



Thermal stability of surface residual stresses in gear shaft subjected to ultrasonic impact treatment: A finite element simulation study

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Abstract. In-service thermal loading of gear shafts may relax the residual compressive stresses introduced by ultrasonic impact treatment (UIT), thereby degrading the surface strengthening effect. In this study, a 30CrMoA gear shaft is investigated using integrated finite element simulation and experimental approaches to systematically evaluate the effects of UIT on surface residual stress, microhardness, surface roughness, and microstructure, as well as the thermal evolution of residual stress at 65 °C. Results demonstrate that UIT introduced a residual compressive stress field extending to approximately 1.25 mm beneath the surface, with a surface stress of −321 MPa and a peak stress of approximately −724 MPa. Surface hardness increased from 231 HV to 318 HV, with a hardened layer depth of approximately 1.14 mm, while surface roughness Ra is reduced from 0.8 μm to 0.18 μm. Under thermal exposure at 65 °C, surface residual compressive stress exhibited an initially rapid relaxation that subsequently decelerated, stabilizing after 8 h with a total relaxation of approximately 23.94%. Meanwhile, surface hardness, hardened layer depth, and microstructure remained essentially unchanged, indicating excellent retention of the work-hardening effect. Finite element predictions showed excellent agreement with experimental measurements, validating the effectiveness of the numerical model. This study elucidates the decoupled thermal evolution of residual stress relaxation and work-hardening retention in UIT-strengthened surface layers, establishing a theoretical framework for the reliable engineering application of ultrasonic impact treatment to gear shafts operating at elevated temperatures.

Keywords: Gear shaft; Ultrasonic impact treatment; Residual stress relaxation; Surface integrity; Work hardening; Finite element simulation.

1 Introduction

As a key component of mechanical transmission systems, gear shafts play a vital role in transmitting torque and motion. During service, they are frequently subjected to complex cyclic loads, friction, and temperature fluctuations, which can readily induce surface wear, fatigue cracks, and even premature failure. The surface integrity of gear shafts, including the residual stress distribution, microhardness, and surface roughness, directly determines their service life and reliability. Improving the surface properties of gear shafts through advanced surface strengthening technologies is therefore of considerable engineering significance. As a medium-carbon quenched and tempered alloy structural steel, 30CrMoA is widely used in the manufacture



of critical components such as gear shafts, sucker rods, and axles, owing to its excellent combination of mechanical properties and good elevated-temperature strength (Wang et al., 2017).

In practical applications, however, gear shaft components are often exposed to the combined action of cyclic loads and
35 corrosive media under complex operating conditions. In hot and humid environments in particular, the surfaces of 30CrMoA
gear shafts are prone to pitting and stress corrosion, which act as the primary sites of fatigue crack initiation. He et al. showed
that corrosion fatigue is the main cause of fracture in 30CrMoA sucker rods, with all cracks originating at the bottoms of
surface corrosion pits (He et al., 2021). Such damage can be classified as corrosion–fatigue failure, the severity of which far
exceeds that of fatigue or corrosion acting alone. Developing effective surface strengthening techniques to mitigate the
40 synergistic damage of corrosion and fatigue is therefore essential for extending the service life of 30CrMoA gear shafts.
Surface strengthening is an effective means of improving the in-service performance of metallic materials (Zhang et al., 2015).
Among the available techniques, ultrasonic impact treatment (UIT) has attracted considerable attention because it introduces
severe plastic deformation into the surface layer while improving the surface finish and generating a deep residual compressive
stress field. Zhang et al. reported that UIT reduced the surface roughness of 30CrMoA specimens from 3.746 μm to
45 approximately 0.2 μm , with the residual compressive stress layer extending to a depth of nearly 2 mm. More broadly, UIT and
related surface processing techniques have been shown to markedly improve fatigue performance and corrosion resistance by
tailoring residual stresses and enhancing the damping capacity of the material (Zhang et al., 2015). Regarding the influence of
UIT parameters, Bai et al. found that, under otherwise identical impact conditions, a larger amplitude produces a deeper
compressive stress layer and higher surface compressive stresses (Bai et al. 2019). Liu et al. investigated the effect of repeated
50 UIT with different fatigue cycle intervals on the fatigue life of S690QL cruciform welded joints (Liu et al. 2014), while Zhang
et al. examined its effect on pre-fatigued high-strength steel welded joints (Zhang et al. 2015). Tehrani Yekta et al. conducted
fatigue tests on non-load-carrying fillet welds subjected to adequate, insufficient, and excessive degrees of UIT (Tehrani Yekta
et al. 2013). All of these studies demonstrated that UIT significantly enhances the fatigue life of welded specimens. Meng et
al. combined laser biomimetic treatment with UIT on AZ31B magnesium alloy welded joints; fatigue tests showed that, at 10^7
55 cycles, the fatigue strength of the composite-treated specimens was 33 % and 125 % higher than that of the singly treated and
untreated specimens, respectively, owing to a shift in crack initiation location, modified fracture morphology, and multiple
strengthening mechanisms (Meng et al. 2023). Okawa et al. examined the influence of preload and stress ratio on the UIT-
induced improvement of the fatigue strength of welded joints (Okawa et al. 2013). To improve the fatigue performance of
underwater wet welded joints, Gao et al. proposed a combined grinding and underwater UIT method; neither treatment alone
60 significantly improved the fatigue life, whereas their combination increased the fatigue strength at 2×10^6 cycles from 135
MPa to 182 and 217 MPa, corresponding to improvements of 35 % and 61 %, respectively (Gao et al. 2015).

In manufacturing and materials processing, UIT also plays an important role in cladding and additive manufacturing, owing
to its unique energy transfer mechanism and complex physical effects (Tehrani Yekta et al., 2013; Zhao et al., 2023). Song et
al. applied UIT during the preparation of CoCrFeNiCu high-entropy alloy coatings and found that it suppressed the
65 interdendritic segregation of copper at the coating surface, transforming the original two-phase structure, a copper-rich FCC1



phase and a copper-poor FCC2 phase, into a single face-centered cubic (FCC) phase (Song et al. 2023). Li et al. fabricated Ni-based, Ni + WC, and Ni + WC + Cr₃C₂ cladding coatings by combining laser cladding of pre-placed powders with UIT and observed the formation of a distinct plastic deformation layer at the coating surface, together with a transformation of the original network eutectic structure into a granular one (Li et al. 2023). Wang et al. prepared CrCoNi medium-entropy alloy coatings by laser cladding followed by UIT at different amplitudes; the wear resistance increased with increasing impact amplitude, and at an amplitude of 20 μm the wear rate was reduced by 85.71 % relative to the untreated coating (Wang et al. 2023). Hua et al. combined interfacial fusion deposition with UIT and laser processing to investigate the microstructure and tribological properties of Ti₂AlC–B particle-reinforced Ti6Al4V self-lubricating coatings (Hua et al. 2023). Wang et al. further examined the microstructure and mechanical properties of 304 stainless steel specimens produced by directed energy deposition (DED) with and without UIT, revealing a transformation from coarse columnar grains to fine equiaxed grains throughout the deposited layer, which was attributed to recrystallization (Wang et al. 2023). Zhao et al. investigated interlayer UIT of thin-walled AZ31 magnesium alloy components and showed that it reduces surface roughness, increases the effective load-bearing area of the deposited material, and promotes the columnar-to-equiaxed grain transition (Zhao et al. 2023). Taken together, these studies indicate that existing research on ultrasonic impact hardening has focused primarily on the evolution of surface conditions in welding, cladding, and additive manufacturing.

To elucidate the thermal stability of the ultrasonic impact-hardened layer under a temperature rise of 65°C, this study uses 30CrMoA steel as the test material to investigate the evolution of the ultrasonic impact-hardened surface roughness, surface geometry, residual stress field in the surface layer, surface microhardness, and surface microstructure during thermal exposure at 65°C.

2 Theoretical analyses of the ultrasonic impact process

2.1 Hertz's theory of contact

Hertz's contact theory, proposed by Hertz in 1882, is primarily used to describe the deformation behavior of two elastic bodies in contact. This theory mainly addresses the stress distribution and deformation at the point of contact and is particularly applicable to circular contact points, such as the contact between a sphere and a plane or between two spheres. A simplified contact configuration between a sphere and a workpiece is shown in Fig. 1. Johnson's Hertz contact theory (Zheng et al., 2026) elucidates the relationship between contact pressures between two elastic bodies:

$$P_r = P_0 \left[1 - \left(\frac{r}{a} \right)^2 \right]^{1/2} \quad (1)$$

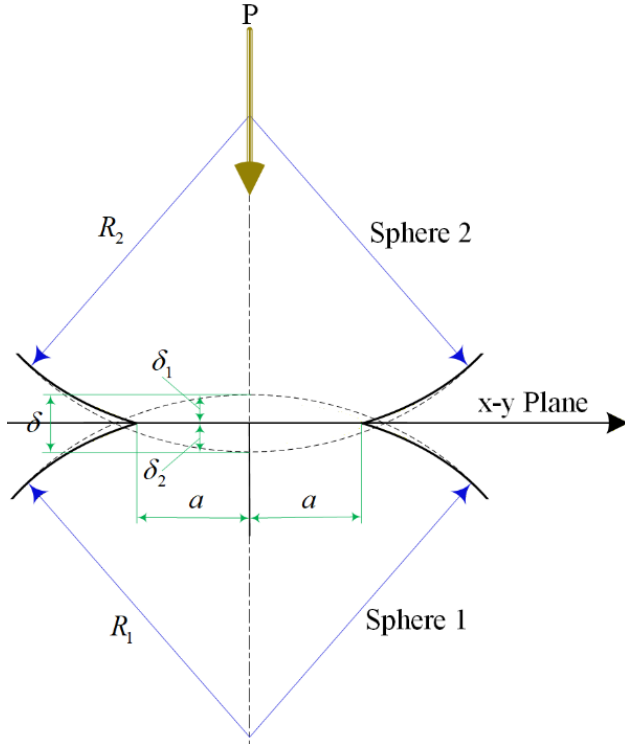
where P_0 is the maximum elastic compressive stress at the center of contact, a is the contact radius, and r is the radial distance from the center of contact to any point.

where E^* is the equivalent elastic modulus of contact between the two objects:

$$\frac{1}{E^*} = \frac{1 - \mu_1^2}{E_1} + \frac{1 - \mu_2^2}{E_2} \quad (2)$$



where E_1 and E_2 represent the elastic moduli of the workpiece and the sphere, respectively; μ_1 and μ_2 represent the Poisson's ratios of the workpiece and the sphere, respectively.



100 **Figure 1. Schematic diagram of a sphere in contact with a workpiece**

During the ultrasonic impact treatment process, the contact pressure between the sphere and the workpiece is composed of both static pressure and dynamic pressure caused by vibration. The formula for calculating the residual stress on the surface after ultrasonic impact treatment is as follows:

$$\sigma_{\text{UIT}}^{\text{res}}(r, A, f) = \left(C_1 (k_1 A)^{1/3} \sqrt{1 - \left(\frac{r}{C_2 (k_1 A)^{1/3}} \right)^2} - \sigma_y \right) \cdot k_f(f) \quad (3)$$

105 where C_1 and C_2 are constants that incorporate material properties and geometric parameters.

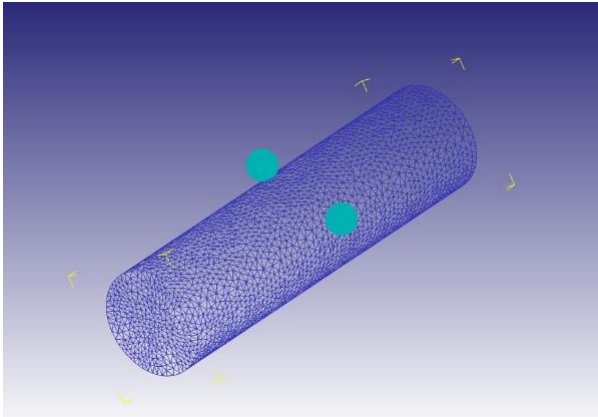
$$\begin{cases} C_1 = \frac{3^{5/3}}{2\pi} \left(\frac{ER}{1-\mu^2} \right)^{2/3} \\ C_2 = \left(\frac{3(1-\mu^2)}{4ER} \right)^{1/3} \end{cases} \quad (4)$$



3 Finite element simulation study of ultrasonic impact

3.1 Create a geometric model

110 The ultrasonic impact process is simulated using the DEFORM finite element software, and the corresponding finite element model is established. To simplify the analysis, the journal and the impact ball were selected as the objects of investigation, as illustrated in Fig. 2. The mechanical and thermal properties of the rolling ball are listed in Table 1. Since the elastic modulus of the rolling ball is considerably higher than that of the 30CrMoA alloy, the rolling ball is modelled as a rigid body. The coefficient of friction at the interface between the WC/Co cemented carbide and the 30CrMoA alloy is set to 0.3.



115 **Figure 2. 3D model for numerical simulation of ultrasonic impact**

3.2 Numerical computation methods

In a stress analysis based on the Lagrangian formulation, a coordinate system is attached to the deformable body under investigation, and its motion and deformation are described by tracking material particles. The governing equations of motion are expressed as follows:

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$$x_i = x_i(X_j, t) \quad (5)$$

where X_j represents the coordinates of the material, t represents the time of particle motion, and $i, j = 1, 2, 3$. This simulation uses the built-in selection solver and the Lagrangian increment algorithm in DEFORM.

In DEFORM deformation–plasticity forming simulations, solving the nonlinear equilibrium equations established by the updated Lagrangian formulation constitutes the core of the numerical procedure, and the efficiency and accuracy of this solution directly determine the reliability of the simulation results as well as the computational cost. Owing to the coupled nature of material, geometric, and boundary nonlinearities inherent to forming processes, iterative methods are the preferred strategy for solving such nonlinear equation systems. Among these, the Newton–Raphson method is adopted due to its quadratic convergence rate and high solution accuracy. To meet the computational requirements of DEFORM, the sparse matrix solver employs methods tailored to symmetric positive-definite sparse matrices, thereby balancing computational accuracy with memory utilization. For small- to medium-scale simulation models, the LDL^T direct method is used to obtain an exact solution of the linear system, thus avoiding the convergence issues associated with iterative methods. For large-scale,

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densely meshed complex forming models, the preconditioned conjugate gradient (PCG) iterative method is employed; by preconditioning the coefficient matrix, this method improves its condition number, accelerates convergence, and significantly reduces memory consumption. The Newton–Raphson method, in combination with the sparse matrix solver, is therefore used to efficiently solve the nonlinear equations arising in the ultrasonic impact simulation.

3.3 Create a constitutive model for the workpiece material

Currently, there are two main material constitutive models widely used in finite element simulations of ultrasonic impact: the Johnson-Cook model (J-C model) and the Zerilli-Armstrong model (Z-A model). The Johnson-Cook model is an empirical model for metallic materials under conditions of large deformation, high strain rates, and high temperatures. Its specific equations are as follows:

$$\sigma = [A + B\varepsilon_p^n][1 + C \ln \varepsilon^*][1 - T^{*m}] \quad (6)$$

where A is the initial yield stress at the reference strain rate and reference temperature; B is the strain hardening modulus of the material; n is the hardening exponent of the material; C is the strain rate strengthening parameter of the material; m is the thermal softening exponent of the material; $\varepsilon^* = \varepsilon/\varepsilon_0$ is the dimensionless strain rate; ε is the actual strain rate; ε_0 is the reference strain rate. $T^* = (T - T_r)/(T_m - T_r)$ is the isothermal coefficient; T_r is the reference temperature; T_m is the melting temperature. Both the J-C model and the Z-A model are empirical models; however, the J-C model requires fewer experimental data and provides a more accurate description of the strain curve during the material's deformation process, making it more widely used. Therefore, the J-C model is selected to describe the mechanical properties of 30CrMoA, as shown in Table 1.

Table 1 Johnson-Cook model parameter values for 30CrMoA

Initial yield stress A/MPa	Strain-hardening modulus B/MPa	Material hardness index n	Reference strain rate $\dot{\varepsilon}_0/(\text{s}^{-1})$	Strain rate hardening parameters C
785	450	0.25	1.0s	0.025

3.4 Loads and boundary conditions

The operating principle of an ultrasonic vibration system is that the electrical energy output by the ultrasonic generator is converted into mechanical vibration by the transducer; the amplifying rod then amplifies this vibrational displacement and transmits it to the impact head. The vibration amplitude at the end face of the impact head is determined by the system's resonant state; it is a controllable and stable kinematic parameter, rather than a constant force output. In actual machining, even when contact conditions fluctuate, the amplitude of the impact head remains essentially constant. Therefore, using displacement amplitude as the input condition better aligns with the physical nature of ultrasonic vibration systems. Furthermore, from the perspective of numerical solution, applying a displacement load directly to the equations of motion as an intrinsic boundary condition avoids the need to iteratively solve for contact forces at each time step, which helps improve the convergence and computational efficiency of solving high-strain-rate impact problems. The radial vibration of the



160 ultrasonic impact head is harmonic, and the variation of its displacement over time can be described by a sine function. The vibrational displacement of the impact head's contact point along the axial direction $z_v(t)$ is

$$z_v(t) = A \sin(\omega t + \varphi) \quad (7)$$

where $\omega = 2\pi f$, f is the ultrasonic vibration frequency; ω is the angular frequency of the ultrasonic vibration; A is the ultrasonic amplitude, which is related to the ultrasonic power and the amplification factor of the horn; φ is the initial phase, which is determined by the initial state of the ultrasonic excitation signal; here, we take $\varphi = 0$.

Based on the motion characteristics of the interaction between the drive shaft and the impact head, a composite feed model is established, in which the impact head moves in a straight line along the axial direction of the drive shaft while the drive shaft rotates simultaneously. When the workpiece undergoes uniform rotational feed, the angular velocity of the workpiece is set to Ω , The center of rotation is at the origin O of the coordinate system. The distance from the point of contact between the impact head and the workpiece surface to the center of rotation is r , and the initial angle is θ_0 , then, the polar coordinates of the point of contact at time t are $(r, \theta(t))$, Among them $\theta(t) = \Omega t + \theta_0$, converting polar coordinates to Cartesian coordinates yields the feed displacement coordinates of the impact head's contact point during rotational feeding:

$$\begin{cases} x_f(t) = r \cos(\Omega t + \theta_0) \\ y_f(t) = r \sin(\Omega t + \theta_0) \\ z_f(t) = 0 \end{cases} \quad (8)$$

where the angular velocity of rotation is $\Omega = \frac{2\pi n}{60}$, n is the rotational speed of the workpiece. At this point, the feed displacement coordinate at the impact head contact point is the sum of the linear feed and the rotational displacement:

$$\begin{cases} x_f(t) = v_f \cdot t + r \cos(\Omega t + \theta_0) \\ y_f(t) = r \sin(\Omega t + \theta_0) \\ z_f(t) = 0 \end{cases} \quad (9)$$

Under the action of preload F_0 , the impact needle comes into contact with the workpiece surface; its motion is divided into a free-flight phase and an impact phase. The dynamic equation for the free-flight phase is:

$$m\ddot{x} + c\dot{x} = F_0 - k(x - x_0) \quad (10)$$

180 where m is the equivalent mass of the impact pin, c is the damping coefficient, k is the stiffness of the return spring, and x_0 is the initial position. The collision phase is modelled using Hertz's contact theory:

$$F_c = \begin{cases} K_h \delta^{3/2} + D_h \dot{\delta} & \delta > 0 \\ 0 & \delta \leq 0 \end{cases} \quad (11)$$



where $\delta = x - x_g$ is the penetration depth, x_g is the position on the workpiece surface, and the contact stiffness is $K_h = \frac{4E^*}{3\sqrt{R}}$,

The Damping Coefficient is $D_h = \alpha K_h \beta^{3/2}$, the material damping factor is α .

185 3.5 Grid division

To ensure the convergence of the finite element simulation results for ultrasonic rolling, the TC11 titanium alloy workpiece was discretized with a mesh size equal to one-twentieth of the ultrasonic impact head diameter (14 mm), a mesh density determined through parametric studies to yield converged equivalent plastic strain (PEEQ) results (Zheng et al., 2026). A multi-level meshing strategy was adopted to balance computational accuracy against efficiency: fine elements with a side
 190 length of 0.7 mm were applied in the processing region, whereas coarser elements (1.5 mm) were gradually introduced in the non-critical regions. In addition, the mesh size in the thickness direction (0.2 mm) was specifically aligned with the residual stress measurement locations. The final model comprised a total of 24,000 elements.

4 Experimental procedures

4.1 Materials and methods

195 A 30CrMoA steel bar was used as the test material, and its chemical composition is listed in Table 2. The specimen preparation procedure consisted of cutting, turning, surface grinding, and heat treatment. The heat treatment comprised a quenching and tempering process: the specimens were oil-quenched at 880 °C and subsequently tempered at approximately 540 °C to achieve a hardness of 220–260 HBW.

Table 2 Chemical composition of 30CrMoA steel

Elements	Mass fraction (%)
C	≤0.31
Mn	≤0.68
Cr	≤1.10
Mo	≤0.23
Si	≤0.32
Ni	≤0.03
S	≤0.025
P	≤0.025

200 4.2 Ultrasonic surface impact process

4.2.1 Components of an ultrasonic impact exercise system

During operation, the ultrasonic transducer converts electrical energy into mechanical vibration, which is amplified by the horn and transmitted to the impact head, thereby generating high-frequency axial vibration of the impact head. Meanwhile, the workpiece motion platform drives the rotation of the workpiece. The orientation of the impact head can be adjusted according
 205 to the machining requirements to maintain an optimal contact angle between the impact head and the workpiece surface. The



ultrasonic impact motion system primarily consists of an ultrasonic generator, an ultrasonic transducer, an amplifying horn, an impact head, and a workpiece motion platform, as shown in Fig. 3.

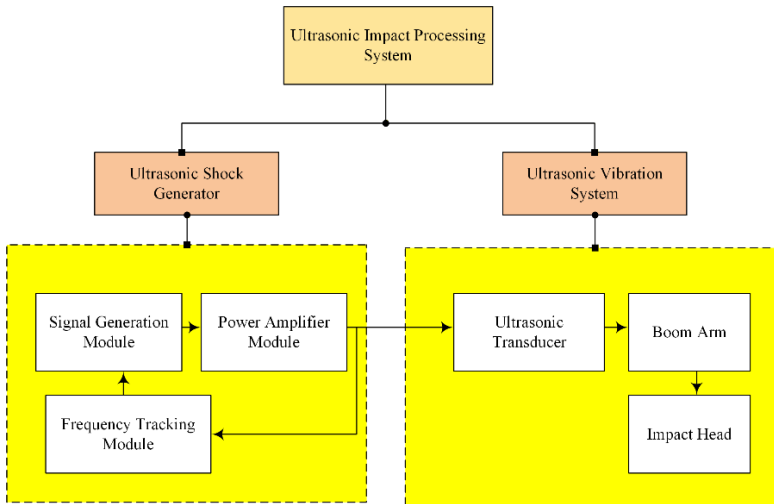


Figure 3. Block diagram of the ultrasonic impact machining system

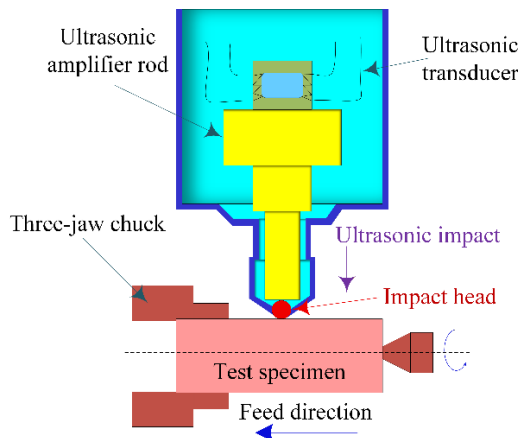


Figure 4. Schematic diagram of ultrasonic impact machining

Prior to the ultrasonic impact process, the contact angle between the impact head and the workpiece surface was adjusted such that the axis of the impact head remained perpendicular to the workpiece surface at all times, thereby ensuring effective transfer of the impact energy, as illustrated in Fig. 4.

The motion of the impact head results from the superposition of multiple movements and can be decomposed into two independent components: (i) the high-frequency, small-amplitude vibration of the impact head along its own axis under ultrasonic excitation, which constitutes the core motion responsible for the impact effect; and (ii) the rotation of the workpiece driven by the motion platform, which determines the overall impact trajectory, together with the linear translation of the impact head along the axial direction of the workpiece. Taking the workpiece as the fixed reference frame (i.e., the inertial coordinate system), a spatial Cartesian coordinate system O-XYZ was established, in which the Z-axis denotes the feed direction of the



workpiece, the Y-axis denotes the tilt direction of the impact head, and the X-axis denotes the normal direction perpendicular to the workpiece surface.

4.2.2 Ultrasonic impact machining process

Table 3 Ultrasonic impact test parameters

Parameter	Value
Vibration frequency	25 kHz
Impact head amplitude	5 μm
Interference	0.01 mm
Spindle speed	100 r/min
Feed rate	0.1 mm/r
Static pressure	600 N

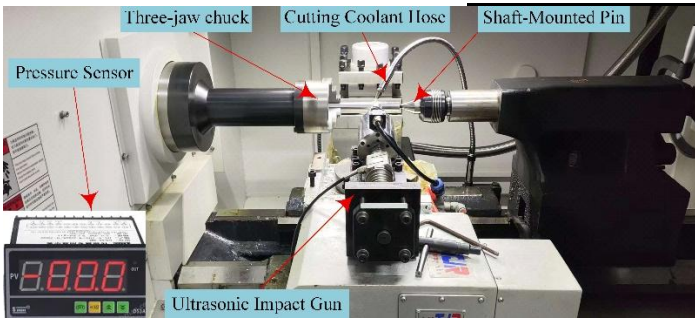


Figure 5. Ultrasonic impact testing equipment

The ultrasonic impact hardening experiments were conducted using an ultrasonic impact device mounted on a CNC machine tool, and the process parameters are listed in Table 3. The impact head was a tungsten carbide ball with a diameter of 14 mm. Prior to the strengthening treatment, a workpiece coordinate system was established on the specimen surface, and the initial gap between the impact head and the workpiece surface was calibrated using a gold foil with a thickness of 0.001 mm to complete the tool setting. Subsequently, a CNC program was compiled according to the interference depth, feed rate, spindle speed, and specimen dimensions, and the machining was performed accordingly. Cooling oil was applied for lubrication throughout the process, and the experimental setup is illustrated in Fig. 5. During the impact process, ultrasonic waves generated by the ultrasonic generator were transmitted to the impact head through the transducer and the amplifying horn. Meanwhile, the impact head was pressed against the workpiece surface with a specified pressure and translated axially at a feed rate f ; thus, the workpiece material, rotating at a rotational speed n , was strengthened by the impact head.

4.3 Equivalent thermal exposure test

4.3.1 Heating equipment and solutions

In the thermal exposure test, the ultrasonically impact-strengthened specimens are held in a constant-temperature environment for a predetermined duration and then removed and cooled to room temperature, thereby simulating the temperature-rise



conditions experienced by a gear shaft in service. The thermal stability of the ultrasonic impact-strengthened layer is evaluated by comparing the residual stress, microstructure, and hardness of the specimens before and after thermal exposure. The thermal exposure test protocol is presented in Table 4.

Table 4 65°C Heat exposure test protocol

Experiment number	Temperature/°C	Processing status
B		Initial state of the matrix
U		Ultrasonic impact
UHT-1	65	1 hour of heat exposure following ultrasonic shock
UHT-2	65	2 hours of heat exposure following ultrasonic shock
UHT-3	65	4 hours of heat exposure following ultrasonic shock
UHT-4	65	8 hours of heat exposure following ultrasonic shock

245 A constant-temperature water bath is selected as the heating equipment for this test, as it heads the specimens uniformly together with the bath medium and provides high temperature-control accuracy as well as excellent temperature uniformity owing to liquid convection, making it suitable for precise constant-temperature simulation. To replicate the operating environment of a gear shaft, the ultrasonic impact-treated specimens are immersed in the water bath, which is set to 65 °C with a temperature-control accuracy of ± 0.5 °C. To avoid additional thermal stress induced by thermal shock, the timing is initiated only after the specimens had been slowly heated from room temperature to 65 °C together with the water bath. The specimens are then held at this temperature for 1, 2, 4, and 8 h, respectively, after which they are removed and allowed to cool naturally to room temperature in air. To exclude the influence of the base material itself, control tests are performed on untreated specimens under identical conditions.

4.4 Methods for characterizing surface conditions

255 4.4.1 Residual stress testing

The surface residual compressive stress is measured using a PROTO residual stress X-ray diffractometer coupled with XrdWin software, as illustrated in Fig. 6. To minimize measurement errors, four measurement points are evenly distributed along the circumference of each specimen; measurements are conducted in the axial direction at each point, and the average value is adopted as the final result.

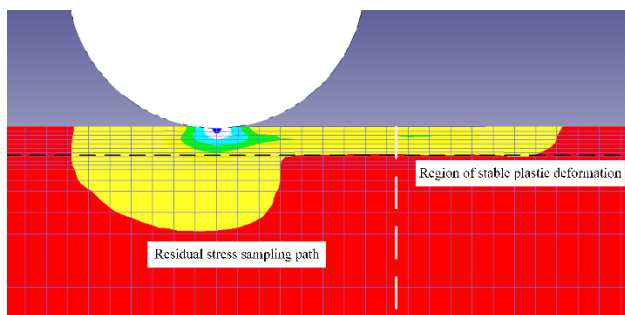


Figure 6 Contour plot of residual stress variation



The testing parameters are as follows: room temperature, target voltage of 24–26 kV, Cr-K α radiation, diffraction angle of 156°, diffraction plane, and X-ray incidence angles ranging from -25° to $+25^\circ$. Prior to testing, the diffractometer is initialized, and the X-ray source is activated and allowed to warm up. Calibration is subsequently performed using a manufacturer-supplied standard sample to ensure the accuracy of the residual stress measurements. The specimens subjected to ultrasonic impact or thermal exposure are then mounted for testing.

To determine the residual stress distribution along the depth direction, the measurement area is progressively removed layer by layer via electrolytic polishing. The polished area is approximately 10 mm $\times$ 10 mm, with a removal depth of approximately 0.05–0.10 mm per layer. Residual stress is measured immediately after each removal step until the measured values approached zero. The residual stress corresponding to each layer depth is recorded, and the depth-dependent residual stress distribution curve is subsequently plotted.

4.4.2 Surface roughness and surface topography

Surface roughness and surface topography are characterized using a MarSurf XR20 profilometer, with a sampling length of 0.8 mm and an evaluation length of 5.6 mm. To minimize measurement errors, four measurement positions are evenly distributed along the circumference of each specimen; surface roughness is measured in the axial direction at each position, and the average value is adopted as the final result. Three-dimensional surface morphology is acquired using the 3D topography module of the MarSurf XR20, with a scanning area of 1 mm $\times$ 1 mm, a sampling interval of 0.02 mm, and a scanning speed of 0.5 mm/s.

4.4.3 Measurement of microstructural hardness gradients and microstructural characterization

Specimens for surface microhardness and microstructure characterization are sectioned by electrical discharge machining (EDM) and subsequently prepared as metallographic samples. The metallographic preparation followed the standard procedure of sectioning $\rightarrow$ mounting $\rightarrow$ grinding $\rightarrow$ polishing. Specifically, the specimens are cut along the axial direction, mounted in JCM1 metallographic mounting resin, ground with SiC abrasive papers, and polished to ensure that the prepared surface is flat and perpendicular to the impacted surface. Surface microhardness is measured using an FMG800 Vickers microhardness tester, and the hardness values are determined from the measured indentation diagonals. To minimize measurement errors, microhardness is measured point by point along the direction perpendicular to the impacted surface (i.e., the depth direction). To avoid interactions between adjacent indentations while keeping them clearly distinguishable, the indentations are arranged in a staggered (diagonal) pattern, with a center-to-center spacing of 10 μ m between adjacent indentations along the depth direction and a lateral offset of less than 30 μ m. A load of 9.807 N is applied for each indentation with a dwell time of 10 s.

5 Analysis of results

5.1 Patterns of residual stress evolution

5.1.1 Simulated residual stress field distribution after ultrasonic impact

Residual stress is a critical indicator for evaluating the effectiveness of ultrasonic impact treatment (UIT). UIT introduces compressive residual stress into the surface layer of the material, which counteracts the external tensile stresses encountered



295 during service, thereby retarding the initiation and propagation of fatigue cracks. However, when the component operates at the service temperature of the gear shaft (65 °C), thermally activated dislocation motion and annihilation may lead to relaxation of the induced compressive residual stress. Excessive stress relaxation would compromise the strengthening effect of UIT. Fig. 7 presents the simulated contour map of the surface residual stress distribution in the 30CrMoA specimen after UIT, obtained using DEFORM software. The contour map reveals a well-defined compressive residual stress zone centered at the impact location, exhibiting an approximately hemispherical distribution with stress magnitude decreasing gradually from the surface toward the interior. The stress-affected region is concentrated within a depth of approximately 1.20 mm from the impacted surface. Figure 7 shows the distribution curve of surface residual stress with depth.

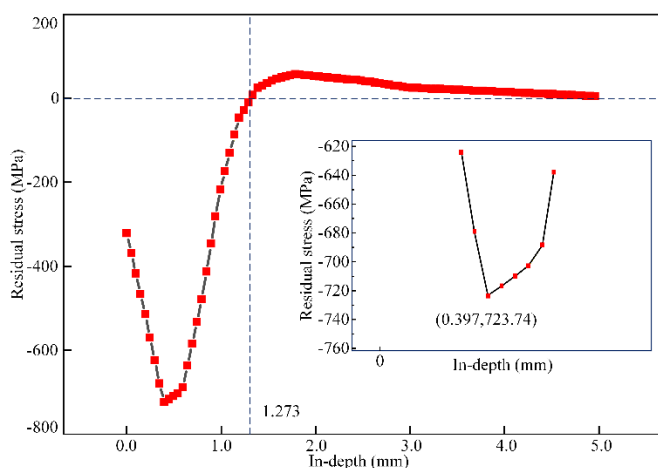


Figure 7 Graph showing the trend of residual stress changes in ultrasonic impact simulation

305 As shown in Fig. 7, the surface residual stress of the untreated specimen (without UIT) is approximately 220 MPa and decayed rapidly to zero with increasing depth. After UIT, pronounced compressive residual stress is introduced into the specimen surface, with a surface compressive stress of approximately 310 MPa and a compressive stress layer extending to a depth of approximately 1273 μm . With increasing distance from the surface, the compressive residual stress first increased and then decreased, reaching a peak value of approximately 720.24 MPa at a depth of about 375 μm , before gradually decaying toward zero. These results indicate that UIT generated a residual compressive stress field with typical gradient characteristics in the surface layer of the 30CrMoA steel.

Regarding the stress gradient, the stress varied most steeply within the depth range of 0–375 μm , with a gradient magnitude of approximately 1.04 MPa/ μm , whereas within the range of 375–1273 μm , the stress decayed gradually. This indicates that the plastic deformation induced by ultrasonic impact was predominantly concentrated in the shallow surface layer.

315 5.1.2 Evolution of residual stresses after thermal exposure

Fig. 8 presents the evolution of surface residual stress in the UIT specimens after thermal exposure at 65 °C for different durations (1, 2, 4, and 8 h). The surface compressive residual stress decreased progressively with increasing exposure duration. After 1 h of thermal exposure, the surface residual stress decreased from an initial value of 320.98 MPa to 273.55 MPa,



corresponding to a relaxation rate of approximately 14.78%; after 2 h, it further decreased to 242.03 MPa; after 4 h, the value is 244.15 MPa; and after 8 h, it stabilized at approximately 239.68 MPa, yielding a total relaxation rate of about 23.94%.

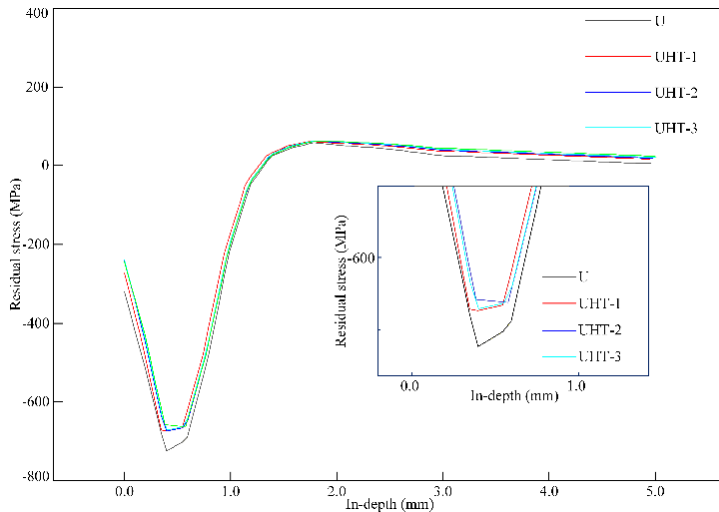


Figure 8 Changes in residual stress after different durations of heat exposure at 65°C

Thermal exposure not only reduced the surface residual stress but also induced a notable reduction in the peak compressive stress. After 8 h of thermal exposure at 65 °C, the peak compressive stress decreased from an initial value of 723.74 MPa to 662.46 MPa. The compressive stress decreased at all depths, with the most pronounced reductions occurring in the surface layer and the peak-stress region, whereas the stress variation in the core (at depths exceeding 961 μm) was negligible. These findings indicate that thermal exposure primarily affects the plastic deformation layer introduced by ultrasonic impact.

5.1.3 Analysis of the mechanism of stress relaxation

Since 65 °C is well below both the phase transformation temperature and the tempering temperature (approximately 540 °C) of 30CrMoA steel, the relaxation of residual stress during thermal exposure does not involve any phase transformation mechanism; rather, it is primarily governed by the thermally activated recovery of dislocations. Under the applied temperature field, the high-density dislocations introduced by ultrasonic impact undergo rearrangement and annihilation, resulting in a reduction of microstrain, which manifests macroscopically as the release of residual stress. This interpretation is corroborated by the subsequent microhardness measurements and microstructural observations.

5.1.4 Comparison and validation of simulation and experiment results

To validate the established finite element model, the simulated results at various state points were compared with the corresponding experimental measurements. Fig. 9 presents a comparison between the simulated and measured residual stress distributions along the depth direction in the as-impacted condition. The two curves exhibit good agreement in their overall trends, both displaying the typical gradient distribution characterized by surface compressive stress → peak value → gradual decay to zero. With regard to the key characteristic parameters, the simulated peak stress (−723.74 MPa) deviates by only approximately 1.9% from the measured value (−710 MPa); the simulated depth of the peak stress is approximately 375 μm,



compared with the measured value of approximately 360 μm , corresponding to a deviation of approximately 4.2%; and the simulated compressive stress layer depth is approximately 1.253 mm, versus the measured value of approximately 1.2 mm, with a deviation of approximately 4.4%. Regarding the stress evolution after thermal exposure, the simulated surface residual stress after 8 h of exposure at 65 °C is -239.68 MPa, compared with the measured value of -244.15 MPa, corresponding to a relative deviation of approximately 1.8%. The deviations at all comparison points remain within 5%, demonstrating that the established DEFORM finite element model can reliably predict both the residual stress field induced by ultrasonic impact and its evolution under thermal exposure at 65 °C.

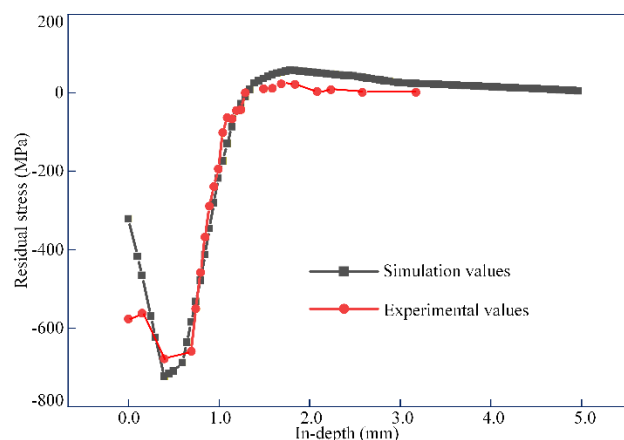


Figure 9 Comparison of residual stress changes between simulation and experiment

The minor discrepancies between the simulation and experimental results can be primarily attributed to the following factors. The parameters of the Johnson–Cook constitutive model were adopted from the literature; subtle differences between these reference values and the actual material properties may have introduced deviations into the stress calculations. The impact ball is modelled as a rigid body in the simulation, neglecting its elastic deformation during the impact process. The residual stress measured by X-ray diffraction represents an average value over the irradiated area, whereas the simulation outputs nodal stress values at discrete locations; this difference in spatial resolution between the two approaches also contributes to the observed deviations. Nevertheless, all deviations remain within an acceptable range, confirming the validity of the model and providing a reliable computational tool for the subsequent optimization of process parameters.

5.2 Patterns of roughness and surface topography evolution

5.2.1 The effect of ultrasonic impact on surface roughness

Surface roughness directly affects the contact stiffness, fitting accuracy, and fatigue resistance of a component. The high-frequency mechanical action of ultrasonic impact not only introduces compressive residual stress but also reduces surface roughness and removes machining marks through rolling and plastic flow of the surface material. Under thermal exposure at 65 °C, potential changes in surface topography (e.g., thermal deformation or localized bulging) may alter the actual contact state and the degree of stress concentration of the component, thereby indirectly influencing the residual stress measurements.



Therefore, quantitative characterization of the surface roughness and three-dimensional topography before and after thermal exposure allows the interference of topography changes on the stress analysis to be excluded and the stability of the surface quality to be evaluated independently.

Fig. 10 presents the surface roughness results of the 30CrMoA specimen before and after UIT. The original ground specimen exhibited a surface roughness (R_a) of approximately $0.8 \mu\text{m}$, which decreased to $0.18 \mu\text{m}$ after UIT, indicating a significant improvement in surface quality. This improvement can be attributed to the high-frequency vibration of the impact head, which exerts a rolling and flattening effect on microscopic surface asperities, while the severe plastic deformation drives surface material flow that fills the microscopic grooves of the original surface

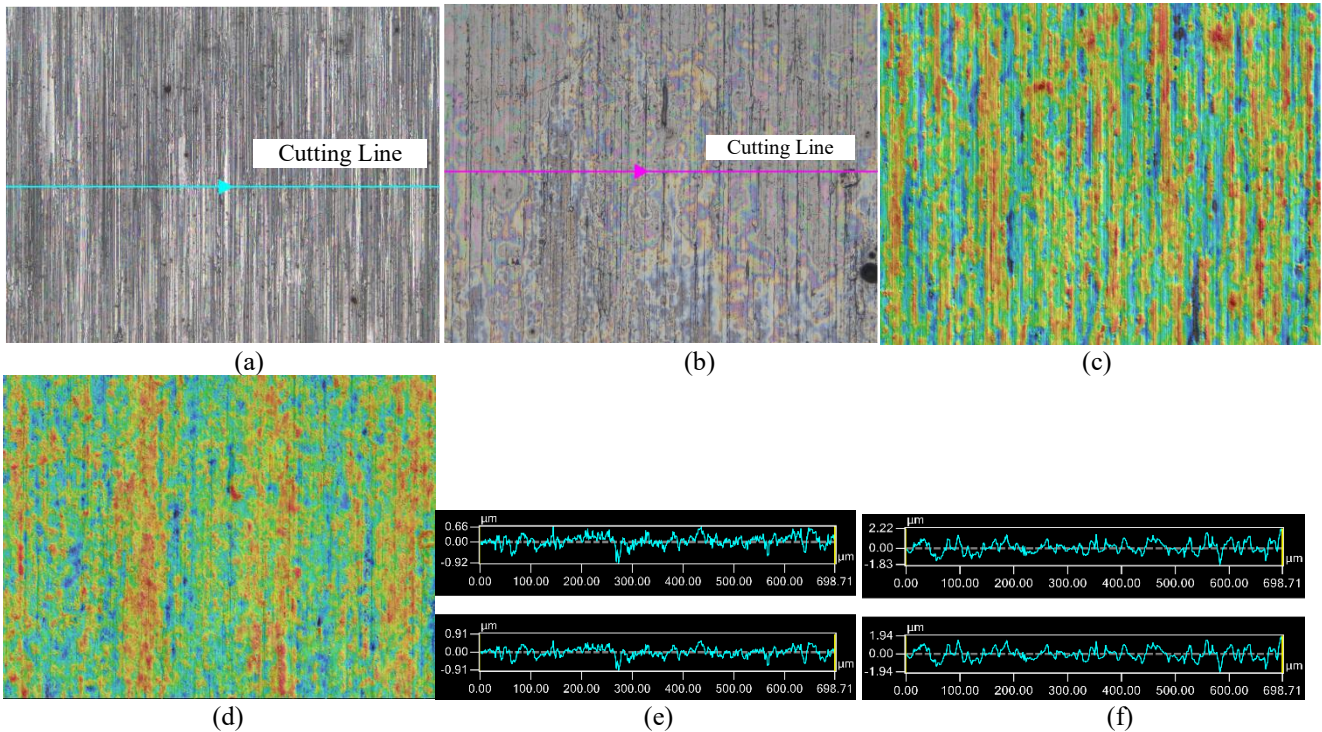


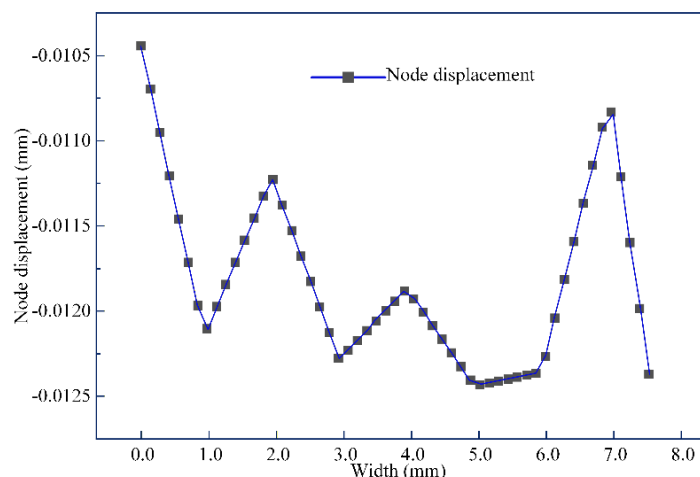
Figure 10 Graph showing changes in surface roughness (a) Surface roughness of the substrate (b) Surface roughness after ultrasonic impact (c) Surface profile diagram of the substrate (d) Surface profile diagram after ultrasonic impact (e) Two-dimensional contour curve of the substrate surface (f) Two-dimensional profile curves after ultrasonic impact

To quantify the surface morphology evolution, multiple rolling trajectories are generated along the circumferential nodal path, as shown in Fig. 11, and the surface nodal displacements in the normal direction are extracted, as depicted in Fig. 12. The arithmetic mean roughness (R_a) is then calculated from the absolute deviations between the displacement of each node and the mean displacement, following the standardized definition of R_a , i.e., the arithmetic average of the absolute distances between the surface profile points and the mean line, as expressed in Eq. (12).

$$Ra = 1/l \sum_{i=1}^l |X_i - \bar{X}| \quad (12)$$



where X_i represents the displacement of each node along the Z-axis, $\bar{X}$ represents the average value of the Z-axis displacement of each node in the circumferential direction; i is the number of nodes in the circumferential direction.



390 **Figure 11 Schematic diagram of node displacements**

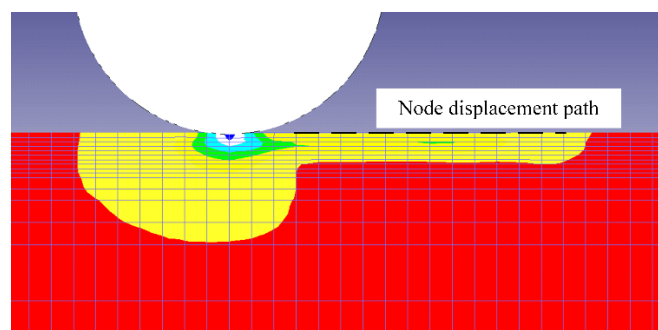


Figure 12 Selection of node displacement paths

5.2.2 The effect of heat exposure on surface roughness

Fig. 13 illustrates the evolution of surface roughness (R_a) after different durations of thermal exposure. After 1, 2, 4, and 8 h of exposure at 65 °C, the R_a values are 0.1854, 0.1869, 0.2124, and 0.1619 μm , respectively. Compared with the as-impacted condition ($R_a = 0.18 \mu\text{m}$), the roughness variation at each time point remained limited, fluctuating within a narrow band around the initial value.

The surface roughness exhibited a non-monotonic variation during thermal exposure: the R_a value rebounded slightly at 4 h and then decreased to 0.1619 μm at 8 h. This fluctuation may be associated with the localized release and redistribution of microstrain in the surface layer during thermal exposure. In the early stage of exposure (1–2 h), the compressive residual stress introduced by ultrasonic impact relaxed rapidly, accompanied by a partial release of microstrain; consequently, the microscopic surface asperities rose slightly due to elastic recovery, leading to a modest increase in roughness. As the exposure duration is extended to 8 h, stress relaxation approached saturation, and the surface material re-established a balanced stress state through localized dislocation rearrangement, causing the surface topography to stabilize accordingly.

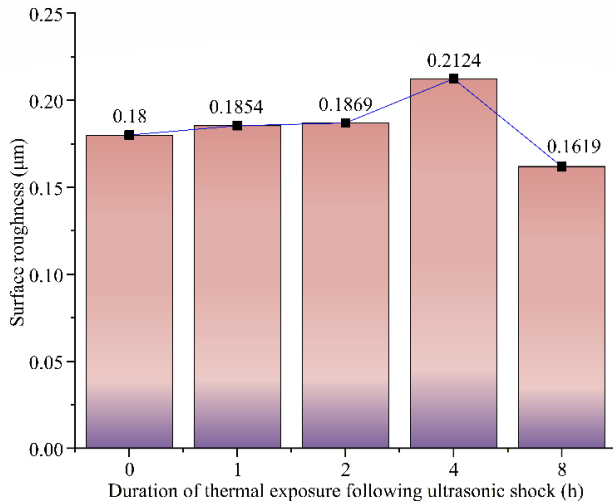


Figure 13 Trends in surface roughness after different durations of heat exposure

Although the Ra value fluctuated at 4 h, the overall variation remained small; in fact, the roughness after 8 h (0.1619 μm) is even lower than that in the as-impacted condition, indicating that thermal exposure at 65 °C do not compromise the surface smoothing effect achieved by ultrasonic impact.

5.3 Hardness and microstructure

5.3.1 Hardness gradient distribution after ultrasonic impact

Microhardness reflects the degree of work hardening in the surface layer and serves as another important indicator for evaluating the effectiveness of ultrasonic impact strengthening. The high-strain-rate plastic deformation induced by ultrasonic impact causes a sharp increase in dislocation density within the surface layer, resulting in pronounced grain refinement and work hardening. Unlike residual stress, hardness characterizes the material's resistance to localized plastic deformation and thus exhibits a distinct response to temperature. By comparing the depth-dependent hardness distributions before and after thermal exposure, it can be determined whether the temperature rise to 65 °C degrades the work hardening effect, thereby providing cross-validation with the residual stress evolution.

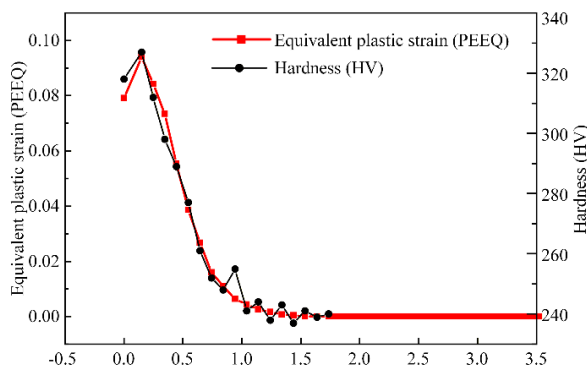


Figure 14 PEEQ and hardness along the depth direction



Fig. 14 presents the microhardness distribution along the depth direction on the cross-section of the specimen after UIT. The untreated specimen exhibited a core hardness of 231 HV and a surface hardness of 240 HV, with no discernible hardened layer. After UIT, the surface hardness increased significantly to 318 HV, representing a 32.5% improvement over the untreated specimen. With increasing distance from the surface, the hardness decreased gradually and returned to the core level at a depth of approximately 1.14 mm, indicating that the hardened layer introduced by ultrasonic impact extends to approximately 1.14 mm, essentially consistent with the depth of the compressive residual stress layer.

The depth-dependent microhardness distribution agrees closely with the simulated PEEQ (equivalent plastic strain) curve: both decrease with increasing depth and eventually level off. This parallel trend confirms that the PEEQ value directly reflects the work hardening intensity induced by ultrasonic impact and provides a basis for further analysis (Zheng et al., 2026). Based on the hardness distribution characteristics, the increase in surface hardness can be primarily attributed to the severe cumulative plastic strain and the associated work hardening introduced by ultrasonic impact. The hardened layer depth is essentially consistent with the compressive residual stress layer depth (Section 5.1.1), and the two results are mutually corroborative.

5.3.2 Hardness evolution following heat exposure

Fig. 15 presents the equivalent plastic strain (PEEQ) distributions across the specimen cross-section as a function of depth after different durations of thermal exposure. As demonstrated by the comparison with Fig. 14, the hardness variation trends are essentially consistent with those of the equivalent plastic strain. Based on Fig. 15, the strain distributions after 1, 2, 4, and 8 h of exposure at 65 °C essentially overlap with that of the as-impacted condition, with the maximum deviation in the surface region not exceeding 5%; correspondingly, neither the peak hardness nor the hardened layer depth exhibits any significant change. These results demonstrate that the work hardening effect introduced by ultrasonic impact does not degrade significantly upon temperature elevation to 65 °C.

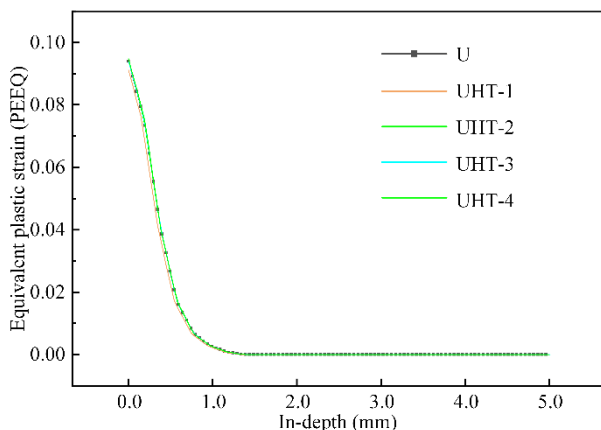


Figure 15 Equivalent plastic strain plots following thermal exposure at 65°C for different durations

The divergent responses of hardness and residual stress to temperature can be attributed to their distinct physical origins: residual stress arises from the accumulation of elastic lattice distortion and is therefore highly sensitive to thermally activated dislocation motion, whereas hardness is primarily governed by grain size and dislocation configuration. Since 65 °C remains



below the threshold temperature for appreciable recovery or recrystallization of 30CrMoA steel, the hardness is essentially preserved. These findings indicate that, although the temperature rise to 65 °C induced partial relaxation of the compressive residual stress, the work hardening of the surface layer retained good thermal stability.

5.3.3 Microstructural analysis

Fig. 16a presents the cross-sectional metallographic microstructure of the surface layer of the 30CrMoA specimen after UIT. The microstructure exhibits a distinct gradient distribution from the impacted surface (top of the image) toward the interior. Within the outermost layer, extending to a depth of approximately 44.61 μm , the original grain morphology is no longer discernible: the microstructure has been severely fragmented and elongated, forming a plastic flow layer aligned along the surface direction, with blurred grain boundaries, features characteristic of severe plastic deformation. This can be attributed to the impact head striking the specimen surface at a high frequency of 25 kHz, which introduces extreme compressive strain into the surface material and causes the grains to fragment and elongate along the deformation direction. With increasing distance from the surface, the degree of deformation diminishes progressively; in the regions deeper than 44.61 μm , the original grain morphology is preserved with clearly distinguishable grain boundaries, and the microstructural transition is well defined.

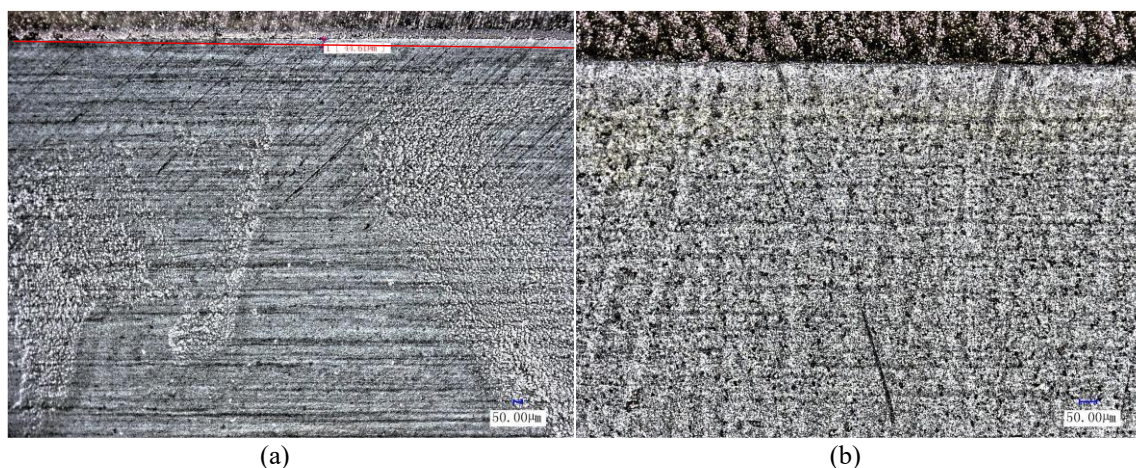


Figure 16 Metallographic microstructure (a) Microstructure after ultrasonic impact (b) Microstructure after 8 hours of heat exposure at 65°C

After 8 h of thermal exposure at 65 °C (Fig. 16b), the thickness of the surface plastic deformation layer showed no noticeable reduction compared with the as-impacted condition, remaining at approximately 50 μm . No nucleation or growth of recrystallized grains is observed, indicating that the temperature rise to 65 °C did not trigger significant microstructural recovery or recrystallization within the impact-affected layer. This observation is consistent with the microhardness results, no significant hardness decrease is detected before or after thermal exposure, further confirming that 65 °C remains below the threshold for appreciable thermal softening of 30CrMoA steel. From the perspective of residual stress evolution, given the absence of microstructural recovery, the stress relaxation observed in Section 5.1 can be primarily attributed to localized stress release driven by thermally activated dislocation motion, rather than to microstructural changes such as grain boundary migration or recrystallization.



6 Conclusion

- (1) Ultrasonic impact treatment introduced a residual compressive stress field with typical gradient characteristics into the surface layer of the 30CrMoA drive shaft. The surface compressive residual stress reached approximately -321 MPa, with a peak value of -723.74 MPa at a depth of about $375\text{ }\mu\text{m}$, and the compressive stress layer extended to a depth of approximately 1.253 mm. Meanwhile, the surface roughness (R_a) decreased from $0.8\text{ }\mu\text{m}$ in the as-ground condition to $0.18\text{ }\mu\text{m}$, demonstrating a significant improvement in surface quality.
- (2) At $65\text{ }^{\circ}\text{C}$, corresponding to the service temperature of the gear shaft, the compressive residual stress exhibited a relaxation behavior characterized by an initial rapid decay followed by a gradual stabilization with increasing exposure duration; the stress stabilized after 8 h, with a total relaxation rate of approximately 23.94%, indicating that the compressive residual stress introduced by ultrasonic impact was only partially released under these thermal conditions.
- (3) In contrast to the relaxation of residual stress, the surface microhardness, hardened layer depth, and microstructure remained essentially unchanged after thermal exposure at $65\text{ }^{\circ}\text{C}$. The surface hardness was maintained at approximately 318 HV, the hardened layer depth at about 1.14 mm, and the thickness of the surface plastic deformation layer at approximately $45\text{ }\mu\text{m}$, with no recrystallization or grain growth observed. This differentiated response demonstrates that $65\text{ }^{\circ}\text{C}$ remains below the threshold temperature for thermal recovery or recrystallization of 30CrMoA steel, and that the work hardening effect possesses good thermal stability.
- (4) The residual stress distributions and evolution trends predicted by the finite element simulation showed good agreement with the experimental results, confirming the validity of the established model. The residual stress relaxation behavior and the thermal stability of work hardening revealed in this study provide a theoretical basis for the process design and engineering application of ultrasonic impact treatment on gear shafts operating under elevated-temperature service conditions.

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