

Article

Effect of Surface Layer Removal After Ultrasonic Surface Rolling Processing on the Tension–Tension Fatigue Performance of AZ31B Magnesium Alloy

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Abstract

This paper investigates the influence of surface ultrasonic rolling treatment on the fatigue performance of Mg-3Al-1Zn extruded alloy and systematically analyzes the evolution laws of fatigue life and mechanical properties with the thickness of the surface removed layer. The results show that after ultrasonic rolling treatment, the fatigue life of the alloy at a stress amplitude of 240 MPa changes significantly and reaches a peak at a specific removal thickness: when the 80 μm surface layer is removed, the fatigue life reaches 7.79×10^6 cycles, which is much higher than that of the untreated sample (3.87×10^4) and the sample only subjected to ultrasonic surface rolling processing (1.8×10^4). With the increase in the removal thickness, the fatigue life shows a trend of first increasing and then decreasing, and a second increase occurs within the range of 400–500 μm . Microstructure analysis indicates that at a depth of 80 μm from the surface, the strength is enhanced due to grain refinement and the peak hardness, thereby inhibiting the initiation of fatigue cracks, while within the depth range of 400–500 μm , there exist high-density dislocations and deformation layers, which also effectively hinder crack propagation. This study reveals the key role of surface state and subsurface microstructure in the fatigue behavior of magnesium alloys, providing a theoretical basis for improving the fatigue performance of magnesium alloys through surface modification.

Keywords: AZ31B magnesium alloy; ultrasonic surface rolling processing; surface removal; fatigue performance



Academic Editor: Dariusz Rozumek

Received: 3 April 2026

Revised: 9 May 2026

Accepted: 11 May 2026

Published: 14 May 2026

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1. Introduction

Magnesium alloys, characterized by their low density, high specific strength, and excellent castability, are increasingly being employed in automotive, aerospace, and electronics industries [1–3]. However, their widespread application is often constrained by inherent limitations associated with the hexagonal close-packed (HCP) crystal structure, such as moderate ductility, formability, and particularly, fatigue resistance [4,5]. Fatigue failure, in particular, poses a significant challenge for the reliability and service life of magnesium alloy components, necessitating effective strategies for performance enhancement.

Ultrasonic surface rolling processing (USRP) has emerged as a promising surface modification technique for improving material performance. The process converts high-frequency electrical signals into mechanical vibrations, which are transmitted through a rolling tool that simultaneously applies static force to the workpiece surface. This combined

action effectively reduces surface roughness by “smoothing peaks and filling valleys” [6]. More importantly, it introduces severe plastic deformation (SPD) within a surface layer, leading to grain refinement, work hardening, and the generation of beneficial residual compressive stresses [7–9]. Collectively, these microstructural and mechanical alterations can significantly influence the fatigue behavior of materials [10].

Recent studies have demonstrated the efficacy of USRP in modifying surface properties of various materials. Research on steels has shown that USRP can induce grain refinement, enhance surface hardness, improve wear and corrosion resistance, and shift crack initiation sites from the surface to subsurface regions [11–13]. In stainless steels, USRP has been shown to create gradient microstructures with significantly increased residual stress and dislocation density [14]. For magnesium alloys, studies confirm that USRP effectively reduces surface roughness and produces a gradient nanostructured surface layer, often involving mechanisms like twinning and texture evolution [8,15,16]. These modifications are generally associated with improved surface properties, including microhardness and corrosion resistance [6,17–22].

The variations in fatigue properties in zero-tension load and tension–compression load were related to deformation mechanisms under different load conditions. The fatigue deformation mechanism in Mg-6Zn-1Mn alloy under zero-tension load was mainly dislocation slip, while under tension–compression load, the activation of twinning in compression half-cycles resulted in deteriorative fatigue strength [23]. The research on the fatigue properties of AZ31 magnesium alloy includes the following aspects. Through fatigue tests conducted on samples with different grain sizes, it was concluded that twinning is the key factor influencing the fatigue life of the AZ31B alloy [24]. At the same time, the research also found that this alloy exhibits cyclic stability at lower strain amplitudes and shows cyclic hardening characteristics at higher strain amplitudes [25]. After the surface mechanical rolling treatment of AZ31B magnesium alloy, the fatigue limit reached 140 MPa, an increase of 55.56%. When combined with pre-tension deformation and surface mechanical rolling treatment, the fatigue limit reached 150 MPa, an increase of 66.67% [26].

Despite these advances, research specifically targeting the tension–tension fatigue performance of USRP-treated magnesium alloys remains limited, and the underlying mechanisms are not fully understood. While some studies report significant improvements in fatigue life for materials like aluminum alloys and steels after USRP, others note that processing parameters must be optimized, as excessive treatment can be detrimental [27–30]. The complex interplay between the introduced gradient microstructure—which may include heavily deformed layers, recrystallized zones, and plastically deformed regions with varying residual stress states—and the resultant fatigue behavior requires systematic investigation. The specific role of different subsurface layers in either promoting or inhibiting fatigue crack initiation and propagation is a critical, yet underexplored, aspect.

Therefore, this study aims to elucidate the mechanisms governing the fatigue performance of AZ31B magnesium alloy following USRP treatment. Through a combination of fatigue testing, systematic surface layer removal, and detailed microstructural characterization, we investigate how the distinct gradient layers induced by USRP—each with unique grain morphology and residual stress characteristics—govern the fatigue life. The findings are expected to provide a theoretical foundation and technical insights for optimizing USRP parameters to enhance the fatigue resistance of magnesium alloys and other metallic materials.

This work clearly distinguishes its novel contributions from previous studies in methodology, microscopic mechanism and quantitative results. Methodologically, unlike most conventional studies that only evaluate the overall fatigue performance of USRP-modified magnesium alloys, an innovative layer-by-layer surface removal strategy is

adopted herein to decouple the independent influence of gradient microstructures at different depth layers on fatigue behavior. Mechanistically, beyond the traditional interpretation that fatigue improvement is merely attributed to surface strengthening and compressive residual stress, the competitive and synergistic effect between microstructural stability and residual stress distribution is further revealed. Meanwhile, the respective roles of the defect-rich surface layer, recrystallized subsurface layer and dislocation-concentrated deformed layer in governing fatigue crack initiation are clarified. In terms of quantitative findings, the quantitative correlation among surface removal depth, microstructure characteristic, hardness evolution and fatigue life is established. The optimal removal depth corresponding to the maximum fatigue life is determined, and the degree of fatigue life enhancement is quantitatively evaluated, which provides reliable theoretical and quantitative reference for parameter optimization and performance regulation of USRP-treated AZ31B magnesium alloy.

2. Experiment Methods

The AZ31B magnesium alloy used in this study was fabricated via hot extrusion of a continuously cast billet with an initial diameter of 180 mm down to a final diameter of 60 mm. The extrusion was performed at 350 °C and immediately followed by water quenching. The chemical composition of the alloy is listed in Table 1.

Table 1. Chemical composition of AZ31B magnesium alloy extruded rods.

Element	Mg	Al	Zn	Mn
Content, wt. / %	96.44	2.43	0.75	0.38

Cylindrical specimens were wire-cut from the as-received rod and subsequently machined to the final dimensions illustrated in Figure 1a using a lathe. The ultrasonic rolling system employed in this study is depicted in Figure 1b. It primarily consists of an ultrasonic generator, a cooling and lubrication system, and a rolling execution mechanism. The ultrasonic generator and the cooling–lubrication unit are integrated into a control cabinet. The rolling execution mechanism comprises a transducer, an amplitude transformer, a rolling head, and an air cylinder. The ultrasonic generator produces an electrical signal, which is converted into mechanical vibration via the transducer. This vibration is amplified by the amplitude transformer and transmitted to the rolling head, which operates at a frequency of up to 30 kHz. The rolling head is equipped with an 8 mm diameter rotatable tungsten carbide ball that directly contacts and rolls over the specimen surface during processing. The static normal force required for the process is supplied by a 63 mm diameter air cylinder. The ultrasonic rolling parameters adopted in this work are as follows: static force 150 N, amplitude 10 µm, spindle speed 140 rpm, feed rate 0.1 mm/min, and two rolling passes.

The tension–tension loading mode ($R = 0.1$) was selected because it better represents the actual service conditions of many structural components made from magnesium alloys, which often experience tensile-dominated cyclic loads [25,26]. Fatigue testing of the as-received material was conducted in accordance with the GB/T 3075-2021 standard [31] under multiple stress amplitudes, with three specimens tested at each level. Axial fatigue experiments were performed using an LF5105 electro-hydraulic servo universal testing machine (Lishi (Shanghai) Scientific Instruments Co., Ltd., Shanghai, China) under tension–tension loading (stress ratio $R = 0.1$), with a sinusoidal waveform at a frequency of 20 Hz. Tests were terminated upon specimen fracture or when 10^7 cycles were reached (run-out).

Uniaxial tensile tests were performed at room temperature using an LD23.104 electronic universal testing machine (Lishi (Shanghai) Scientific Instruments Co., Ltd., Shanghai, China) at a speed of 1.0 mm/min.

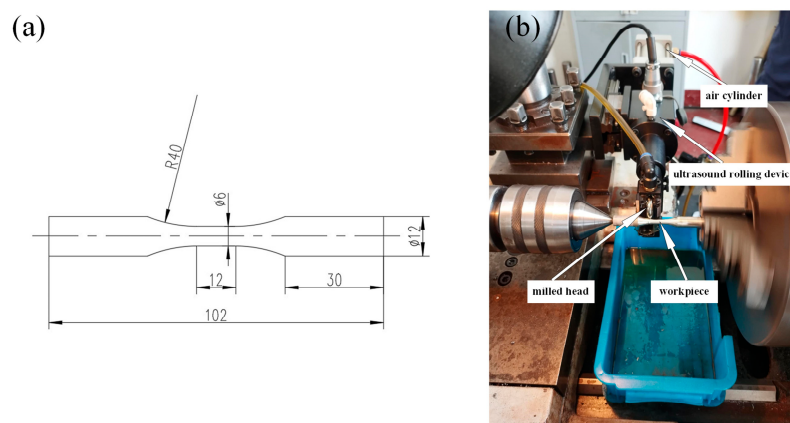


Figure 1. (a) Sample dimensions. (b) Ultrasonic surface rolling processing diagram.

The microhardness profile from the treated surface to the substrate was measured using an HV-1000STA micro-Vickers hardness tester (Ningbo Economic and Technological Development Zone Kainuo Instrument Co., Ltd., Ningbo, China). A load of 10 g and a dwell time of 15 s were applied during the tests. Measurements were taken at intervals of 15 μm , with three datasets collected for each sample. The average value of these measurements was used for subsequent analysis.

Both the as-received material and the USRP treated specimens were prepared for microstructural characterization through a standardized metallographic procedure. The samples were sequentially ground using 800, 2000, 3000, and 7000 grit sandpaper, followed by polishing with a diamond spray polishing agent on a woolen cloth. Subsequently, electrolytic polishing was applied to minimize surface deformation. The processed surfaces were then examined using electron backscatter diffraction (EBSD) for microstructural and crystallographic analysis.

The USRP-treated specimen was subjected to a surface removal process by immersing the specimen in a uniformly mixed solution of 500 mL deionized water and 2.5 g oxalic acid. The specimen was suspended in the beaker and rotated continuously at a constant speed during immersion. Subsequently, it was mounted on a lathe and rotated at 300 rpm for sequential grinding with abrasive paper and polishing until the predetermined dimensions and desired surface roughness were achieved. The surface removal procedure was designed to minimize unintended effects. Chemical etching using a mild oxalic acid solution is a stress-free removal method that does not alter the underlying residual stress state or microstructure. The subsequent light mechanical polishing was performed with fine abrasive paper (up to 7000 grit) and diamond paste under low pressure, which affects only the outermost few micrometers and has been shown not to introduce significant residual stress or microstructural changes [32]. After each removal depth, surface roughness was measured and maintained at $R_a = 0.2 \pm 0.03 \mu\text{m}$ to ensure consistency across all specimens.

3. Result and Discussion

3.1. Initial Materials

The cross-sectional microstructure of the as-received material was characterized by EBSD. Figure 2a presents the inverse pole figure (IPF) map, in which a distinct extrusion zone can be observed. This zone resulted from plastic flow and the corresponding adaptive transformation of grains during the extrusion process. The microstructure consists of grains with non-uniform sizes distributed relatively uniformly throughout the cross-section, yielding an average grain size of approximately 14.79 μm . Figure 2c shows the distribution of grain sizes, and the average grain size of the recrystallized fine grains is about 10 μm .

Figure 2b shows the corresponding pole figure (PF). Variations in color represent differences in texture intensity. A strong texture is evident along the rolling direction (RD), with a maximum texture intensity of 11.322.

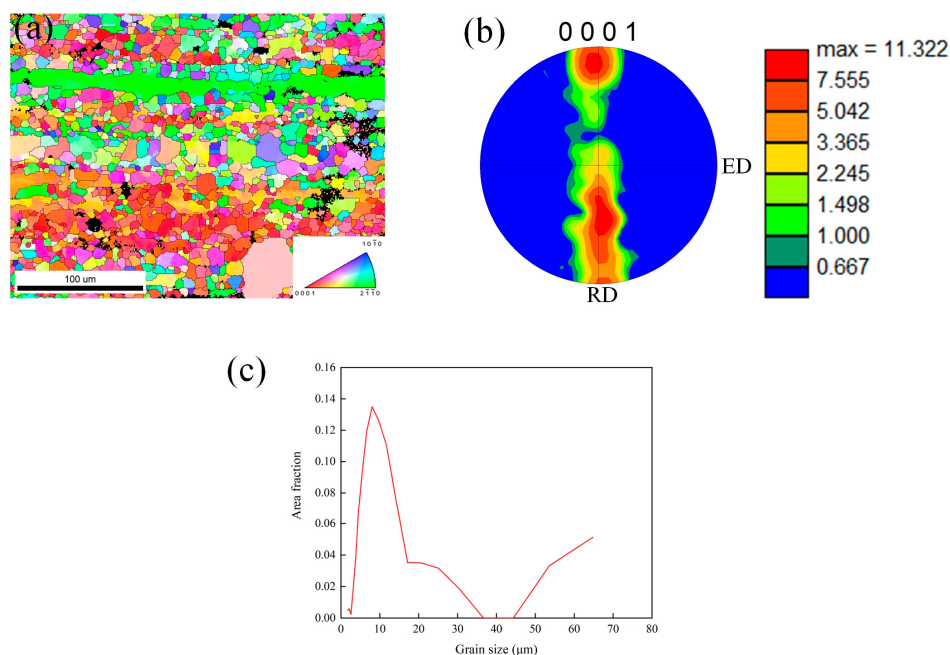


Figure 2. Microstructure of the extruded AZ31B alloy: (a) IPF diagram. (b) PF diagram. (c) Grain size distribution diagram.

The fatigue behavior of the as-received material was evaluated under tension–tension loading, and the resulting S–N curve is presented in Figure 3a. Using 0.8 times the yield strength (approximately 180 MPa) as the baseline stress, fatigue tests were performed with a 10 MPa stress increment. Testing was conducted at eight stress amplitudes (250, 240, 230, 220, 210, 200, 190, and 180 MPa), with three specimens tested at each level. The fatigue limit of 210 MPa was identified as the highest stress amplitude at which specimens consistently survived beyond 10^7 cycles. This determination is supported by the fact that all three specimens tested at 200 MPa reached run-out, while specimens tested at 210 MPa exhibited both run-out and failure, and specimens tested at 220 MPa consistently failed before 10^7 cycles. This staircase-like behavior provides statistical confidence in the reported fatigue limit. The line shown in Figure 3 is the linearly fitted trend lines, whose purpose is to visually demonstrate the overall negative correlation between the stress amplitude and the fatigue life. The S–N relationship exhibits two distinct regimes: low-cycle fatigue at stress amplitudes above 240 MPa and high-cycle fatigue below 240 MPa. A linear fit was applied to the high-cycle fatigue regime (stress amplitudes ≤ 240 MPa) to determine the fatigue strength, defined as the stress amplitude corresponding to 10^7 cycles. The fatigue strength is determined to be 210 MPa. In the present study, the fatigue strength obtained from tensile fatigue tests was higher than that from compressive fatigue tests. This material exhibits a strong crystallographic texture and mainly deforms through basal slip, while the occurrence of twinning is relatively limited. These factors collectively contribute to the enhanced fatigue strength. Quasi-static tensile tests were also performed, and the corresponding stress–strain curve is shown in Figure 3b. The measured yield strength, tensile strength, and elongation at fracture are 233 MPa, 289 MPa, and 19.0%, respectively. The relatively high fatigue strength and yield strength can be attributed to the fine-grained microstructure characterized in the material [33].

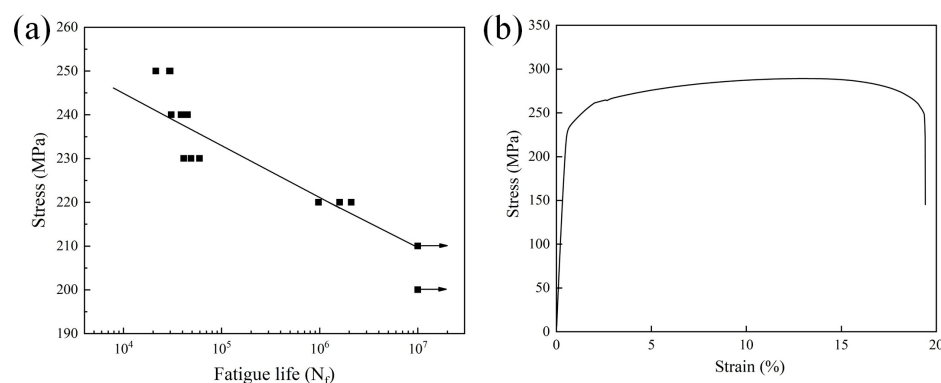


Figure 3. (a) Fatigue life curve and (b) stress–strain curve of extruded AZ31B alloy.

The relatively small difference between yield strength (233 MPa) and fatigue strength (210 MPa) can be attributed to the strong crystallographic texture of the extruded AZ31B alloy (Figure 2b) and its deformation characteristics. The material primarily deforms through basal slip due to the strong basal texture, while twinning activity is relatively limited under tension–tension loading. These factors collectively contribute to the enhanced fatigue strength, which is consistent with previous reports on textured magnesium alloys [24,32].

3.2. Effect of USRP on Microstructure Evolution and Mechanical Property

After USRP, the samples were subjected to fatigue testing at a stress amplitude of 240 MPa, and the results were compared with those of the initial material, as shown in Figure 4. The experimental results indicate that under the similar surface roughness condition of $R_a = 0.2 \mu\text{m}$, the fatigue life of the USRP-treated samples showed a slight decrease to 1.80×10^4 cycles at a stress amplitude of 240 MPa.

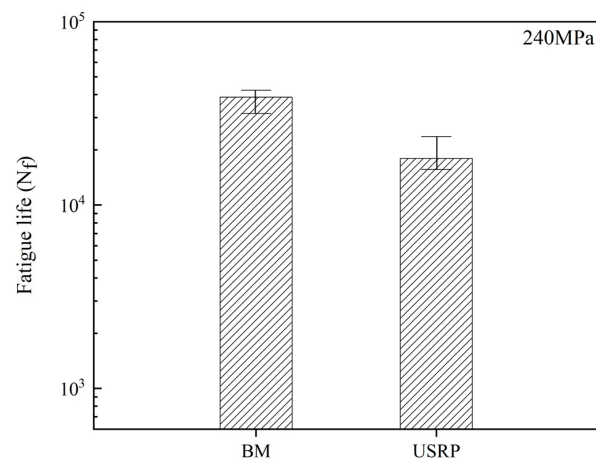


Figure 4. Fatigue life of the matrix and USRP samples at 240 MPa.

The USRP parameter scheme adopted in this study was formulated based on the accumulated experimental experience of our research group, which was intentionally set to obtain typical gradient microstructure characteristics along the depth direction of the material. Admittedly, the selected USRP parameters result in a certain degree of reduction in fatigue life of the specimen. However, the research orientation of this work is not to screen out the optimal process parameters for improving fatigue resistance. The core research goal is to quantitatively reveal the influence law of gradient microstructures at different depth layers on fatigue damage behavior by adopting the layer-by-layer surface stripping characterization method. In this context, the current USRP parameter selection is in line with the design idea of controlled variables in this experiment, which can effectively

guarantee the pertinence and rationality of exploring the correlation between gradient microstructure and fatigue performance.

This marginal decline in fatigue life suggests that while USRP generally enhances surface integrity, the specific parameters used here may have introduced a near-surface region with excessive defect density, which counteracts the potential benefits of grain refinement and work hardening. Such a phenomenon underscores the sensitivity of fatigue performance to the balance between surface strengthening and defect introduction in USRP-treated Mg alloys. Similarly, studies have shown that the USRP treatment introduces a high density of dislocation defects in the gradient structure layer on the surface of the Zr705 alloy, making it more susceptible to corrosion than untreated samples. In highly corrosive media, the fatigue life of the USRP samples is actually lower than that of the untreated samples [34]. Chen et al. found that the fatigue life of the samples treated solely by USRP was 3.36×10^6 cycles, while the fatigue life of the samples treated with the USRP combined with DLC coating was increased to 6.55×10^6 cycles [35]. This indicates that although the USRP treatment without coating optimization has some effect, its surface damage mechanism still limits the potential for increasing the fatigue life.

To systematically investigate the influence of USRP on the microstructure of AZ31B magnesium alloy, EBSD analysis was performed on the cross-sections of treated samples, with data processed using OIM Analysis™ 7.0 software. The results are presented in Figure 5. It should be noted that the formation of the gradient microstructure is governed by multiple USRP parameters, including static load, amplitude, tool geometry, feed rate, and thermophysical conditions such as heat dissipation. Under the fixed processing conditions employed in this study (static force 150 N, amplitude 10 μm , two passes), the resulting gradient structure shown in Figure 5 was consistently observed across multiple samples, confirming its repeatability.

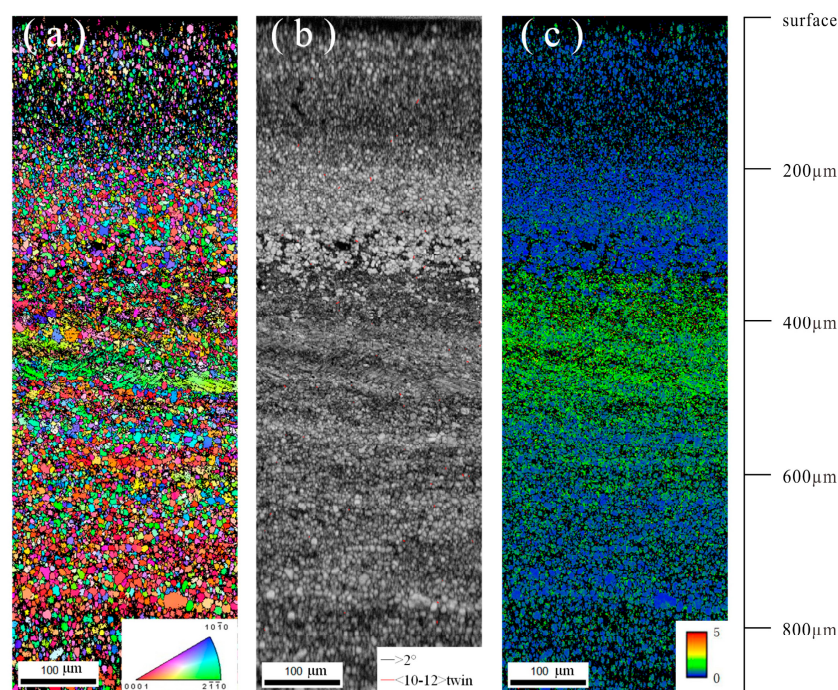


Figure 5. Microstructure of the cross-section of AZ31B magnesium alloy after USRP: (a) IPF diagram; (b) IQ diagram; (c) KAM diagram.

Figure 5a shows the inverse pole figure (IPF) map of the USRP-treated cross-section. Significant variations in microstructure are observed with increasing distance from the rolled surface. After ultrasonic rolling treatment, the samples exhibit significant grain

refinement due to the formation of twinned crystals and dislocations [36,37]. In the surface region, severe plastic deformation and lattice distortion result in a low EBSD indexing rate. As the distance from the rolled surface increases, fine grains formed through complete recrystallization are evident. This is basically consistent with the previous changes in the microstructure of the material after ultrasonic rolling [8,38]. Within the range of 80–180 μm from the surface, a mixture of unindexed areas and finer recrystallized grains is present, indicating partial recrystallization in this subsurface region induced by surface rolling treatment.

In the region 180–360 μm from the surface, a relatively fine and homogeneous grain distribution is observed. Image quality (IQ) maps reveal reduced lattice distortion in this zone, while kernel average misorientation (KAM) maps indicate lower stored plastic strain energy, suggesting that recrystallization is fully accomplished in this layer. Between 360 μm and 540 μm from the surface, the grains exhibit a distinctly deformed morphology. KAM analysis confirms a higher level of stored plastic strain energy in this region, indicative of substantial plastic deformation introduced by the rolling process. Beyond 540 μm from the surface, the stored plastic strain energy gradually decreases with increasing depth, as reflected in the KAM results.

The observed gradient microstructure is consistent with the typical response of Mg alloys to severe plastic deformation processes. The near-surface region with low EBSD indexing rate likely contains nanocrystalline or amorphous-like structures, which are mechanically unstable under cyclic loading. The intermediate recrystallized zone (180–360 μm) represents a structurally stable region with enhanced grain boundary strengthening. The deeper deformed zone (360–540 μm) retains dislocations, which contribute to strain hardening and residual stress retention. This layered microstructure creates a complex mechanical response under fatigue conditions.

Figure 6 illustrates the variation in hardness as a function of distance from the rolled surface. As shown in the figure, the hardness initially increases and then decreases with increasing distance from the surface. The increase in hardness can be attributed to grain refinement, twinning and dislocations [17]. The peak hardness, reaching a maximum value of approximately 80.6, occurs at a distance of about 115 μm from the rolled surface, which represents an increase of about 58% compared to the base material hardness of 51.

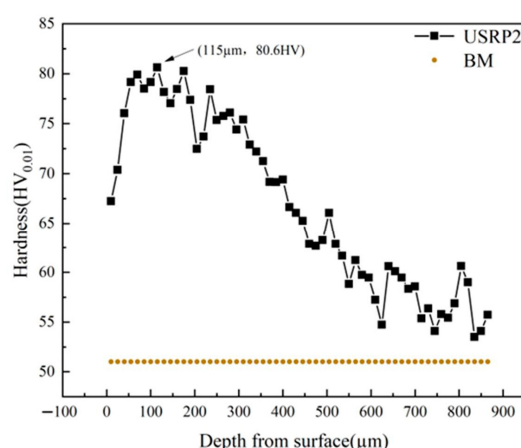


Figure 6. Microhardness of AZ31B samples after USRP.

In contrast, the region closest to the rolled surface exhibits a hardness of only 67, significantly lower than the peak value. This can be attributed to the high density of crystalline defects introduced by the rolling process in the near-surface region. The presence of these substantial defects also explains why the fatigue performance of the rolled material at a stress amplitude of 240 MPa is inferior to that of the base material.

The hardness profile further corroborates the microstructural observations. The sub-surface peak hardness at $\sim 100\ \mu\text{m}$ corresponds to the partially recrystallized zone where work hardening and grain refinement synergistically enhance strength. The lower surface hardness suggests that excessive plastic deformation may have caused micro-void formation or surface damage, reducing load-bearing capacity. This hardness distribution implies that the beneficial effects of USRP are concentrated beneath the immediate surface, highlighting the importance of subsurface microstructure control for fatigue enhancement.

The total thickness of the microstructurally modified layer, as reflected by the hardness profile and EBSD analysis, extends to approximately 1 mm. This depth is greater than the typical 100–300 μm range reported in some USRP studies on other materials, which can be attributed to the relatively low yield strength of the AZ31B magnesium alloy and the specific processing parameters used (high static load and amplitude).

The slight decrease in fatigue life of the as-USRP sample (Figure 4) is attributed to the presence of a severely deformed, defect-rich surface layer characterized by a low EBSD indexing rate and reduced near-surface hardness (Figures 5 and 6). This phenomenon is consistent with previous reports indicating that non-optimized USRP conditions can lead to surface damage and diminished fatigue performance [29,30]. This observation underpins the rationale for the subsequent surface removal experiments, which aim to identify the optimal subsurface microstructure for fatigue enhancement.

During ultrasonic surface rolling processing (USRP), the reduction in hardness of the near-surface region is not only attributed to high-density lattice defects and residual stress induced by severe plastic deformation. Alternative mechanisms also play important roles, including transient thermal effects, microstructure recovery and surface micro-damage. The high-frequency cyclic impact of USRP produces instantaneous local temperature rise, which facilitates partial dislocation recovery and annihilation in the outermost layer, thereby weakening the work-hardening effect. Meanwhile, repeated strong impact inevitably introduces slight surface micro-damage and lattice distortion. Consequently, the decreased surface hardness is a comprehensive result of defect accumulation, thermal-assisted recovery and surface micro-damage under the coupling effect of USRP.

Figure 7 shows the average grain size distribution along the depth of the AZ31 alloy after ultrasonic rolling, exhibiting a fluctuating gradient trend. Severe plastic deformation refines the surface grains to 4.536 μm (0–20 μm). The grain size then slightly increases to 5.782 μm (20–100 μm), decreases to 4.366 μm (100–180 μm), and rises to 5.668 μm (180–360 μm) due to recrystallization and slight growth. The maximum grain size (11.320 μm) appears at 360–550 μm , while values decrease to 7.572 μm at 550–800 μm , approaching the matrix. This gradient evolution provides microstructural evidence for the enhanced fatigue performance of the treated alloy.

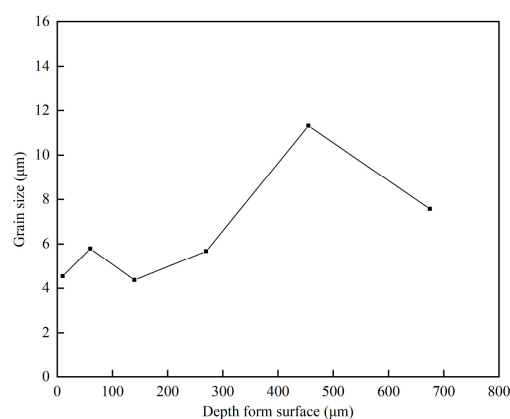


Figure 7. The grain size varies with distance.

Figure 8 displays the statistical distributions of grain boundary misorientation angles at six typical depth layers from the surface to the matrix of AZ31 magnesium alloy after ultrasonic rolling treatment. Generally, misorientation angles of 2–15° are defined as low-angle grain boundaries (LABs) related to dislocation accumulation and deformed substructures, while angles above 15° correspond to high-angle grain boundaries (HABs) associated with recrystallized boundaries and twin boundaries. The fractions of LABs and HABs at different depths are quantitatively marked to reveal the gradient evolution of grain boundary characteristics.

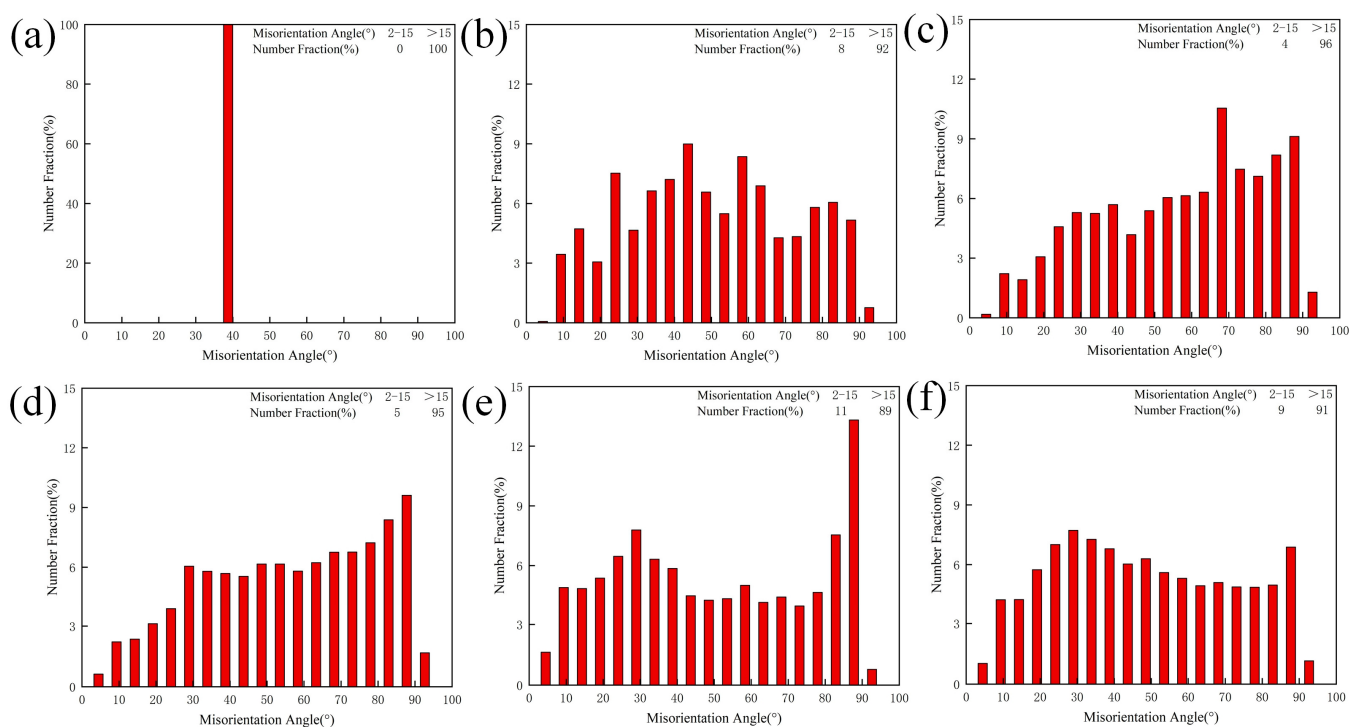


Figure 8. Grain boundary misorientation angle distribution of AZ31 magnesium alloy at different depth layers after ultrasonic rolling treatment. (a) 0–20 μm; (b) 20–100 μm; (c) 100–180 μm; (d) 180–360 μm; (e) 360–550 μm; (f) 550–800 μm.

In the 0–20 μm surface layer (Figure 8a), misorientation angles are highly concentrated around 40°, with HABs occupying 100%. However, the low EBSD indexing rate in this layer weakens the statistical representativeness. For the 20–100 μm subsurface layer (Figure 8b), LABs and HABs account for 8% and 92%, respectively. The dispersed misorientation distribution indicates retained deformed substructures and ongoing dynamic recrystallization, which transforms partial LABs into HABs and generates a small number of compression twins. In the 100–180 μm region (Figure 8c), the LAB fraction decreases to 4% and the HAB fraction increases to 96%, with the main misorientation peak shifting to 60–80°, suggesting further recrystallization and gradual elimination of deformed sub-boundaries. The 180–360 μm region (Figure 8d) maintains an extremely low LAB fraction of 5%, accompanied by a dominant peak at 80–90°, which is a typical feature of fully recrystallized microstructure with released deformation stored energy.

The 360–550 μm region (Figure 8e) presents an obvious bimodal misorientation distribution, with LABs up to 11% and HABs of 89%. A prominent peak near 90° confirms the retention of deformed sub-boundaries and twin boundaries, which is consistent with the high KAM value and residual stress, and serves as the microstructural origin of improved fatigue life in this layer. In the 550–800 μm region (Figure 8f), the misorientation distribu-

tion becomes uniform with 9% LABs and 91% HABs, indicating that the microstructure gradually recovers to the initial matrix state.

Overall, ultrasonic rolling induces an obvious gradient evolution of grain boundary structure. The LAB fraction decreases first, rises to the maximum at 360–550 μm , and then gradually returns to the matrix level, while the HAB fraction shows an opposite trend. The distribution of twin boundaries also presents a gradient feature along the depth direction, matching the variation in KAM value and residual stress. Combined with the gradient microstructure and stress field, the evolved grain boundary structure dominated by high-angle twin boundaries and residual stress can effectively restrain crack initiation, thereby dominating the gradient improvement of fatigue performance for ultrasonic rolled AZ31 magnesium alloy.

It should be emphasized that the above microstructural and mechanical responses are highly sensitive to USRP parameters. Improper parameter selection (e.g., excessive static load, amplitude, or passes) can cause several detrimental consequences. First, severe near-surface damage may lead to a fatigue life even lower than that of the untreated material (as seen in the as-USRP sample in Figure 4). Second, the peak hardness can shift to subsurface layers while the outermost surface softens (Figure 6). Third, the beneficial residual compressive stress field may become mismatched, potentially accelerating crack propagation. Fourth, the desired gradient microstructure (deformed layer—recrystallized zone—high-dislocation layer) can be disrupted. As a result, improper USRP parameters may produce a poor as-treated fatigue performance, while selective surface removal (e.g., 80 μm or 400–500 μm in this study) reveals significant improvement. Hence, when applying USRP to a specific alloy, parameter selection should be carefully considered; nevertheless, the primary goal of this work is not to determine optimal parameters but to elucidate how the depth-dependent gradient microstructure governs fatigue behavior.

3.3. Effect of Surface Layer Removal on Microstructure Evolution and Mechanical Property

The USRP-treated specimens were subjected to layer removal at different depths, followed by grinding and polishing to maintain a consistent surface roughness. Fatigue tests were then performed at a stress amplitude of 240 MPa using an electro-hydraulic servo universal testing machine, with three samples tested repeatedly for each removal depth and error bars included to evaluate the fatigue life of the base material, the as-rolled specimen, and specimens with different material removal depths; the corresponding results are summarized in Figure 9. The error bars (standard deviations) under all the removal depth conditions are also included in Figure 9. The number of repeated experiments for each condition is indicated in the figure legends. The repeatability analysis shows that the standard deviation ranges from 8% to 15% of the average fatigue life, which is a common situation in fatigue tests of magnesium alloys. The fatigue strength of the untreated sample was 210 MPa. After ultrasonic rolling processing to remove 80 micrometers, the fatigue strength became 230 MPa. Considering all these factors, it was decided to conduct the test at 240 MPa which lies above the yield strength (233 MPa) but within the high-cycle fatigue regime for this material ($N > 10^4$ cycles). This would better reflect the differences in fatigue life at different removal depths. And given the large number of conditions investigated (as-USRP plus seven removal depths, each requiring multiple specimens), conducting full S-N curves for each condition would require an impractical number of tests. The single stress amplitude approach allows systematic comparison across conditions while maintaining statistical validity through replication.

The fatigue life as a function of removal depth (Figure 9) exhibits a distinct non-monotonic pattern with two peaks. This behavior can be systematically explained by the microstructural gradient characterized in Section 3.2. At 80 μm removal, fatigue life

peaks at 7.79×10^6 cycles, corresponding to removal of the defect-rich surface layer and exposure of the recrystallized zone with high mechanical stability. At 200–300 μm removal, fatigue life declines as the recrystallized zone—lacking sufficient compressive stress—is exposed. At 400–500 μm removal, fatigue life recovers to $\sim 7.2 \times 10^6$ cycles, coinciding with exposure of the plastically deformed layer that retains beneficial residual compressive stress. At 600 μm removal, fatigue life decreases again as this deformed layer is fully removed.

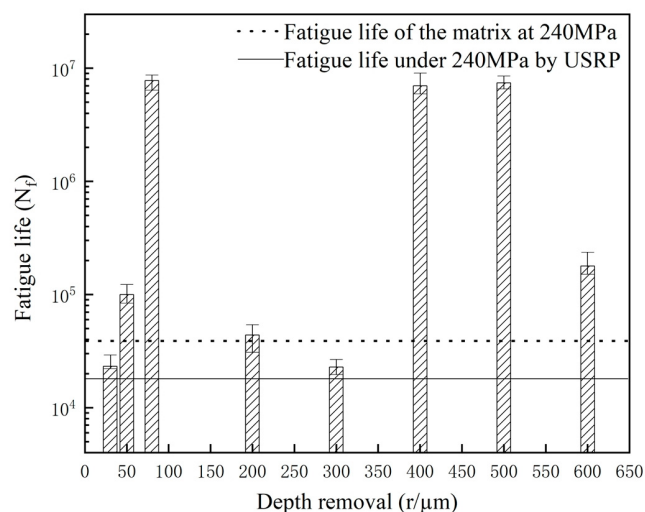


Figure 9. Comparison of fatigue life at a stress amplitude of 240 MPa with respect to different removal depths.

The double-peak fatigue characteristic is only observed at the specific stress amplitude of 240 MPa in the present work, rather than occurring universally across all tested stress levels. Although this phenomenon is limited to this particular loading condition, it is regarded as a typical structural response feature and an important research breakthrough point. By analyzing the formation mechanism of the double-peak behavior at 240 MPa, the competitive relationship between surface and internal crack initiation, the evolution of dislocation configuration, and microstructural modification are further clarified. On this basis, the intrinsic mechanism responsible for the effective improvement of fatigue life is reasonably revealed, which provides a clear physical insight into the fatigue enhancement effect of the material.

Gao et al. improved the fatigue strength of AZ31B magnesium alloy through pre-stretching deformation and surface mechanical rolling treatment. They found that the fatigue strength after pre-stretching was 115 MPa, after surface mechanical rolling it was 140 MPa, and the fatigue strength after combining pre-stretching and surface mechanical rolling was 150 MPa [26]. Zhao et al. processed AZ31B magnesium alloy with high-frequency impacting and rolling, increasing its fatigue strength from 140 MPa to 180 MPa [39]. In this study, by removing the surface layer, the fatigue strength of AZ31B magnesium alloy was increased to 230 MPa, and the fatigue life at a stress of 240 MPa could reach 7.79×10^6 .

When the surface layer is removed to a depth of 80 μm , the peak fatigue life is achieved. This corresponds to the removal of the near-surface region characterized by severe lattice distortion and a low EBSD indexing rate (Figure 4). Eliminating this heavily deformed and defect-rich surface layer effectively removes the primary sites for early crack initiation. Furthermore, the exposed subsurface material at this depth resides within the region of complete recrystallization (180–360 μm from the original surface), which features a fine, homogeneous grain structure with minimal lattice distortion and low stored strain energy (as indicated by IQ and KAM maps). This recrystallized microstructure

possesses high mechanical stability and resistance to crack propagation, leading to optimal fatigue performance.

The significant reduction in fatigue life at removal depths of 200 μm and 300 μm can be primarily attributed to the exposure of the fully recrystallized microstructure (originally at 180–360 μm beneath the surface). Although this region exhibits fine grains and low strain energy, as indicated by KAM analysis, its low level of residual compressive stress is less effective in suppressing crack initiation during cyclic deformation. Consequently, despite the absence of severe deformation defects, the fatigue performance deteriorates due to the diminished crack initiation resistance provided by the low-stress state.

Conversely, the remarkable recovery and enhancement of fatigue life at removal depths of 400 μm and 500 μm are linked to the exposure of the plastically deformed layer (originally at 360–540 μm from the surface). This region exhibits a deformed morphology with a higher level of stored plastic strain energy (Figure 4). This condition corresponds to the presence of beneficial high residual compressive stresses. Previous studies have established a strong correlation between KAM and residual stress in plastically deformed metals [40]. And based on published studies on USRP-treated AZ31B magnesium alloys, the residual stress profile typically follows a characteristic distribution: Geng et al. [8] reported that after USRP with similar parameters (static force 150 N, amplitude 10 μm), the maximum compressive residual stress reaches approximately -120 to -150 MPa at a depth of 100–150 μm beneath the surface. The compressive stress gradually decays with increasing depth but remains significant (approximately -40 to -60 MPa) at depths of 400–500 μm . Liu et al. [36] and Huang et al. [17] reported similar trends in USRP-treated Mg alloys, with the plastically deformed zone extending to depths of 500–600 μm . This literature-supported residual stress profile is consistent with our KAM and hardness observations, providing indirect validation that the 360–540 μm depth range retains beneficial compressive residual stress. These compressive stresses effectively impede the initiation and early growth of surface cracks under cyclic loading, leading to a dramatic improvement in fatigue life that surpasses even the base material. While direct measurement of residual stress would provide more robust evidence, such measurements were not feasible in the present study due to the unavailability of dedicated testing equipment. Instead, we indirectly supported this interpretation using multiple complementary characterizations, including KAM maps and other microstructural indicators. Direct residual stress measurements (e.g., XRD) will be systematically conducted in future investigations to further validate the present findings. Notably, the existing literature has consistently demonstrated that ultrasonic surface rolling processing effectively introduces beneficial residual stresses within the material, thereby contributing to enhanced fatigue performance [41]. Specifically, higher KAM values correspond to higher dislocation density and, consequently, higher residual compressive stress.

The final decline in fatigue life at a removal depth of 600 μm occurs as the majority of the plastically deformed layer, along with its associated beneficial compressive stress field, is removed. The exposed surface condition approaches that of the untreated base material. Consequently, the fatigue performance reverts to a level closer to that of the substrate, losing the significant enhancement provided by the deep compressive stress layer.

This layered fatigue response underscores a fundamental principle in surface-engineered materials: fatigue resistance is not merely a function of surface quality or hardness, but rather a complex outcome of the interplay between defect distribution, microstructure, and residual stress through the affected depth [42]. The results suggest that for USRP-treated Mg alloys, achieving superior fatigue performance requires not only grain refinement but also the preservation of a deep, stable compressive stress zone. This

has important implications for optimizing USRP parameters to maximize the depth and magnitude of beneficial residual stresses while minimizing surface damage.

The localized micro-plastic deformation occurring at 240 MPa plays a critical role in understanding the fatigue mechanism. Such micro-plastic activity preferentially concentrates at grain boundaries, twin interfaces, and defect-concentrated gradient layers, promoting dislocation pile-up and acting as preferential sites for fatigue crack initiation. Despite the absence of macroscopic plastic strain, cyclic micro-plastic accumulation accelerates the formation of persistent slip bands and further facilitates early crack nucleation. Meanwhile, the gradient-refined microstructure and abundant grain boundaries induced by USRP can effectively constrain the expansion of local plastic zones, hinder crack propagation, and ultimately enhance the high-cycle fatigue performance of AZ31 magnesium alloy under slightly over-yield cyclic loading.

In summary, the fatigue life after layer removal is determined by the sequential exposure of microstructural zones with distinct combinations of grain morphology and residual stress states. The trend illustrates that while a refined, recrystallized structure is beneficial, the presence of high residual compressive stresses in the deformed subsurface layer plays a more dominant role in achieving superior fatigue resistance in this USRP-treated magnesium alloy.

Figure 10 presents the SEM morphologies of the fatigue fracture surfaces of AZ31B magnesium alloy specimens under different conditions at a stress amplitude of 240 MPa. The fracture surface of the BM exhibits typical fatigue failure characteristics, with crack initiation occurring at the surface and a large stable crack propagation zone. In contrast, the fracture surfaces of the specimens with removal depths of 80 μm and 500 μm are relatively smooth. Specifically, the crack initiation zone in Figure 10b shifts to a subsurface depth of 2277 μm , while that in Figure 10c moves to 1570 μm beneath the surface. The transition of crack initiation sites from the free surface to the subsurface well explains the significant improvement in fatigue life presented in the preceding sections.

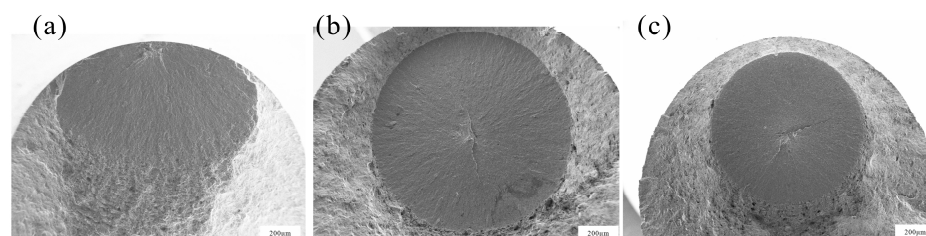


Figure 10. SEM images of fatigue fracture surfaces of AZ31B magnesium alloy under different treatment processes: (a) BM; (b) removal depth of 80 μm ; (c) removal depth of 500 μm .

Figure 11 presents the subsurface layered structure from the treated surface to the bulk matrix and its correlation with the non-monotonic fatigue life trend. The outermost defect-rich layer (0–80 μm) with severe surface damage and low EBSD indexing rate acts as a preferential crack initiation site, leading to low fatigue life for the as-USRP sample. Removing this layer exposes the fully recrystallized fine-grained layer (80–360 μm) with high hardness and low strain energy, resulting in the first fatigue life peak at 80 μm removal. The underlying high-dislocation-density deformed layer (360–540 μm) with beneficial residual compressive stress hinders crack propagation, giving rise to the second fatigue life peak at 400–500 μm removal. Beyond 540 μm , the microstructure approaches the original coarse-grained matrix, and the fatigue life gradually decreases to the base material level. This schematic clarifies that the subsurface gradient structure, rather than the outermost surface, dominates the fatigue performance of USRP-treated AZ31B Mg alloy.

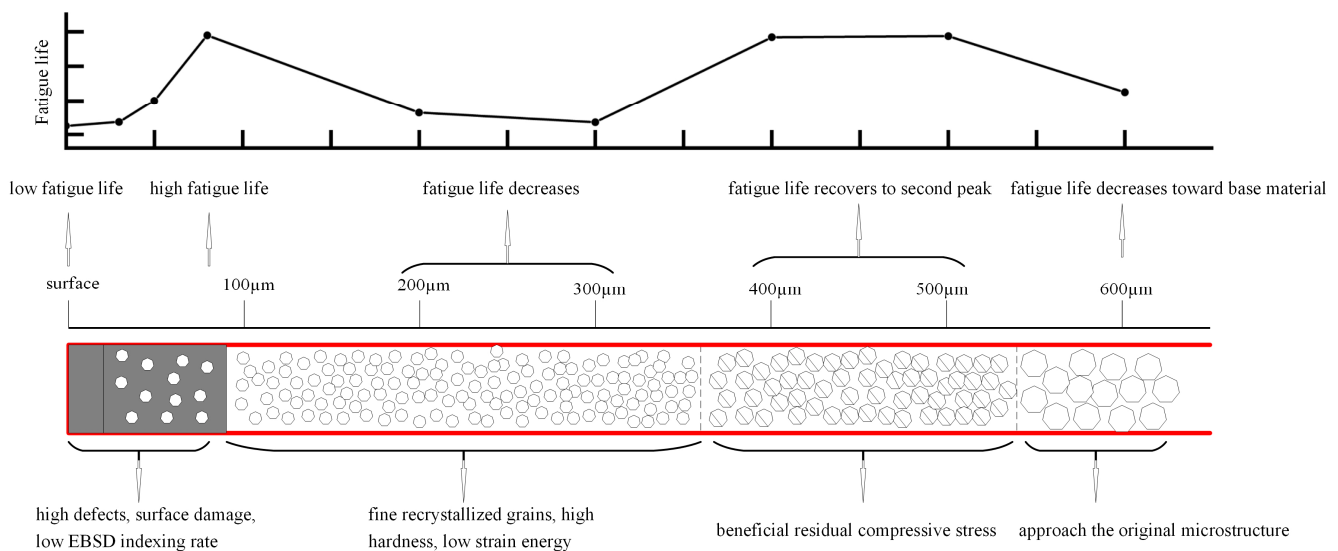


Figure 11. Schematic diagram of the layered fatigue mechanism.

The microstructure (EBSD) identifies three distinct zones: a defect-rich surface layer, a recrystallized zone (180–360 μm) with high mechanical stability, and a deformed zone (360–540 μm) with retained dislocations indicating compressive stress. Hardness profiling quantifies depth-dependent strengthening, with sustained hardness in the deformed zone confirming retained work hardening. Indirect residual stress evidence from KAM values and the literature supports the presence of compressive stress in the 360–540 μm range. Fractography shows subsurface crack initiation at 80 μm and 400–500 μm removal depths, correlating with the exposure of either the stable recrystallized zone or the compressive-stressed deformed zone. Together, these integrated analyses demonstrate that fatigue performance is governed by the interplay between microstructural stability and residual stress across distinct subsurface layers.

4. Conclusions

The influence of surface layer removal on the tension–tension fatigue performance of USRP-treated AZ31B magnesium alloy was systematically investigated. This study focuses on revealing how the gradient microstructure at different depth layers governs fatigue behavior by means of a layer-by-layer surface removal strategy, rather than seeking optimal USRP process parameters. The main conclusions are summarized as follows:

- (1) Fatigue life exhibits a significant and non-monotonic dependence on removal depth. The as-rolled state shows limited improvement, while removing specific surface layers leads to dramatic fatigue life enhancement—by up to two orders of magnitude—demonstrating that the subsurface gradient structure, rather than the immediate surface, governs fatigue behavior.
- (2) Optimal fatigue resistance is achieved after removing approximately 80 μm of the surface material under the specific test conditions employed in this study (stress amplitude 240 MPa, stress ratio $R = 0.1$, room temperature). At this depth, fatigue life reaches a maximum of 7.79×10^6 cycles. This improvement results from eliminating the severely deformed, defect-rich surface layer and exposing a refined, recrystallized subsurface zone that effectively resists crack initiation and propagation.
- (3) A secondary fatigue life peak occurs after removing 400–500 μm of material. At 400 μm removal, fatigue life reaches 7.00×10^6 cycles, and at 500 μm removal, a comparable life of 7.42×10^6 cycles is observed. This recovery is attributed to the exposure of a deeper plastically deformed region containing high dislocation density and a

favorable residual compressive stress field, which strongly suppresses crack initiation under cyclic loading.

- (4) The fatigue response is structurally layered, governed by the gradient microstructure introduced by USRP. The variation in fatigue life with removal depth reflects the sequential exposure of distinct microstructural zones—each with unique combinations of grain morphology, defect concentration, and residual stress state—highlighting the importance of integrated surface–subsurface design in fatigue-resistant engineering.

Author Contributions: Conceptualization, Q.C.; Methodology, F.W.; Software, Z.W., J.M., S.L. and F.W.; Validation, J.M. and Q.C.; Formal analysis, J.M.; Investigation, S.L.; Data curation, J.M.; Writing—original draft, J.M.; Writing—review & editing, Z.W. and J.S.; Visualization, S.L.; Supervision, Z.W., Q.C. and J.S.; Project administration, Z.W.; Funding acquisition, Z.W. and J.S. All authors have read and agreed to the published version of the manuscript.

Funding: This work was supported by the Natural Science Foundation of Shandong Province (ZR2022ME186), the National Natural Science Foundation of China (52001188) and the Jinan Municipal-University Integration Development Strategy Project (JNSX2025022).

Data Availability Statement: The original contributions presented in this study are included in the article. Further inquiries can be directed to the corresponding author.

Conflicts of Interest: The authors declare no conflict of interest.

References

1. Miura, H.; Maruoka, T.; Yang, X.; Jonas, J.J. Microstructure and mechanical properties of multi-directionally forged Mg–Al–Zn alloy. *Scr. Mater.* **2012**, *66*, 49–51. [\[CrossRef\]](#)
2. Joost, W.J.; Krajewski, P.E. Towards magnesium alloys for high-volume automotive applications. *Scr. Mater.* **2017**, *128*, 107–112. [\[CrossRef\]](#)
3. Prasad, S.V.S.; Prasad, S.B.; Verma, K.; Mishra, R.K.; Kumar, V.; Singh, S. The role and significance of Magnesium in modern day research-A review. *J. Magnes. Alloys* **2022**, *10*, 1–61. [\[CrossRef\]](#)
4. Agnew, S.R.; Duygulu, Ö. Plastic anisotropy and the role of non-basal slip in magnesium alloy AZ31B. *Int. J. Plast.* **2005**, *21*, 1161–1193. [\[CrossRef\]](#)
5. Basu, S.; Dogan, E.; Kondori, B.; Karaman, I.; Benzerger, A.A. Towards designing anisotropy for ductility enhancement: A theory-driven investigation in Mg-alloys. *Acta Mater.* **2017**, *131*, 349–362. [\[CrossRef\]](#)
6. Ye, H.; Sun, X.; Liu, Y.; Rao, X.X.; Gu, Q. Effect of ultrasonic surface rolling process on mechanical properties and corrosion resistance of AZ31B Mg alloy. *Surf. Coat. Technol.* **2019**, *372*, 288–298. [\[CrossRef\]](#)
7. Tong, J.L.; Zheng, Y.H.; Yang, S.K.; Li, X.B.; Liu, L.X.; Xiang, D.H.; Gao, G.F. Research on the Surface Characteristics of TC4 Titanium Alloy Subjected to Eccentric Ultrasonic Surface Rolling Process. *J. Mater. Eng. Perform.* **2025**, 1–17. [\[CrossRef\]](#)
8. Geng, J.L.; Yan, Z.F.; Zhang, H.X.; Liu, Y.Q.; Dong, P.; Yuan, S.; Wang, W.X. Microstructure and Mechanical Properties of AZ31B Magnesium Alloy via Ultrasonic Surface Rolling Process. *Adv. Eng. Mater.* **2021**, *23*, 2100076. [\[CrossRef\]](#)
9. Wang, S.; Yu, T.; Pang, Z.; Yin, X.; Liu, X. Effect of Repeated Processing Passes during Ultrasonic Rolling on Fatigue Performance and Corrosion Resistance of Ti6Al4V Alloy. *Metals* **2023**, *13*, 1719. [\[CrossRef\]](#)
10. Wang, H.B.; Song, G.L.; Tang, G.Y. Enhanced surface properties of austenitic stainless steel by electropulsing-assisted ultrasonic surface rolling process. *Surf. Coat. Technol.* **2015**, *282*, 149–154. [\[CrossRef\]](#)
11. Ma, X.Q.; Xu, S.B.; Xue, X.M.; Zhao, X.H.; Pan, Y.F.; Li, J.N.; Zheng, W. Effect of Multi-pass Ultrasonic Surface Rolling Process on Surface Properties and Microstructure of GCr15 Steel. *J. Mater. Eng. Perform.* **2025**, *34*, 17354–17366. [\[CrossRef\]](#)
12. Wu, L.J.; Lv, Y.X.; Zhang, Y.L.; Yang, L.J.; Yang, Y.F.; Li, A.H. Surface integrity and rolling contact fatigue behavior of 18CrNiMo7-6 steel subjected to ultrasonic surface rolling process. *Eng. Fail. Anal.* **2024**, *162*, 108442. [\[CrossRef\]](#)
13. Liang, C.; Yan, H.; Yin, Y.; Hu, H.; Li, L. Effect of Ultrasonic Surface Rolling Step Size on the Wear and Corrosion Behavior of Shot-Peened Cr8 Steel. *Metals* **2026**, *16*, 51. [\[CrossRef\]](#)
14. Jiang, L.K.; Feng, X.W.; Wu, H.C.; Su, G.S.; Yang, B. Improved Microstructure of 316LN Stainless Steel Performed by Ultrasonic Surface Rolling. *Metals* **2025**, *15*, 545. [\[CrossRef\]](#)
15. Yang, H.M.; Yang, K.; Wei, G.B.; Li, R.G. Optimization of Surface Layer Properties of Mg-9Li-1Zn Alloy by Ultrasonic Surface Rolling Process and its Impact on Corrosion Behavior. *Acta Metall. Sin. (Engl. Lett.)* **2025**, *38*, 1421–1435. [\[CrossRef\]](#)

16. Zheng, G.Y.; Luo, X.; Kou, Z.D.; Huang, B.; Yang, Y.Q. Microstructural evolution of Al-Zn-Mg-Cu alloy during ultrasonic surface rolling process. *Mater. Charact.* **2022**, *194*, 112418. [\[CrossRef\]](#)
17. Huang, L.S.; Liu, Y.; Wu, J.H.; Guo, Y.X.; Feng, M.L.; Lian, J.M.; Zhou, M.H.; Huang, H. Effect of ultrasonic surface rolling process on microstructure, mechanical properties, and wear resistance of Mg-1.9Mn-0.3Ce alloy. *J. Alloys Compd.* **2025**, *1034*, 181372. [\[CrossRef\]](#)
18. Xu, F.M.; Huang, L.S.; Liu, G.H. Effect of ultrasonic surface rolling process on the surface properties of Mg-Gd-Zn-Zr alloy. *Mater. Lett.* **2024**, *365*, 136398. [\[CrossRef\]](#)
19. Han, J.; Wang, C.; Song, Y.M.; Liu, Z.Y.; Sun, J.P.; Zhao, J.Y. Simultaneously improving mechanical properties and corrosion resistance of as-cast AZ91 Mg alloy by ultrasonic surface rolling. *Int. J. Miner. Metall. Mater.* **2022**, *29*, 1551–1558. [\[CrossRef\]](#)
20. He, J.H.; Geng, X.; Dong, Q.S.; Zhang, Y.; Zhang, X.B. Simultaneously improving mechanical properties and corrosion resistance of Mg-3Gd-1Zn-0.4Zr alloy via ultrasonic surface rolling. *J. Alloys Compd.* **2025**, *1038*, 182897. [\[CrossRef\]](#)
21. Wang, S.; Li, W.; Chen, L.B.; Luo, L.; Meng, G.; Chang, G.Q.; Liu, Q.D. Formation mechanism of surface gradient microstructure and mechanical properties evolution of Mg-Y-Nd-Gd-Zr alloy by ultrasonic rolling. *J. Mater. Res. Technol.* **2024**, *30*, 6482–6497. [\[CrossRef\]](#)
22. Yang, Z.L.; Liu, Y.D.; Tang, Y.; Li, W.; Zhang, S.; Wang, H.J.; Ran, X. Effect of USRP on corrosion properties of Mg-Y-Nd-Gd-Zr magnesium alloy. *Mater. Chem. Phys.* **2025**, *344*, 131042. [\[CrossRef\]](#)
23. Yu, D.L.; Zhang, D.F.; Dai, Q.W.; Lan, W.; Peng, J.; Xu, J.Y.; Qi, F.G.; Pan, F.S. Effect of stress ratio on high cycle fatigue properties in Mg-6Zn-1Mn alloy. *Mater. Sci. Eng. A-Struct. Mater. Prop. Microstruct. Process.* **2018**, *711*, 624–632. [\[CrossRef\]](#)
24. Tsushida, M.; Shikada, K.; Kitahara, H.; Ando, S.; Tonda, H. Relationship between fatigue strength and grain size in AZ31 magnesium alloys. *Mater. Trans.* **2008**, *49*, 1157–1161. [\[CrossRef\]](#)
25. Begum, S.; Chen, D.L.; Xu, S.; Luo, A.A. Low cycle fatigue properties of an extruded AZ31 magnesium alloy. *Int. J. Fatigue* **2009**, *31*, 726–735. [\[CrossRef\]](#)
26. Gao, J.; Wang, S.B.; Cai, J.X.; Xin, W.L.; Yan, Z.F. Study on the synergistic strengthening mechanism of pre-tension deformation and surface mechanical rolling treatment on the fatigue performance of AZ31B magnesium alloy. *J. Magnes. Alloys* **2025**, *13*, 4500–4516. [\[CrossRef\]](#)
27. Xu, X.C.; Liu, D.X.; Zhang, X.H.; Liu, C.S.; Liu, D.; Zhang, W.C. Influence of ultrasonic rolling on surface integrity and corrosion fatigue behavior of 7B50-T7751 aluminum alloy. *Int. J. Fatigue* **2019**, *125*, 237–248. [\[CrossRef\]](#)
28. Lan, S.L.; Qi, M.; Zhu, Y.F.; Liu, M.X.; Bie, W.B. Ultrasonic rolling strengthening effect on the bending fatigue behavior of 12Cr2Ni4A steel gears. *Eng. Fract. Mech.* **2023**, *279*, 109024. [\[CrossRef\]](#)
29. Yang, J.; Liu, D.X.; Ren, Z.C.; Zhi, Y.L.; Zhang, X.H.; Zhao, R.M.; Liu, D.; Xu, X.C.; Fan, K.F.; Liu, C.S.; et al. Grain growth and fatigue behaviors of GH4169 superalloy subjected to excessive ultrasonic surface rolling process. *Mater. Sci. Eng. A* **2022**, *839*, 142875. [\[CrossRef\]](#)
30. Wang, X.; Chen, L.; Liu, P.; Lin, G.; Ren, X. Enhancement of Fatigue Endurance Limit through Ultrasonic Surface Rolling Processing in EA4T Axle Steel. *Metals* **2020**, *10*, 830. [\[CrossRef\]](#)
31. GB/T 3075-2021; Metallic Materials — Fatigue Testing — Axial Force-Controlled Method. Standards Press of China: Beijing, China, 2021.
32. Yazdanmehr, A.; Jahed, H. On the Surface Residual Stress Measurement in Magnesium Alloys Using X-Ray Diffraction. *Materials* **2020**, *13*, 5190. [\[CrossRef\]](#)
33. Liu, R.X.; Song, Y.F.; Wang, Y.Z.; Yan, H.; Zhao, L.Y.; Chen, R.S. High strength AZ31 magnesium alloy sheets without sacrificing ductility enabled by ultrasonic surface rolling processing introducing micro/ nano-gradient structure. *J. Alloys Compd.* **2025**, *1036*, 181718. [\[CrossRef\]](#)
34. Ma, C.; Zhang, C.H.; Zhu, W.G.; Zeng, X.K.; Xie, Z.H.; Wang, X.Y. Effect of Ultrasonic Surface Rolling Process on Pre-corrosion Fatigue Behavior of Zr705 Alloy. *Rare Met. Mater. Eng.* **2024**, *53*. [\[CrossRef\]](#)
35. Chen, L.T.; Wang, Y.S.; Xu, S.H.; Zhang, M.Y.; Zheng, G.H. The Effect of USRP-Composite DLC Coating on Bearing Fatigue Life. *Coatings* **2025**, *15*, 616. [\[CrossRef\]](#)
36. Liu, Y.D.; Li, W.; Sun, Y.D.; Chen, L.B.; Chang, G.Q.; Deng, G.S. Effect of ultrasonic surface rolling process on microstructure and properties of rolled Mg-Y-Nd-Zr alloy. *J. Mater. Sci.* **2023**, *58*, 9362–9381. [\[CrossRef\]](#)
37. Liu, R.X.; Li, X.H.; Song, Y.F.; Wang, Z.J.; Zhao, L.Y.; Ning, F.Q.; Yan, H.; Chen, R.S. The effect of ultrasonic surface rolling process on the gradient microstructure and wear resistance of AZ31 thin sheet. *J. Magnes. Alloys* **2025**, *13*, 5077–5090. [\[CrossRef\]](#)
38. John, M.; Ralls, A.M.; Dooley, S.C.; Thazhathidathil, A.K.V.; Perka, A.K.; Kuruveri, U.B.; Menezes, P.L. Ultrasonic Surface Rolling Process: Properties, Characterization, and Applications. *Appl. Sci.* **2021**, *11*, 10986. [\[CrossRef\]](#)
39. Zhao, X.H.; Zhang, Y.J.; Liu, Y. Surface Characteristics and Fatigue Behavior of Gradient Nano-Structured Magnesium Alloy. *Metals* **2017**, *7*, 62. [\[CrossRef\]](#)

40. Serrano-Munoz, I.; Mishurova, T.; Thiede, T.; Sprengel, M.; Kromm, A.; Nadammal, N.; Nolze, G.; Saliwan-Neumann, R.; Evans, A.; Bruno, G. The residual stress in as-built Laser Powder Bed Fusion IN718 alloy as a consequence of the scanning strategy induced microstructure. *Sci. Rep.* **2020**, *10*, 14645. [[CrossRef](#)]
41. Zhao, W.D.; Liu, D.X.; Zhang, X.H.; Zhou, Y.; Zhang, R.X.; Zhang, H.; Ye, C. Improving the fretting and corrosion fatigue performance of 300M ultra-high strength steel using the ultrasonic surface rolling process. *Int. J. Fatigue* **2019**, *121*, 30–38. [[CrossRef](#)]
42. Xu, S.B.; Wang, Y.R.; Sun, K.W.; Zhou, H.L.; Pan, Y.F.; Lin, X.J.; Li, T.T.; Li, T.H.; Xu, C.; Zheng, W.; et al. Ultrasonic Surface Rolling Process with Multi-Parameter Coupling for Surface Integrity and Fatigue Life Enhancement of 20CrNiMo Steel. *J. Mater. Eng. Perform.* **2025**, *35*, 9. [[CrossRef](#)]

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