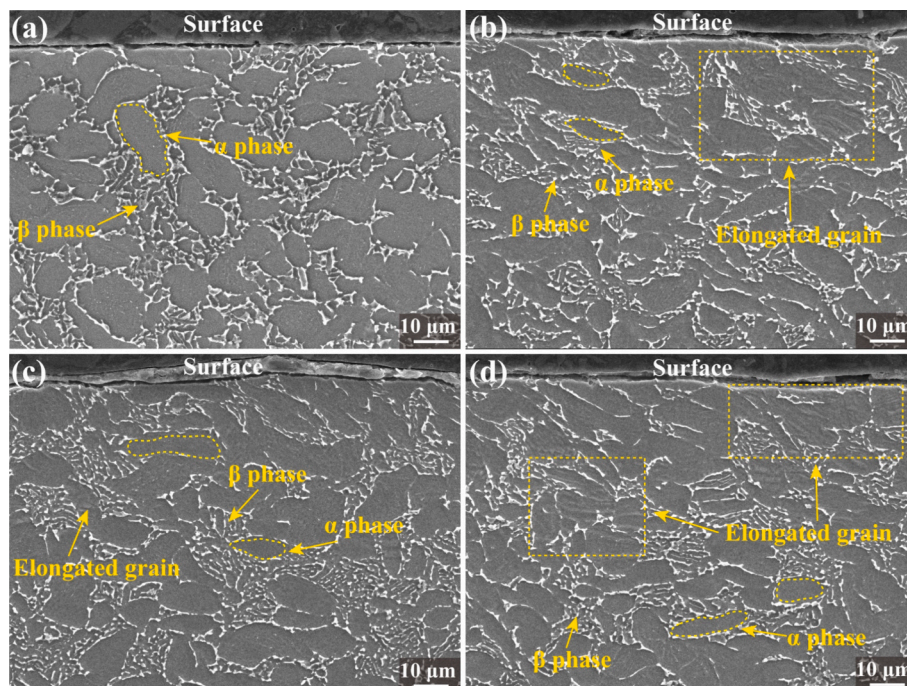


Fig. 5. Cross-sectional SEM images of the specimens: (a) Untreated, (b) SP-treated, (c) LSP-treated and (d) LSP + SP-treated.

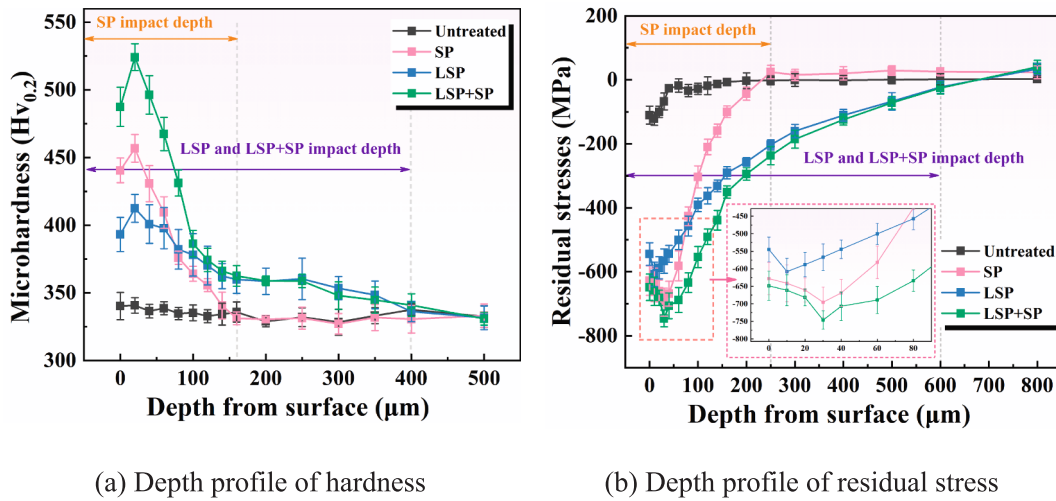


Fig. 6. Hardness and residual stress distribution statistics for all specimens.

depth, the CRS layer induced by SP extends to ~ 250 μm , whereas both LSP and LSP + SP produce much deeper profiles, reaching ~ 600 μm . These results clearly demonstrate the complementary nature of the composite treatment: SP contributes stronger near surface hardening and higher CRS, whereas LSP provides a substantially deeper affected layer [25,33].

3.3. Physical phases

The XRD diffraction peaks of all specimens are basically the same, indicating that the surface treatments do not change the phase constitution of the material, as shown in Fig. 7. The magnified XRD patterns show that the (101) and (002) diffraction peaks of the SP treated, LSP treated, and LSP + SP treated specimens shift toward lower diffraction angles compared with the untreated specimen [34]. This is can be associated with the surface grain refinement and compressive stress enhancement induced by mechanical treatment. Notably, the SP and LSP + SP specimens exhibited reduced peak intensities together with peak broadening, suggesting lattice distortion and elevated dislocation densities induced by severe plastic deformation [35,36].

3.4. FW behavior

Fig. 8(a–d) displays the friction force–displacement (F–D) curves for all specimens in the FW tests. The enclosed areas of these curves approximate parallelogram shapes, characteristic of a complete gross slip regime, which is generally associated with severe material damage caused by intense wear [37]. The COF, calculated according to Equation

(1), is depicted in Fig. 8(e). The FW process can be divided into two distinct stages. In the initial running-in stage, the COF increases rapidly, stabilizing within approximately the first 3000 cycles. This is followed by a steady state wear stage, during which the COF remains relatively constant, exhibiting only minor fluctuations. As illustrated by the bar chart in Fig. 8(e), the mean COF values in the steady state wear regime were 0.395 for the untreated specimen, 0.377 for the SP specimen, 0.371 for the LSP specimen, and 0.368 for the LSP + SP specimen. A lower COF is generally associated with improved wear resistance, which is further supported by the FW volume and wear rate data illustrated in Fig. 8(f). All treated specimens exhibited significant improvements in FW resistance compared to the untreated specimen. Notably, the LSP + SP treated specimens showed the greatest improvement, with a 31.3% reduction in wear volume relative to the untreated condition. In comparison, the SP and LSP treatments individually resulted in wear volume reductions of 8.4% and 11.2%, respectively. This superior performance of the combined LSP + SP treatment can be attributed to the synergistic effect between increased surface hardness and the distribution of CRS with depth [38].

It should be noted that the COF measured in FW tests is highly dependent on the testing configuration and parameters, rather than being an intrinsic material constant. Previous studies on Ti-6Al-4V have reported a wide range of COF values under different fretting conditions: for example, approximately 0.15 to 0.40 under ball on flat contact against an AISI E52100 steel ball at 10 to 20 N [39], 0.4225 to 0.6024 under flat on flat self-mated contact with varying temperature, load, displacement, and frequency [40]. Therefore, the COF values in this study are mainly used to compare the relative frictional responses of

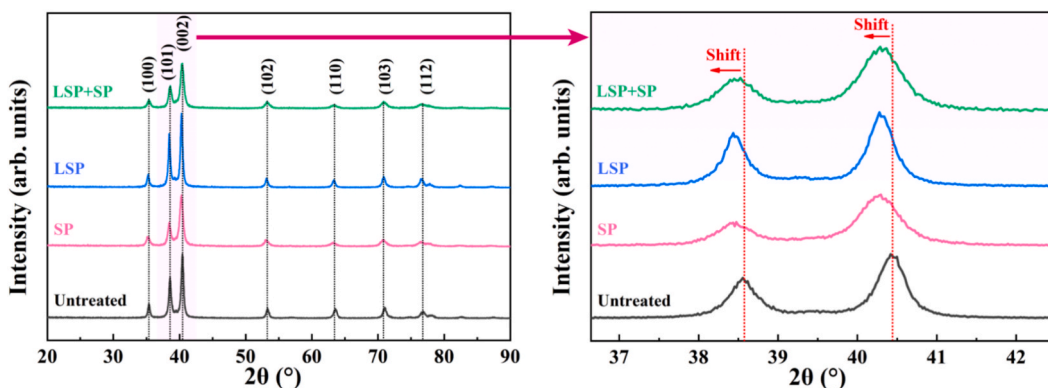


Fig. 7. The XRD patterns of all specimens are presented alongside magnified views of the (101) and (002) diffraction peaks.

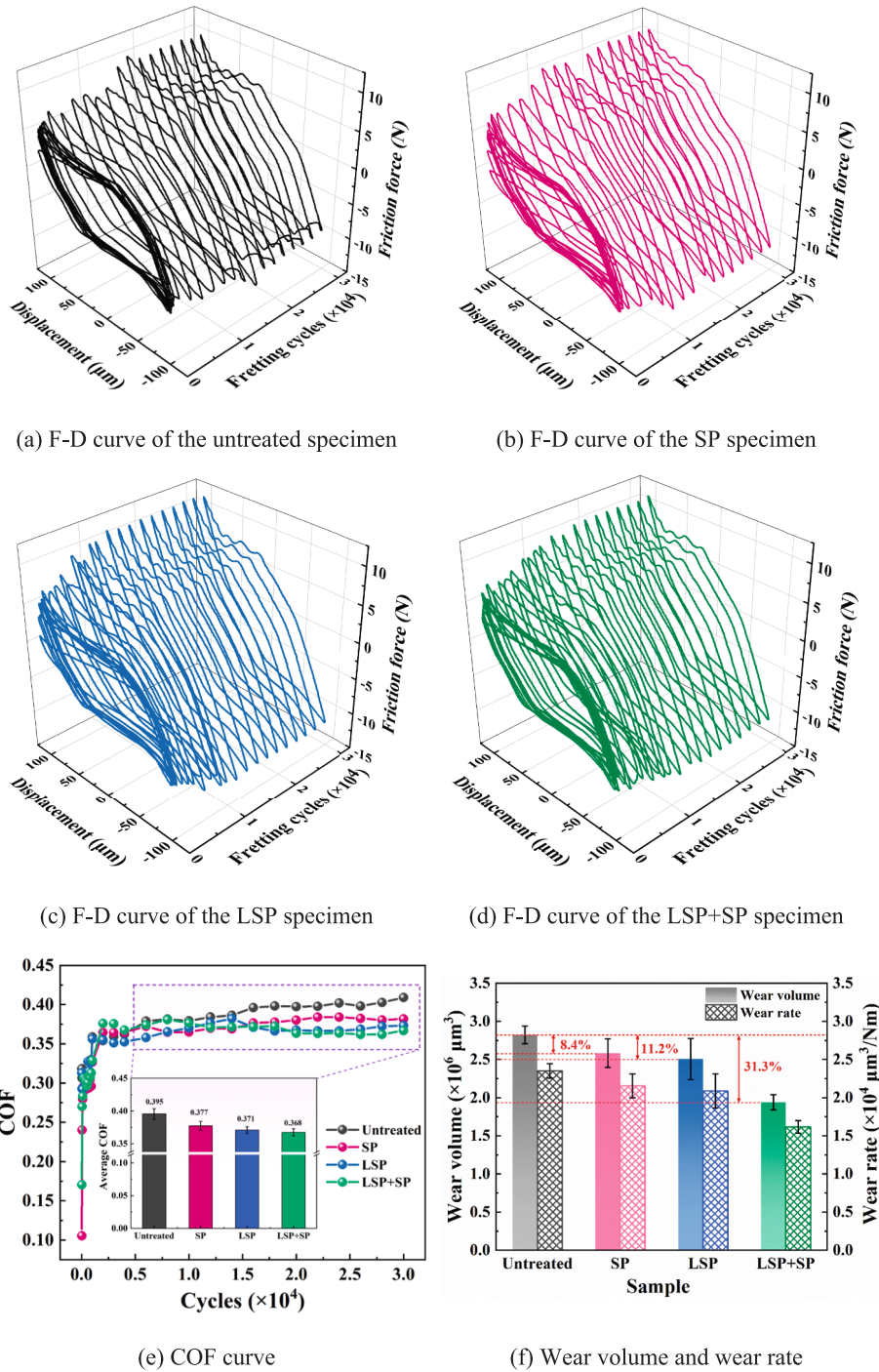


Fig. 8. F-D curves, COF, wear volume and wear rate results for all specimens.

different surface conditions under identical testing parameters.

Fig. 9 presents the comparative wear scar morphologies, three-dimensional surface topographies, and corresponding cross-sectional profiles of the untreated, SP treated, LSP treated, and LSP + SP treated specimens after FW testing. The optical morphologies reveal that a substantial amount of dark debris adhered to the wear scars of all specimens, indicating that some wear particles were not effectively expelled from the contact interface and thus participated in the wear process. The untreated specimen exhibited the largest wear scar area, with a scar width of approximately 1098.1 μm , followed by the LSP treated specimen (1031.0 μm). In contrast, the SP and LSP + SP treated specimens showed narrower wear scars, measuring 990.7 μm and 941.0

μm , respectively. This reduction in wear scar width may be attributed to the increased surface roughness and hardness resulting from the surface treatments. The three-dimensional surface morphologies reveal the presence of multiple spalling pits distributed across the wear surfaces of all specimens. The corresponding cross-sectional profiles of the wear scars are shown in the third row of Fig. 9. The untreated specimen exhibited the largest wear depth, reaching 29.49 μm , indicating poor resistance to FW. In contrast, the SP, LSP, and LSP + SP treated specimens showed reduced wear depths of 25.45 μm , 21.61 μm , and 19.18 μm , respectively. These results demonstrate that the LSP + SP treatment provides superior FW resistance, as evidenced by its smaller wear area and reduced wear depth.

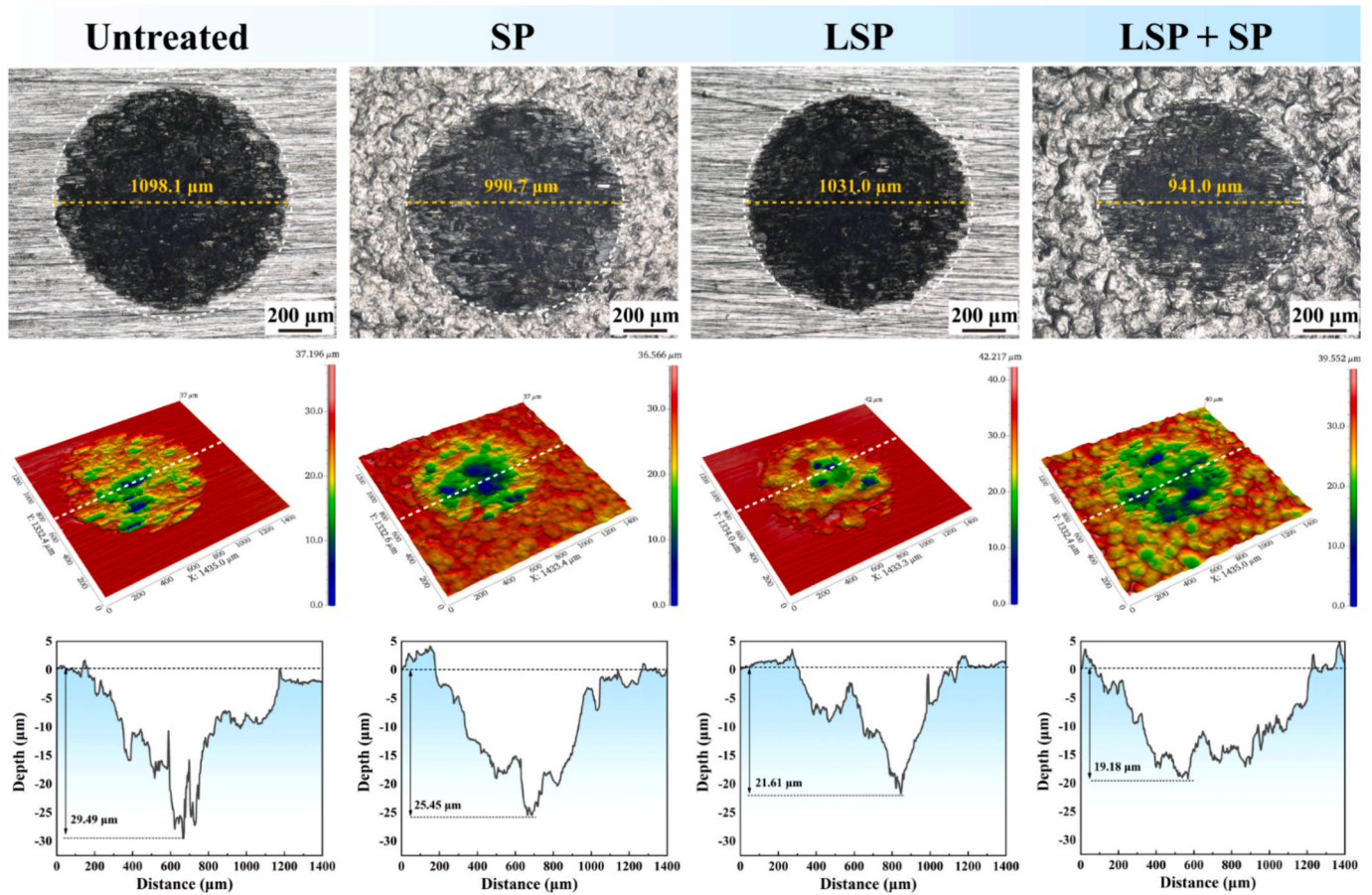


Fig. 9. Optical morphology and contour curves of wear scars for all samples.

Fig. 10 shows the SEM morphologies of the worn surfaces along with EDS analysis for all specimens. As displayed in Fig. 10(a), the worn surface of the untreated specimen displays a marked increase in oxygen content, with a mass fraction of approximately 19.1%. Magnified views of the wear scar reveal numerous grooves and debris-filled spallation pits. For the SP treated specimen, the wear scar morphology (Fig. 10(b)) shows a noticeable reduction in the size and number of spallation pits compared to the untreated specimen. Although significant delamination can be observed in the enlarged image, the remaining debris within the pits appears compact, and no pronounced grooves are visible. The LSP specimen (Fig. 10(c)) also demonstrates notable oxygen enrichment on its worn surface, though at a relatively lower concentration of 16.8 wt%. The surface morphology of this specimen closely resembles that of the untreated specimen. Significant delamination and spalling are also observed, together with grooves and wear debris within the spalling pits. Among all specimens, the LSP + SP treated specimens exhibits the most favorable wear scar characteristics (Fig. 10(d)). Its morphology is similar to that of the SP treated specimen, but with the highest detected oxygen content, approximately 21.4 wt%. High magnification images show that the worn surface retains good integrity, with only minor delamination observed.

3.5. LCF life

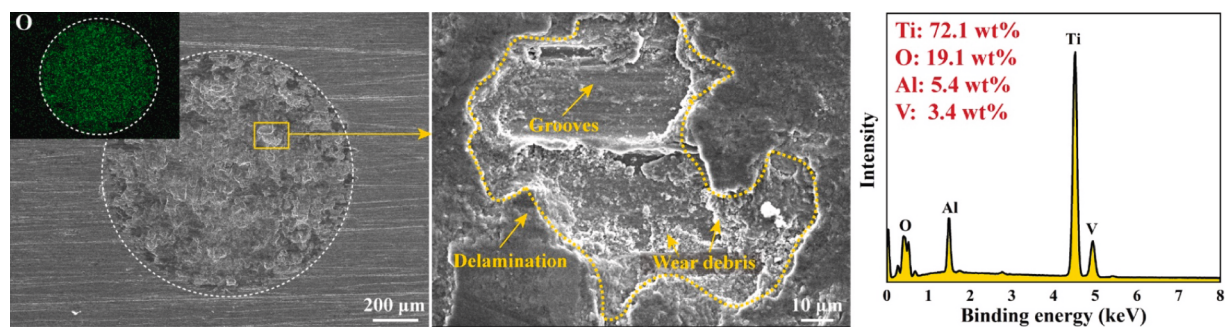
Fig. 11 displays the fatigue life and associated fracture morphologies for all specimens. Given that the applied peak stress was close to the yield strength of Ti-6Al-4V alloy, and that all specimens exhibited fatigue lives on the order of 10^4 cycles, the failure mode can be classified as LCF. The average fatigue life of the untreated specimens was approximately 10,662 cycles. In contrast, the SP, LSP, and LSP + SP

treated specimens exhibited significantly enhanced fatigue life values, averaging 16122, 22439, and 26,666 cycles, respectively. These results represent fatigue life improvements of 51.2%, 110.4%, and 150.1%, respectively, compared to the untreated condition. The observed enhancements can be attributed to the combined effects of surface hardening and the introduction of beneficial CRS, both of which act to delay crack initiation and retard crack propagation under cyclic loading. Notably, the LSP + SP specimen showed the greatest improvement, indicating a clear synergistic effect between the two surface treatments in enhancing fatigue performance.

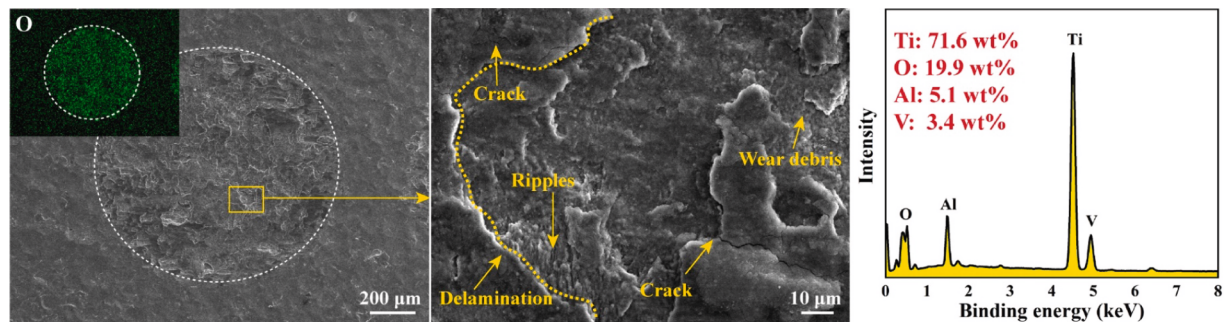
3.6. Fatigue fracture morphology

A typical fatigue fracture consists of three regions: the crack initiation region, crack propagation region, and final rupture region [16]. As shown in Fig. 12(a), multiple surface crack initiation sites can be identified on the fracture surface of the untreated specimen in the optical image. SEM analysis of a selected initiation area reveals cleavage planes and step features typical of brittle fracture. In addition, a groove like region is observed on the surface, which is evidently caused by machining scratches. During fatigue testing, such deep grooves can induce significant stress concentration and thus serve as preferential sites for crack initiation [41]. At a region approximately 1 mm from the surface, distinct fatigue striations are observed. The spacing of these striations corresponds to the crack growth increment per loading cycle. Therefore, by comparing the striation spacing in different specimens, the relative crack propagation rates can be assessed. In the untreated specimen, the distance across three adjacent fatigue striations is measured as 3.26 μm .

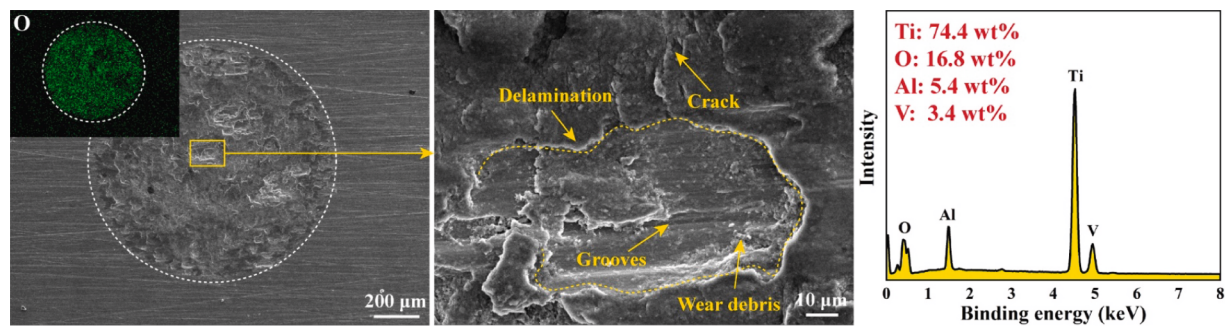
For the SP specimen, multiple crack initiation sites are also



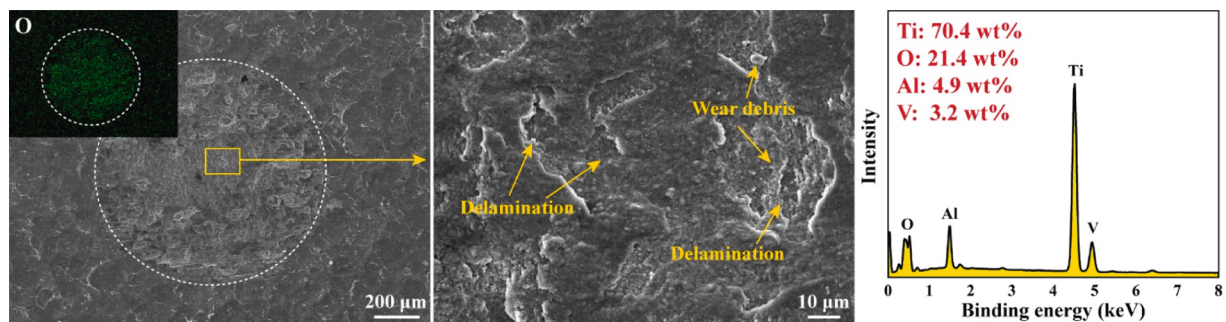
(a) Untreated



(b) SP-treated



(c) LSP-treated



(d) LSP+SP-treated

Fig. 10. Wear scar SEM morphology and EDS mapping.

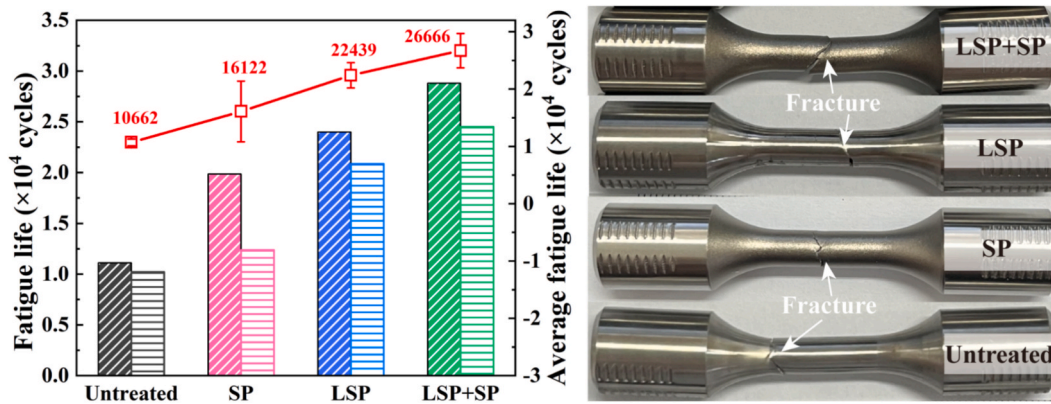


Fig. 11. LCF life and failure photos of all specimens.

identified, but these origins are located in the subsurface region, approximately 24.69 μm below the surface, as shown in Fig. 12(b). This shift to subsurface initiation is a known benefit of surface SP treatments, contributing to enhanced fatigue life by delaying crack nucleation [42,43]. Similar to the untreated specimen, the vicinity of the crack origins in the SP specimen still displays cleavage planes and step features, confirming that brittle fracture remains the dominant local failure mode. Pronounced fatigue striations are also observed in the SP specimen at a location about 1 mm beneath the surface. The distance across three consecutive striations is measured to be 5.97 μm , suggesting a higher crack growth rate at this site compared with the untreated specimen.

The fracture morphology of the LSP specimen, shown in Fig. 12(c), is similar to that of the untreated specimen in terms of crack initiation. Multiple cracks still originated from machining-induced grooves and scratches on the surface, indicating that while LSP introduces CRS, severe surface defects can still act as dominant initiation sites if not mitigated. Consistent with other specimens, brittle fracture remained the predominant failure mechanism. However, at a region about 1 mm from the surface, the fatigue striations appeared significantly denser. The measured distance across between three consecutive striations is 2.88 μm , indicating that the crack propagation rate in the LSP specimen is lower than that of both the untreated and SP specimens. This reduction in crack growth rate is likely associated with the greater depth of the CRS layer introduced by LSP.

The LSP + SP specimen, shown in Fig. 12(d), demonstrated the most favorable fatigue fracture characteristics. Although multiple crack initiation sites are still observed, these crack origins are located significantly deeper in the subsurface region, approximately 42.74 μm below the surface. This deeper subsurface initiation, compared to the SP specimen, highlights the enhanced ability of the combined treatment to suppress crack nucleation at the surface. Furthermore, at approximately 1 mm from the surface, the fatigue striations exhibited the smallest spacing among all specimens, with an average distance of 2.39 μm . This indicates the most effective suppression of crack propagation. These results collectively underscore the synergistic effect of combining LSP and SP treatments.

The final rupture regions of all specimens all exhibited a large number of microvoids and dimples, as shown in Fig. 13. The formation of dimpled fracture surfaces is likely related to the nucleation and growth of microvoids, which accumulate and coalesce under plastic deformation to accommodate local material incompatibilities [2]. The dimple size distributions for the four types of specimens were statistically analyzed, and the average dimple sizes were calculated, as depicted in Fig. 13(a-d). The untreated specimen exhibited an average dimple size of 5.48 μm . In comparison, the SP, LSP, and LSP + SP specimens displayed a significant increase in the number and density of small dimples, with average sizes of 5.15 μm , 4.44 μm , and 4.31 μm ,

respectively. Small dimples are generally associated with enhanced plasticity and are considered beneficial for retarding crack propagation [44].

4. Discussion

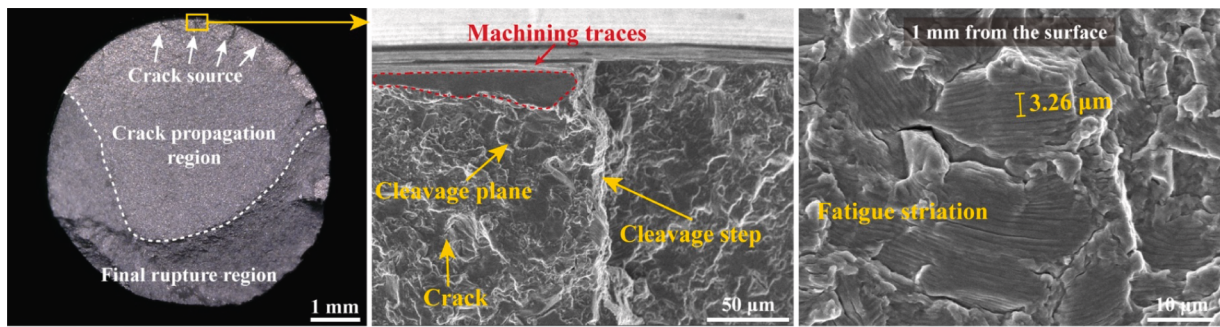
4.1. Microstructure evolution mechanisms

Figs. 14 to 16 show the cross-sectional EBSD results of the SP, LSP, and LSP + SP specimens, including IPF, KAM, geometrically necessary dislocation (GND) density, and grain orientation spread (GOS) maps, together with the grain size and KAM distributions at different depths. To enable a direct comparison among different surface treatments, the color scales of the KAM, GND density, and GOS maps were kept identical for the three specimens. Compared with the initial Ti-6Al-4V substrate shown in Fig. 1 (average grain size: 2.80 μm , mean KAM: 0.65°), the three surface-treated specimens exhibit different degrees of grain subdivision, lattice rotation, and dislocation accumulation, reflecting the distinct deformation characteristics introduced by LSP, SP, and their sequential combination.

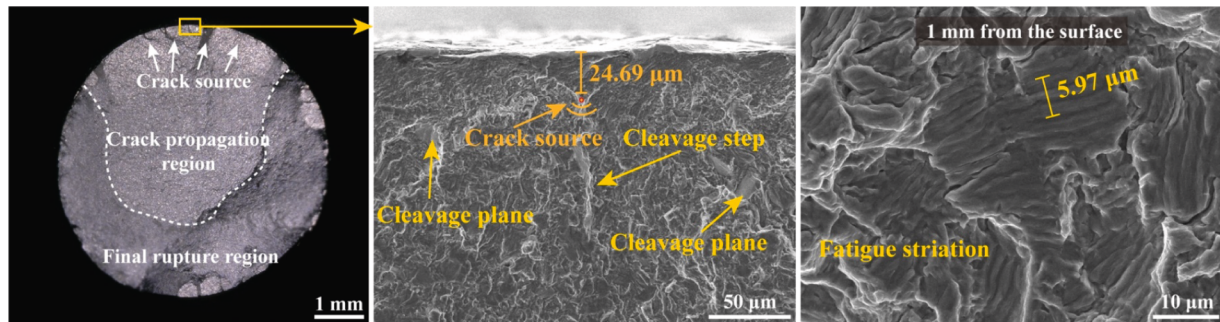
For the SP specimen, Fig. 14 shows that the microstructural modification is mainly concentrated near the surface. The average grain size in the 0–50 μm region decreases to 1.62 μm , whereas it increases to 2.49 μm and 2.78 μm in the 50–150 μm and 150–250 μm regions, respectively. The KAM distribution follows a similar depth dependent attenuation, with the mean KAM decreasing from 1.08° in the surface layer to 0.70° and 0.66° in the two deeper regions. The GND and GOS maps also show pronounced deformation features near the surface, while the subsurface regions exhibit much weaker lattice distortion. These results indicate that SP induces severe plastic deformation, grain subdivision, and dislocation accumulation in the outermost layer, but the resulting microstructural influence decreases rapidly with depth.

In contrast, the LSP specimen exhibits a less pronounced effect of grain refinement but a deeper deformation response, as shown in Fig. 15. The average grain sizes in the 0–50 μm , 50–150 μm , and 150–250 μm regions are 2.44 μm , 2.57 μm , and 2.70 μm , respectively, which are only slightly smaller than that of the initial substrate. This indicates that LSP alone does not produce severe grain fragmentation in the near surface region. However, the mean KAM values remain at 0.89°, 0.83°, and 0.69° from the surface to the 150–250 μm region, respectively. Together with the GND and GOS maps, these results suggest that LSP introduces a relatively deep field of lattice curvature and stored plastic strain, even though its ability to refine grains at the surface is weaker than that of SP. This feature is consistent with the shock wave loading nature of LSP, which promotes subsurface plastic deformation over a larger depth rather than highly localized surface fragmentation.

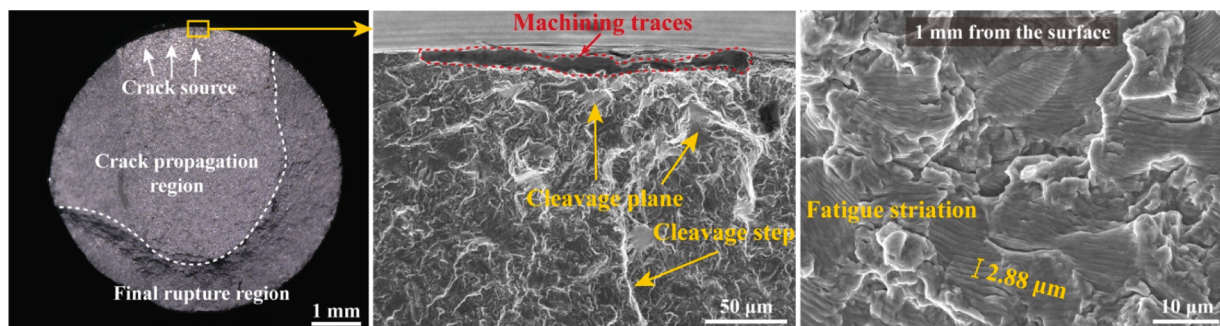
As shown in Fig. 16, the LSP + SP specimen exhibits the most pronounced gradient deformation structure among the three treated



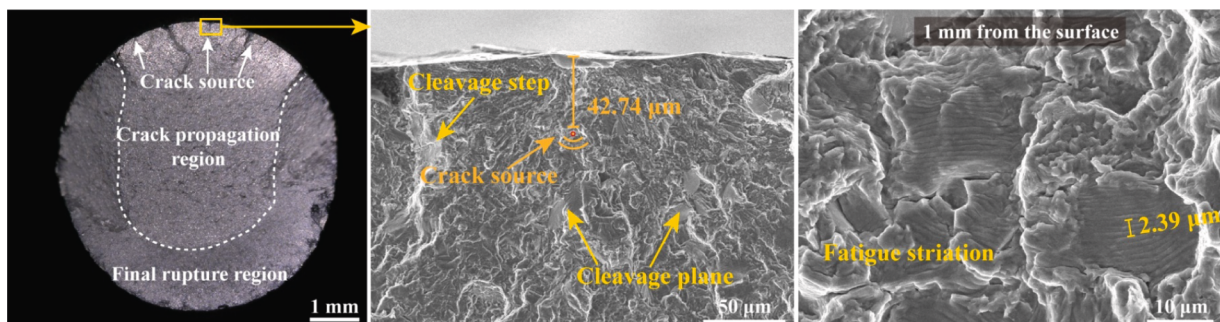
(a) Untreated



(b) SP treated

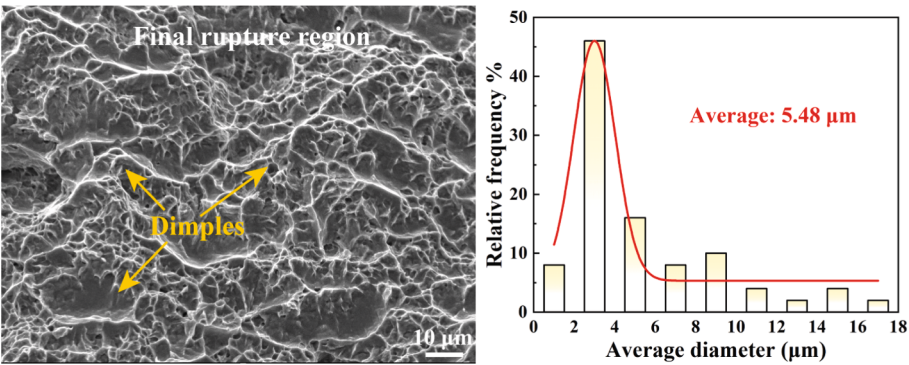


(c) LSP treated

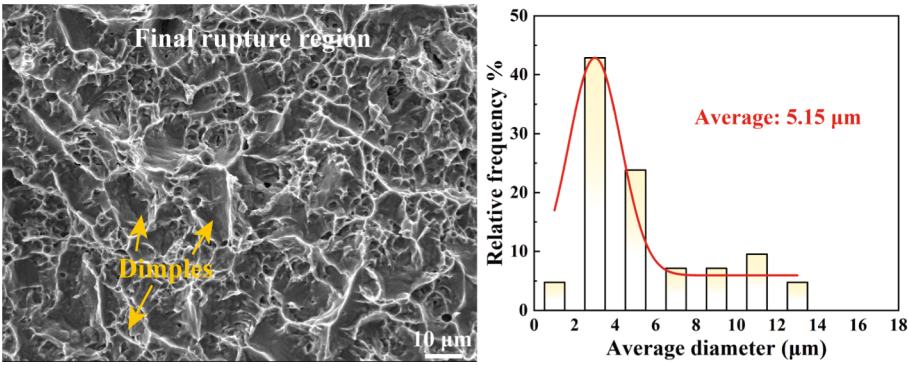


(d) LSP+SP treated

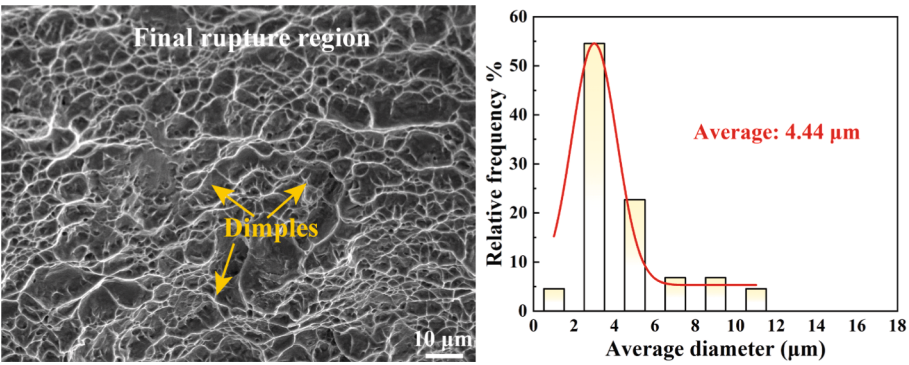
Fig. 12. Fatigue fracture optical and SEM morphology of all specimens.



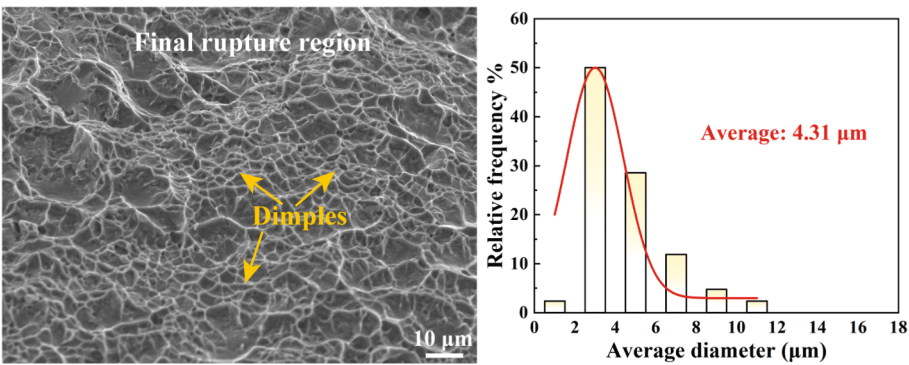
(a) Untreated



(b) SP treated

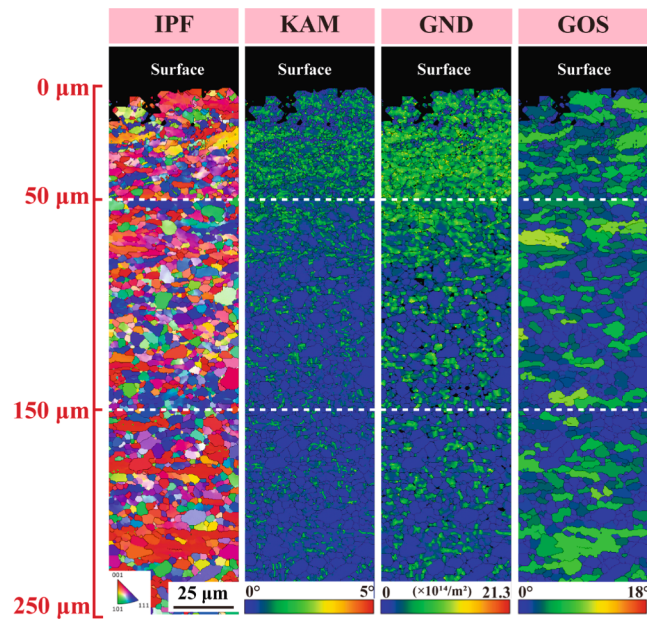


(c) LSP treated

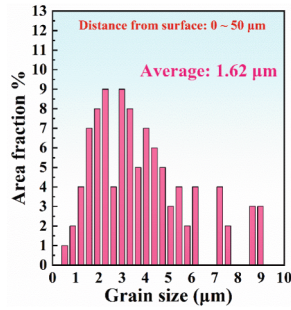


(d) LSP+SP treated

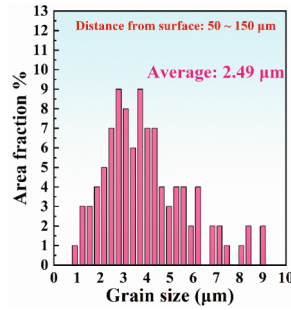
Fig. 13. SEM morphology and statistical distribution of dimple sizes in the final rupture region for all specimens.



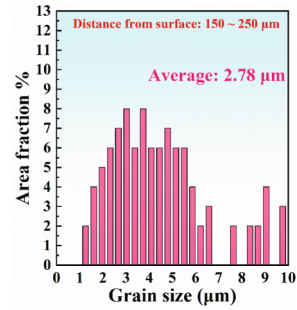
(a) IPF, KAM, GND, and GOS maps



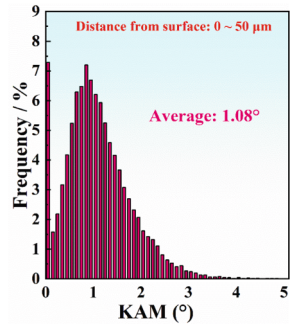
(b) Grain size within the range of 0–50 μm



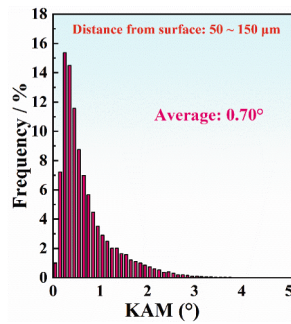
(c) Grain size within the range of 50–150 μm



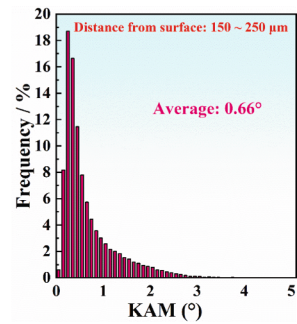
(d) Grain size within the range of 150–250 μm



(e) KAM distribution in the range of 0–50 μm



(f) KAM distribution in the range of 50–150 μm



(g) KAM distribution in the range of 150–250 μm

Fig. 14. SP specimen cross-section EBSD results.

conditions. In the 0–50 μm region, the average grain size is reduced to 1.38 μm, which is smaller than that of the SP specimen (1.62 μm) and much smaller than that of the LSP specimen (2.44 μm). Meanwhile, the mean KAM value increases to 1.22°, exceeding those of the SP (1.08°) and LSP (0.89°) specimens at the same depth. These comparisons indicate that the sequential LSP + SP treatment produces the strongest near surface grain subdivision and lattice distortion. In the 50–150 μm

region, the LSP + SP specimen still shows a smaller average grain size of 2.17 μm and a higher KAM value of 0.98°, compared with 2.49 μm and 0.70° for the SP specimen and 2.57 μm and 0.83° for the LSP specimen. This suggests that the combined treatment retains a stronger subsurface deformation response than either individual treatment. In the 150–250 μm region, the grain size of the LSP + SP specimen increases to 2.68 μm, close to the corresponding values for the SP and LSP specimens, but its

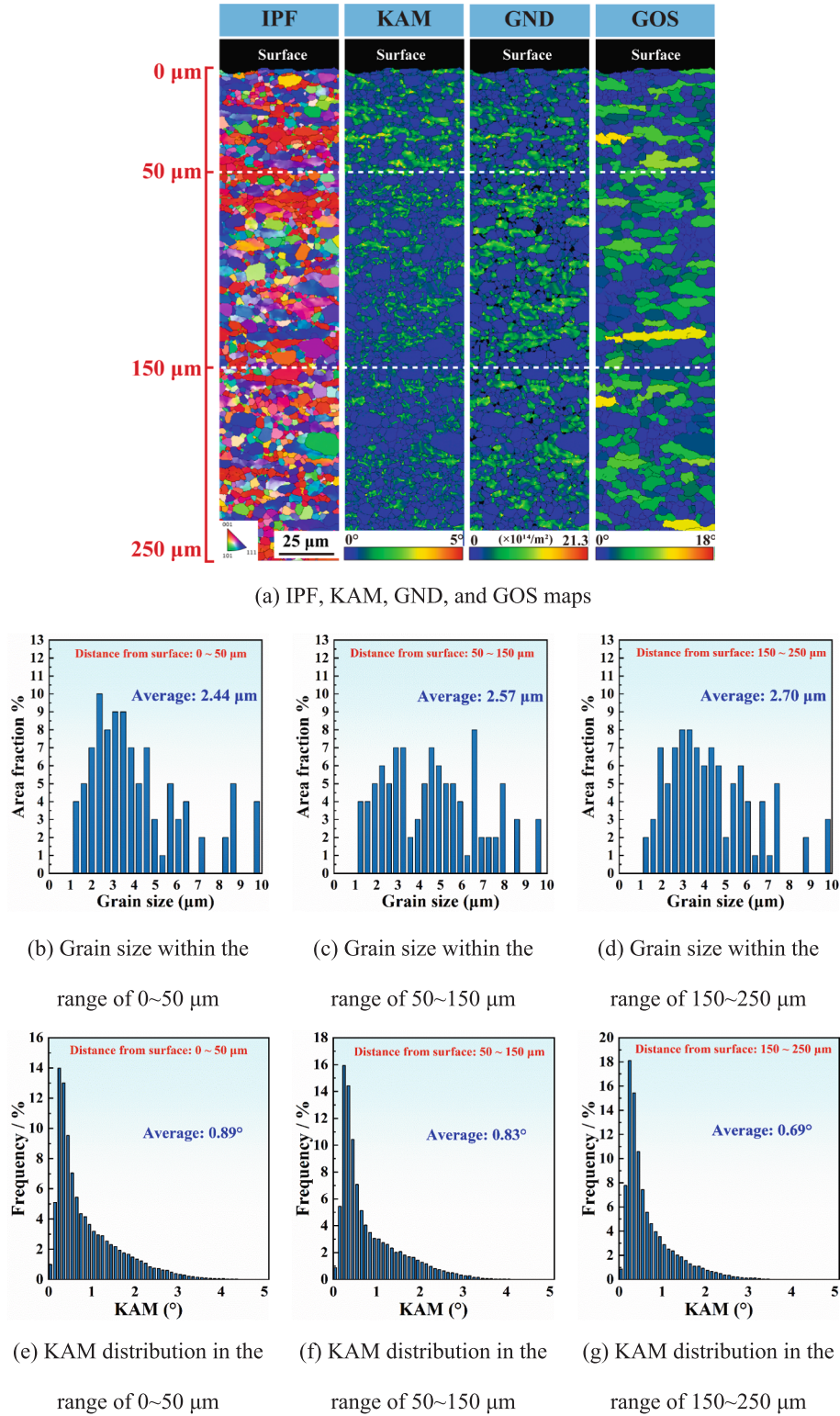
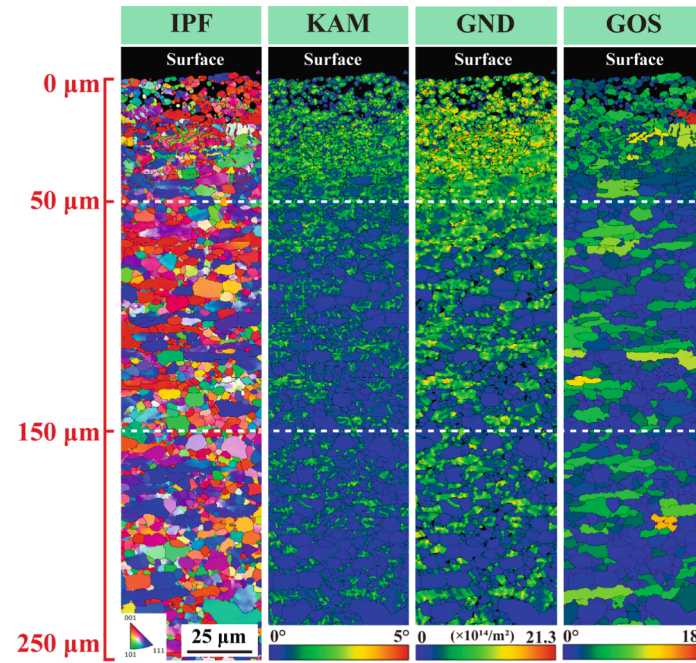


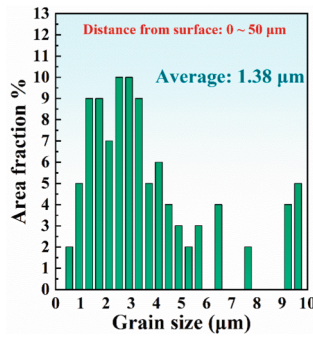
Fig. 15. LSP specimen cross-section EBSD results.

mean KAM value remains slightly higher at 0.72° , compared with 0.66° for SP and 0.69° for LSP. Therefore, although effect of grain refinement gradually weakens with depth, the LSP + SP specimen still preserves more pronounced lattice distortion in the subsurface. The GND density map further supports this conclusion, showing stronger near surface GND accumulation and a more continuous deformation gradient along the depth direction.

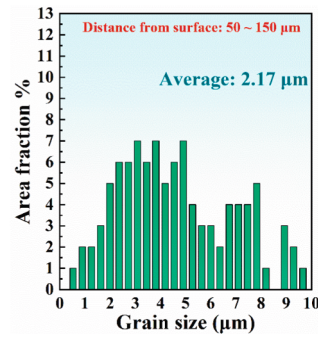
Some studies [18,45] on Ti-6Al-4V show that SP and LSP impose different modes of plastic deformation. SP is dominated by repeated near surface mechanical impacts, which promote severe plastic strain, dislocation multiplication, low angle boundary formation, and grain subdivision within a shallow surface layer. In contrast, LSP generates plastic deformation induced by shock waves over a larger depth, favoring substructures rich in dislocations, lattice curvature, and subgrain



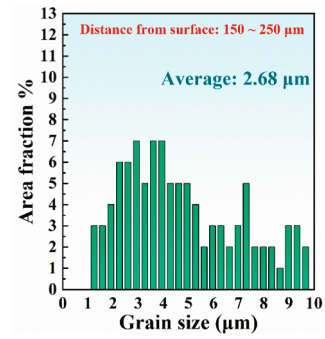
(a) IPF, KAM, GND, and GOS maps



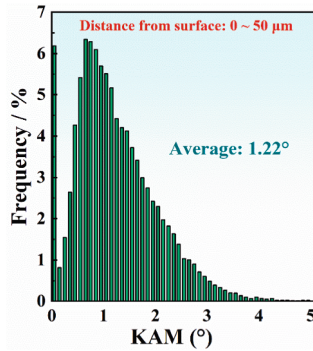
(b) Grain size within the range of 0~50 μm



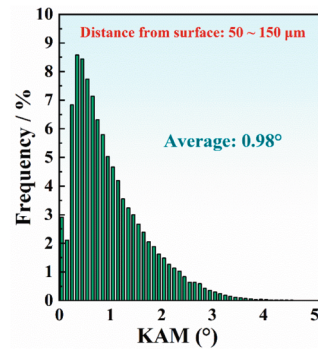
(c) Grain size within the range of 50~150 μm



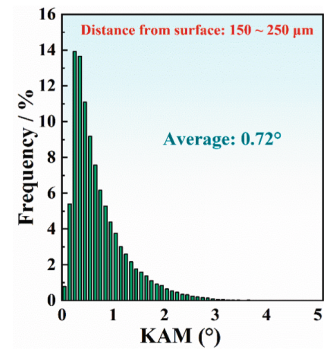
(d) Grain size within the range of 150~250 μm



(e) KAM distribution in the range of 0~50 μm



(f) KAM distribution in the range of 50~150 μm



(g) KAM distribution in the range of 150~250 μm

Fig. 16. LSP + SP specimen cross-section EBSD results.

formation rather than severe surface fragmentation [46]. Accordingly, the gradient deformation structure observed in the LSP + SP specimen can be rationalized by the sequential contribution of the two treatments. LSP first produces a relatively deep deformation affected layer containing stored defects, lattice rotation, and orientation gradients. The subsequent SP process then superimposes intense impact deformation on this previously deformed surface region, accelerating dislocation accumulation, low angle boundary development, and grain subdivision in the outermost layer. At greater depths, however, the imposed strain is insufficient to produce equally strong grain fragmentation. Therefore, the microstructure is characterized mainly by retained lattice distortion and geometrically necessary dislocation accumulation rather than by substantial further grain refinement. This interpretation is supported by the evolution with depth of grain size, KAM, GND density, and GOS: grain refinement is most pronounced near the surface, whereas elevated misorientation and stored deformation persists into the subsurface. Thus, the affected layer produced by LSP + SP forms through coupled mechanisms of progressive grain subdivision and graded dislocation accumulation. The roles of the two treatments are therefore distinct but complementary: LSP defines the depth of the layer affected by deformation through shock wave loading, while SP intensifies near surface plastic deformation and promotes stronger refinement and dislocation storage. As a result, the composite treatment generates a gradient microstructure that maximizes near surface hardening and CRS without compromising the affected depth imparted by LSP.

4.2. FW mechanisms

Fig. 17(a) shows a schematic diagram of the FW mechanisms for the four specimens. For the untreated specimen, the high oxygen content and the presence of grooves and spallation pits point to a combination of oxidative wear, abrasive wear, and delamination wear. Under gross sliding, material is readily removed, generating wear debris. If not efficiently expelled, this debris can compact and undergo work hardening under cyclic loading to form a tribolayer, typically composed of metal and metal oxides [47]. While such layers can act as protective barriers against direct contact between metal surfaces, tribolayers

formed at ambient temperature are often porous and structurally inhomogeneous, making them susceptible to crack initiation and propagation [48]. The observed delamination and spalling indicate fracture and local detachment of the tribolayer. The detached fragments then act as abrasive third bodies, which further plough the substrate, intensifying the wear process through abrasive action [49].

For the SP specimen, the dominant mechanism remains tribolayer formation followed by local cracking and spallation, but the damage sequence is modified by the surface state induced by SP. The roughened peak and valley topography improves debris retention and promotes rapid tribolayer buildup, while near surface work hardening enhances resistance to ploughing [50,51]. However, because the SP affected layer is intrinsically shallow, the tribolayer still undergoes local rupture under continued cycling. In contrast, LSP primarily improves subsurface capacity to bear load through a deeper hardened layer and a deformation zone rich in dislocations, which suppress subsurface plastic collapse and reduce the tendency for delamination over large areas. Its smoother surface, however, is less effective in retaining debris, so the protective tribolayer remains thinner and less continuous than that formed on the SP surface.

The superior FW resistance of the LSP + SP specimen arises from the synergistic coupling of a stable tribolayer with a subsurface with gradient hardening. The increased surface roughness and larger height difference between peaks and valleys favor the entrapment and compaction of wear debris, thereby promoting the formation of a thicker and more continuous tribolayer during cyclic slip. At the same time, the composite treatment produces the strongest near surface microstructural refinement and defect accumulation, together with a deeper layer affected by deformation inherited from LSP. This gradient structure enhances the capacity of the contact zone to bear load and reduces the susceptibility of the underlying substrate to plastic collapse. Consequently, the tribolayer formed on the LSP + SP surface is less prone to rupture and detachment than that on the SP or LSP treated surfaces. Once third body abrasion and tribolayer spallation are suppressed, the contributions of abrasive wear and delamination are both markedly reduced [52]. Therefore, the wear advantage of LSP + SP is not attributable solely to enhanced debris retention or solely to higher hardness

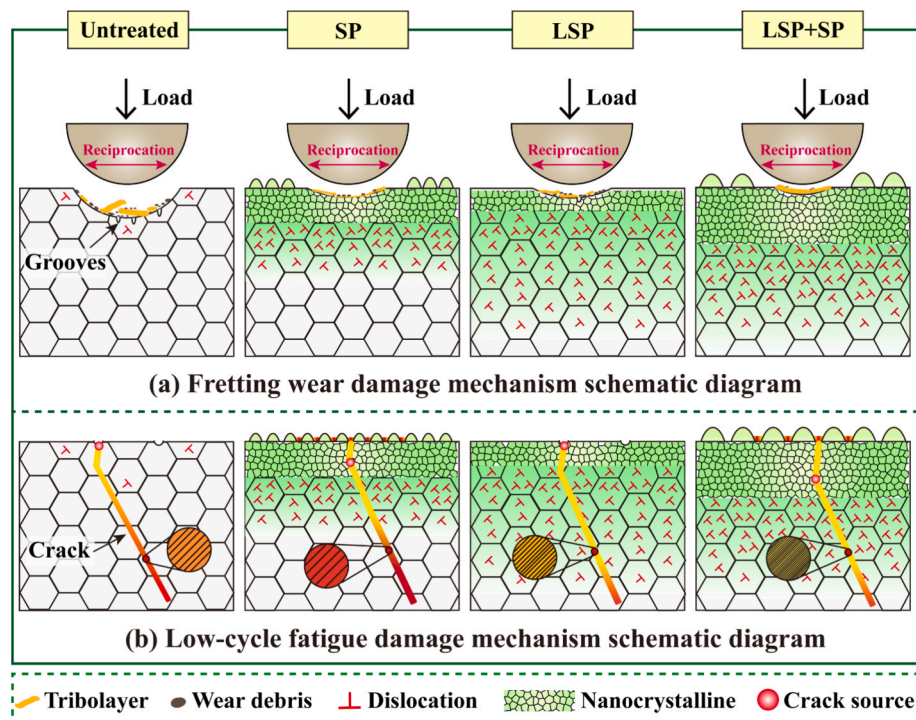


Fig. 17. Schematic representation of the FW and LCF damage mechanism for all specimens.

and CRS, but to their cooperative effect: SP facilitates the establishment of a protective tribolayer, whereas LSP preserves sufficient subsurface mechanical support to sustain this layer during repeated fretting, resulting in the lowest wear damage and the best overall tribological performance.

4.3. Fatigue fracture mechanisms

Fig. 17(b) schematically illustrates the LCF fracture mechanisms of the four specimens. For the untreated specimen, fatigue cracking is dominated by the surface condition inherited from machining. Surface scratches and other preexisting defects act as preferential stress concentrators, leading to localized cyclic slip, crack nucleation at the free surface, and subsequent crack propagation into the substrate. This mechanism is consistent with the widely reported fatigue behavior of Ti-6Al-4V, in which surface damage and local strain accumulation govern early crack formation under cyclic loading [53].

For the SP specimen, the fatigue response is improved primarily through enhanced resistance to crack initiation. SP introduces a refined and work hardened surface layer together with surface CRS, both of which reduce cyclic slip localization at the free surface and promote crack initiation in the subsurface rather than directly at the surface [54,55]. In addition, near surface grain refinement shortens the effective slip length and increases the resistance to the development of persistent slip bands, thereby suppressing early crack nucleation [11]. However, the beneficial effect of SP is fundamentally limited by the relatively shallow depth of the affected layer. Once a crack extends beyond the zone strengthened by compressive stress and hardening, the shielding effect rapidly diminishes and crack growth accelerates, which is consistent with reports that shot peening is highly effective in retarding short crack initiation and early propagation, but less effective once the crack penetrates beyond the CRS layer [56,57].

The fatigue mechanism of the LSP specimen is different. Because LSP largely preserves the original surface morphology, geometric surface defects are not eliminated and can still serve as crack initiation sites. Its principal contribution therefore lies less in suppressing crack nucleation at the surface than in reducing the subsequent driving force for crack growth [32]. Specifically, LSP introduces a deeper CRS field and a deeper subsurface with gradient deformation, which provide more persistent shielding against crack opening and cyclic plasticity ahead of the crack tip. Accordingly, although crack initiation may still occur at or near preexisting surface defects, the propagation stage is more effectively retarded than in the untreated and SP specimens. This interpretation agrees with recent studies showing that LSP in Ti-6Al-4V is particularly effective in establishing a deep deformation and stress field that improves resistance to crack propagation [58].

The LSP + SP specimen exhibits the most favorable fatigue behavior because it integrates the respective advantages of SP and LSP into a single gradient surface architecture. The SP step strengthens the first tens of micrometres most effectively by increasing hardness, dislocation density, and grain subdivision, thereby maximizing resistance to crack initiation in the near surface region [59]. The LSP step, in contrast, preserves a much deeper layer affected by deformation and CRS field, which continue to retard crack growth after initiation. In other words, SP mainly controls where a crack nucleates, whereas LSP mainly controls how fast it propagates once formed. This complementarity between the two processes is consistent with the present EBSD results: the LSP + SP condition exhibits the strongest near surface refinement and the highest local strain storage, while still retaining deformation signatures at greater subsurface depths. Consequently, the number of effective crack initiation sites decreases, the crack source shifts away from the free surface, and the initial crack growth rate is further reduced. The fatigue advantage of LSP + SP therefore arises not from a simple superposition of two treatments, but from the simultaneous optimization of near surface resistance to crack initiation and subsurface resistance to crack propagation.

5. Conclusions

This study systematically investigated the effects of LSP, SP, and their composite treatment (LSP + SP) on the surface integrity, FW resistance, and LCF performance of Ti-6Al-4V alloy. The key findings are summarized as follows:

- Sequential LSP + SP produced a gradient microstructure in Ti-6Al-4V. Grain refinement and lattice distortion were strongest in the outermost 0 to 50 μm region, while elevated KAM, GND and GOS extended into the subsurface, indicating graded deformation and strain storage.
- The combined LSP + SP treatment integrated the advantages of both processes, achieving the highest surface hardness (524.1 HV) and maximum CRS (-746.3 MPa) while maintaining a CRS affected depth comparable to that of LSP alone. This indicates that SP primarily enhanced the near-surface strengthening effect, whereas LSP preserved the strengthening depth.
- LSP + SP reduced the FW volume by 31.3% relative to the untreated specimen. This behavior is associated with the coupling of a more stable tribolayer and a subsurface with gradient hardening, which reduced tribolayer spallation, abrasive wear, and delamination.
- LSP + SP extended LCF life by 150.1%. Crack initiation shifted to a deeper subsurface region and fatigue striation spacing decreased, indicating that the combined treatment improved resistance to both crack initiation and early crack propagation.

Future work should focus on the stability and applicability of the gradient surface structure induced by LSP + SP. In particular, the evolution of the refined and deformed layer and CRS field under prolonged cyclic contact and thermomechanical loading needs to be clarified.

CRedit authorship contribution statement

Zhiguo Wang: Writing – original draft, Data curation, Investigation, Methodology. **Jianxin Liu:** Investigation, Methodology. **Ke Li:** Data curation. **Xiuyang Fang:** Resources. **Magd Abdel Wahab:** Writing – review & editing, Supervision, Investigation. **Zhenbing Cai:** Conceptualization, Funding acquisition, Project administration, Supervision.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Data availability

Data will be made available on request.

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