

Review

Evolution Mechanisms of Diffusion-Induced Phase Transformation Layers in Gun-Barrel Bores Under Thermochemical Coupling

Jinghua Cao ¹, Yiming Liu ¹, Mengran Zhu ¹, Jiawei Fu ^{2,*}, Yao Jiang ^{1,*}, Zheng Li ¹, Ying Liu ¹ and Jingtao Wang ¹

¹ School of Materials Science and Engineering, Nanjing University of Science and Technology, Nanjing 210094, China; caojinghua@njust.edu.cn (J.C.)

² School of Mechanical Engineering, Nanjing University of Science and Technology, Nanjing 210094, China

* Correspondence: jwfu@njust.edu.cn (J.F.); jiangyao@njust.edu.cn (Y.J.)

Abstract

This study focuses on a 155 mm 32CrNi3MoV steel barrel and presents a thermochemically coupled phase transformation and diffusion dynamics model. The model leverages the significant disparity between radial and axial temperature gradients to simplify the heat conduction problem to a one-dimensional transient formulation. The temperature field distribution during firing sequences is solved analytically, accounting for the dynamic shift in critical phase transformation temperatures under high heating rates. The evolution of the martensitic layer thickness under repeated thermal shock is subsequently calculated. A numerical model for the pulsed diffusion of C and N is established based on Fick's second law, incorporating the competitive diffusion–phase transformation mechanisms that govern martensite/austenite interface migration. To quantitatively evaluate the synergistic contribution of C and N to austenite stabilization, a carbon equivalent (Ceq) model is introduced, with the weight coefficient of N relative to C determined to be 0.68 and the critical Ceq required to lower the martensite start temperature below 25 °C calculated as 1.15 wt%. Concurrently, the microstructure and elemental distribution within the austenite layer of the retired barrel are systematically characterized using multi-scale techniques. The results indicate that the austenite layer on the inner bore surface arises from the synergistic effects of cyclic thermal-shock-induced phase transformation and elemental diffusion. Based on the Ceq criterion, the austenite layer thickness increases rapidly during the initial ~100 firing cycles, after which the growth rate slows significantly: it reaches approximately 1.27 μm after the first cycle and 2.94 μm after 1000 cycles, with only 0.2 μm of additional thickening between 100 and 1000 cycles—consistent with the experimentally observed range of 1.52–4.16 μm. The martensitic layer formed during the first firing cycle exhibits low thermal conductivity, which impedes subsequent heat transfer and leads to stabilization of its thickness at a characteristic depth. Grain refinement induced by repeated thermal shock provide short-circuit diffusion paths for elemental diffusion, accelerating compositional homogenization within the austenite layer and resulting in a stepped concentration profile at the interface. This study provides a representative example of non-equilibrium coupled phase transformation–diffusion phenomena under extreme transient loading. The established thickness prediction model can provide guidance for service life assessment of large-caliber barrels, offering both theoretical foundations and practical engineering guidance for their material design and performance optimization.



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Keywords: phase transformation–diffusion model; barrel erosion; gun barrel; multi-field coupling; austenite layer

1. Introduction

Elevated chamber pressure, high muzzle velocity, and high firing rate are the dominant development trends of modern large-caliber artillery systems [1]. With the continuous advancement of related technologies, the gun bore is subjected to increasingly higher peak temperatures and transient pressures of propellant gas, which aggravate severe thermo–mechanical–chemical coupling effects. Such extreme service conditions pose a serious threat to barrel service life and the operational performance of artillery systems. Figure 1 presents a comprehensive thermo-mechanical-chemical ablation model for a typical gun barrel [2]. During ballistic firing, the inner bore surface undergoes ultra-fast thermal cycling on a millisecond timescale [3–5]. Meanwhile, chemically active atoms (e.g., C, N, O) from the high-pressure gas undergo transient interstitial diffusion [4], inducing violent interfacial reactions and non-equilibrium phase transformations. Consequently, a unique multi-layered radial gradient microstructure is formed [6]. As shown in Figure 2, this cross-sectional gradient structure is the key carrier of ablation failure mechanisms. Its local microstructural features and mechanical properties directly determine the anti-ablation performance and structural integrity of the barrel.

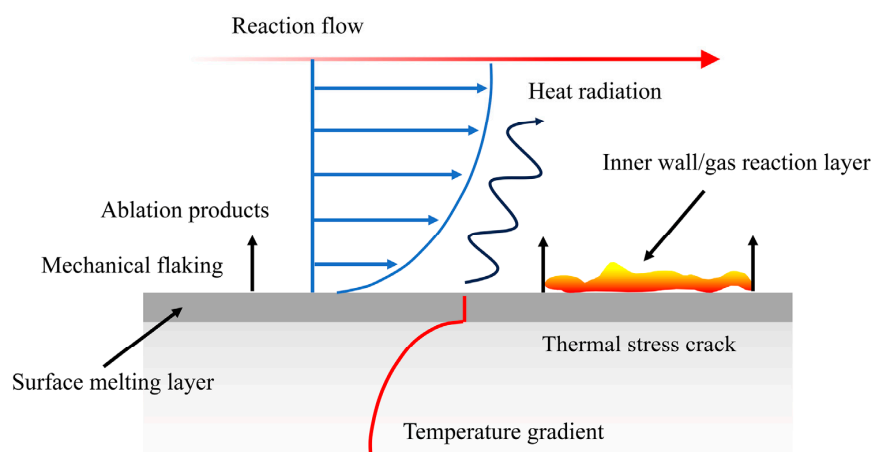


Figure 1. General schematic of the thermal–chemical–mechanical erosion mechanisms (Reprinted from Ref. [2]).

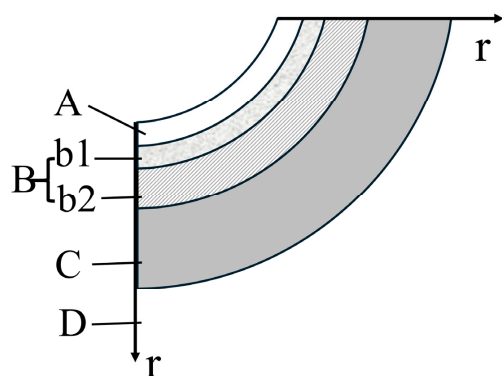


Figure 2. Schematic diagram of radial gradient organization of the barrel. (A) Outer white layer; (B) inner white layer, b1: diffusion layer, b2: phase transition layer; (C) heat-affected zone; (D) matrix.

Regarding the phase composition of the outer white layer, Ahmad [7] explicitly stated that this region comprises a mixture of carbides, nitrides, and minor fractions of oxides. Botstein et al. [8] further demonstrated that the carbides within the outer white layer do not exist as a generic Fe_3C phase, but rather consist of various discrete carbide species, such as Fe_2C , Fe_5C_2 , and Fe_{20}C_2 . Utilizing scanning electron microscopy combined with taper sectioning techniques, Turley et al. [9] confirmed that the outer white layer is predominantly composed of Fe_3C with a thickness of approximately $0.25\text{ }\mu\text{m}$. More recently, through simulated ablation experiments, Zhang [10,11] and Chen [12] et al. proposed that the outer white layer is a reaction product layer derived from thermochemical corrosion, primarily enriched with carbon, oxygen, and sulfur-bearing compounds. Although a general consensus has been reached regarding the phase constitution and formation mechanisms of this outer white layer—characterizing it as a low-strength, low-melting-point compound zone [8,13–16] that is easily exfoliated under high-velocity gas erosion and mechanical shearing [9]—the underlying inner white layer remains less understood. Consequently, the sub-surface inner white layer represents the immediate frontier of the interaction between the barrel substrate and the high-temperature, chemically active propellant gas. Its formation mechanism and localized thickness variations directly govern the phase stability, interfacial bonding strength, and crack-initiation behaviors of the material, thereby constituting a pivotal theoretical cornerstone for elucidating barrel ablation mechanisms and accurately predicting service life.

Unlike the outer white layer, the inner white layer is a metastable structure formed via transient thermochemical coupling within the barrel bore, and it remains a subject of intense academic debate. Currently, this controversy centers around two competing schools of thought: the austenitic viewpoint and the martensitic/multiphasic viewpoint. Proponents of the former assert that the inner white layer is predominantly composed of austenite; for instance, Ahmad [7] suggested that this layer consists of an austenitic matrix stabilized by the dissolution of substantial fractions of carbon (C) and nitrogen (N), while Turley et al. [9] utilized X-ray diffraction (XRD) to confirm that high-carbon austenite is the primary constituent. Conversely, an alternative school of thought advocates for predominantly martensitic or multiphasic architectures. Gong [17] proposed, based on metallographic evidence, that the white layer comprises a mixture of martensite and retained austenite, a conclusion echoed by Zhang et al. [10,11], who characterized the inner white layer as a refined multiphasic mixture of these two constituents. Furthermore, Yu [18] dissected a 14.5 mm chromium-coated machine gun barrel and revealed that the white layer at the forcing cone region consisted of a mixture of high-carbon austenite and cementite. To elucidate the dynamic evolution of these multi-layered architectures, researchers have widely implemented laser-pulse simulations [19–21] and semi-closed explosive test configurations [22,23]. However, conventional laboratory environments fall short of reproducing the cumulative thermal profiles and pulsed interstitial diffusion characteristics of C and N atoms that occur during sustained, multi-shot ballistic sequences under operational conditions. Consequently, the distinct austenite layer has rarely been captured in simulation-based trials, leaving its precise formation and evolutionary mechanisms unresolved. At present, direct experimental observations regarding the formation of the austenite layer on actual gun bore surfaces remain exceptionally scarce, and to date, a quantitative, multi-physics-coupled dynamic predictive model for this specific layer remains absent from the open literature.

Our preliminary research [24] demonstrated that a uniform austenite layer, approximately $2\text{--}3\text{ }\mu\text{m}$ in thickness, indeed forms at the rifling start of decommissioned large-caliber barrels. However, owing to the severe mechanical wear induced by rotating band engraving and projectile scouring within the rifled zone, this fragile austenite layer is highly

susceptible to physical destruction, making its complete characterization exceptionally difficult. Conversely, the gun barrel chamber sustains primarily the transient thermal shocks and chemical reactions induced by the ultra-high-pressure propellant gas, remaining virtually free from the direct mechanical abrasion caused by projectile-bore interactions. This isolation provides a decoupled, pristine platform for comprehensively investigating the standalone spatiotemporal evolution of the austenite layer under dual thermochemical coupling fields. Extensive literature has established that among the numerous degradation mechanisms of gun bore surfaces, the thermal shock from high-temperature propellant gas is the dominant cause of barrel ablation [15,25–27], while pure mechanical erosion plays a subordinate role. In view of this, the present study focuses on the chamber of a 155 mm large-caliber 32CrNi3MoV steel barrel, targeting the core multi-physics effects of thermochemical interactions while successfully eliminating the interference of mechanical exfoliation. A coupled thermochemical phase-transformation and diffusion-kinetics model was constructed to achieve quantitative predictions of both the transient formation and the long-term evolution of the austenite layer during cyclic ballistic firing. To quantitatively evaluate the synergistic contribution of C and N to austenite stabilization, a carbon equivalent (C_{eq}) model is introduced, with the critical C_{eq} required to retain austenite at room temperature determined via thermodynamic calculations. This numerical framework was further validated through bidirectional correlation with multi-scale microstructural characterizations. Consequently, this work establishes an explicit quantitative relationship between the austenite layer thickness and the cumulative number of firing rounds. This bridges a critical gap in the quantitative numerical prediction of austenitic layer evolution in large-caliber barrels and provides a robust theoretical foundation for barrel life assessment and anti-ablation structural design.

2. Test Results

2.1. Materials and Methods

The material investigated in this study is 32CrNi3MoV steel, used to manufacture 155 mm large-caliber thick-walled gun barrels. Its chemical composition is provided in Table 1. To validate the proposed model, a retired 155 mm large-caliber barrel made of 32CrNi3MoV steel was examined. Samples were extracted from the chamber region for detailed microstructural analysis. First, a specimen with dimensions of $10 \times 40 \times 10 \text{ mm}^3$ was sectioned from the chamber region, which then underwent degreasing, acid pickling, activation, and water rinsing, followed by nickel electroplating to protect the pristine microstructure of the bore surface layer. This electroplated specimen was then sectioned radially to obtain a $10 \times 10 \times 2 \text{ mm}^3$ coupon. The exposed surface was subsequently ground, polished, and etched to produce a clean, smooth surface suitable for SEM and EBSD analysis. APT needle specimens were prepared from the inner bore surface using focused ion beam (FIB) milling. Quantitative elemental analysis at various radial positions on the inner bore surface was conducted by three-dimensional atom probe tomography using a LEAP 4000X Si instrument (CAMECA, Fitchburg, MA, USA).

Table 1. Chemical composition of 32CrNi3MoV Steel (wt%).

Element	C	Si	Mn	Cr	Cu	Ni	Mo	V	Fe
32CrNi3MoV	0.35	0.16	0.62	0.68	0.03	3.26	0.55	0.20	Bal.

The numerical simulations were performed using MATLAB R2024a. The one-dimensional unsteady heat conduction equation and Fick’s second law were discretized using the finite difference method (FDM). The spatial domain (0–500 μm) was discretized

into 2000 uniform nodes, corresponding to a spatial step size of $\Delta r = 0.25 \mu\text{m}$. The temporal discretization for temperature field calculation employed a fixed step size of $\Delta t = 0.5 \text{ ms}$ over a total duration of 500 ms per firing cycle. For the diffusion simulation, an adaptive time-stepping scheme was implemented, with the step size dynamically adjusted between $1 \times 10^{-9} \text{ s}$ and $1 \times 10^{-6} \text{ s}$ based on the Courant-Friedrichs-Lewy stability condition. The Crank-Nicolson scheme was adopted for the diffusion equation to ensure numerical stability. The simulation was iterated for up to 1000 firing cycles. Finally, verify the simulation results with the experimental results. All model parameters, including material properties, boundary conditions, and diffusion coefficients, are detailed in Section 3.

2.2. Results

Figure 3 presents the microstructural features of the inner bore surface layer of the barrel after service. As shown in Figure 3a, the inner bore surface layer exhibits a three-layered microstructure: an outermost corrosion-resistant layer approximately $1.52\text{--}4.16 \mu\text{m}$ thick, an intermediate needle-like layer with a thickness of about $13.30\text{--}16.82 \mu\text{m}$, and an inner fine equiaxed layer. In Figure 3a, the red dashed line denotes the interface between the nickel layer and the inner bore surface layer, while the white dashed line marks the boundary between the austenite layer and the martensite. The region outlined by the yellow dashed box in Figure 3a corresponds to the EBSD analysis area shown in Figure 3b. Figure 3b displays the EBSD phase map of the surface layer. The phase map reveals that the corrosion-resistant layer identified in Figure 3a corresponds to a face-centered cubic (FCC) austenite structure, with a measured thickness of $2.45 \pm 0.43 \mu\text{m}$. Adjacent to this austenite layer lies a region with a body-centered cubic (BCC) structure corresponding to $\alpha\text{-Fe}$, within which cementite (Fe_3C) particles are distributed both intragranularly and along grain boundaries. The corrosion-resistant layer shown in Figure 3a and the yellow region in the EBSD phase map of Figure 3b, collectively confirming the formation of a continuous austenite layer, $1.52\text{--}4.16 \mu\text{m}$ in thickness, on the inner bore surface of the large-caliber barrel after service.

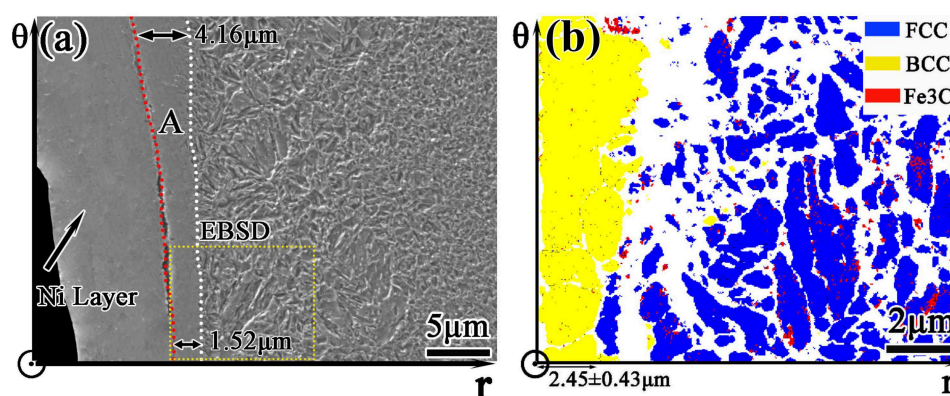


Figure 3. Microstructural characterization of the barrel surface layer after service: (a) SEM micrograph showing the overall morphology; (b) EBSD phase map revealing phase distribution. Note: z-axis: axial direction, with the positive direction pointing along the propellant gas flow; r-axis: radial direction; θ -axis: circumferential direction, with the positive direction defined as clockwise. The r- θ plane (i.e., the cross-section with z as the normal direction) defines the transverse section of the barrel.

Figure 4 presents a schematic illustration of the APT sampling locations. Prior to sampling, the thickness of the austenite layer at this location was measured to be $2.22 \pm 0.08 \mu\text{m}$. Accordingly, APT nanoneedles were prepared from two positions within the austenite layer: one at $r = 0 \mu\text{m}$ and another at $r = 2.20 \mu\text{m}$. Subsequently, an additional APT nanoneedle was prepared from a position located $2.20 \mu\text{m}$ from the right edge of the

austenite layer, corresponding to $r = 4.40 \mu\text{m}$. Table 2 lists the corresponding quantitative measurements obtained from the three needle tips. Table 2 indicates that, within the measurement uncertainty, the distributions of C and N in the austenite layer can be considered uniform, with no significant concentration gradient detected as a function of radial distance. Furthermore, nitrogen is detected exclusively within the austenite layer; it is absent from both the underlying matrix material and the adjacent martensite region.

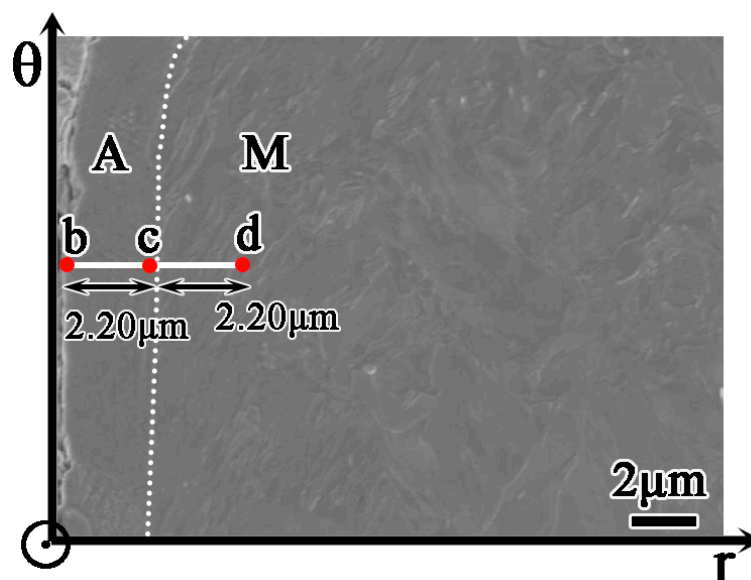


Figure 4. Schematic diagram of APT sampling location.

Table 2. Composition of the sample (at%).

	b-Tip	c-Tip	d-Tip
Fe	87.41	86.95	92.61
C	5.50	6.04	1.60
Si	0.45	0.40	0.34
Mn	0.68	0.81	0.58
Cr	1.12	1.05	0.70
Cu	0.07	0.09	0.06
Ni	4.00	3.86	3.67
Mo	0.08	0.12	0.24
V	0.19	0.19	0.12
N	0.23	0.22	-
O	0.25	0.26	0.09

3. Multi-Field Coupled Phase Transition Diffusion Model

During artillery firing, the inner bore surface is subjected to coupled thermal, chemical, and mechanical fields, encompassing high temperatures, reactive gases, and mechanical stresses. Under thermochemical exposure, the carbon and nitrogen potentials of the propellant gases must be considered, with temperature and exposure duration being the primary factors governing diffusion depth. Furthermore, Yaakoubi et al. [28] noted that stress and strain can influence phase transformation by altering the transformation temperatures. Given the absence of a unified theoretical framework describing the influence of stress state on phase transformation kinetics, the effect of mechanical stress was not considered in the present model. To validate the model, samples were extracted from the barrel chamber region, where thermochemical effects are dominant and mechanical contributions are min-

imal. A schematic representation of the coupled phenomena considered in the model is provided in Figure 5.

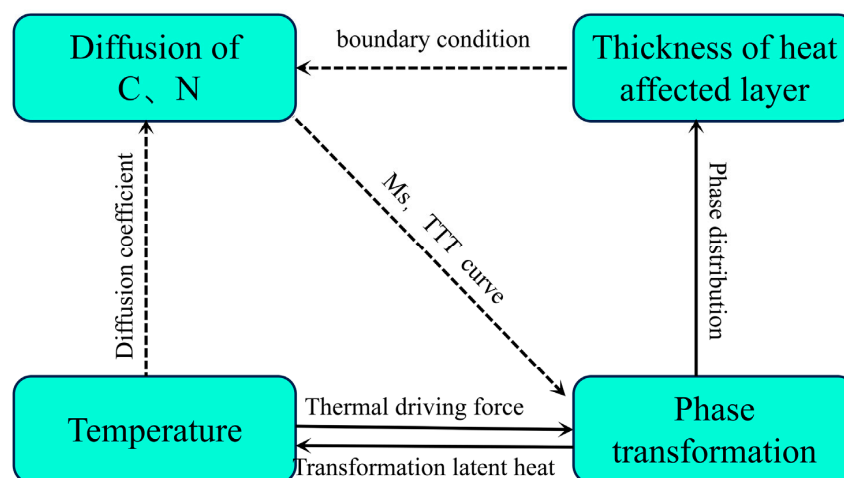


Figure 5. Multi-field coupled model.

3.1. One-Dimensional Radial Heat Conduction Model for Gun Barrels

Following a firing event, the heat transfer process within the barrel wall is inherently transient. Strictly speaking, heat conduction within the barrel wall constitutes a two-dimensional (axial and radial) transient problem. However, a full two-dimensional analysis is computationally demanding. Moreover, previous investigations have shown that, except for a discontinuity at the projectile location, the heat flux from the propellant gases to the barrel wall varies only weakly with axial position. Quantitative analyses [29] further indicate that the radial temperature gradient is typically more than three orders of magnitude larger than its axial counterpart, i.e., $\frac{\partial T_s}{\partial r} \gg \frac{\partial T_s}{\partial x}$, and similarly $\frac{\partial^2 T_s}{\partial r^2} \gg \frac{\partial^2 T_s}{\partial x^2}$. Consequently, axial heat conduction can be justifiably neglected. Additionally, assuming axial and circumferential symmetry of the temperature field is reasonable, given the geometry of the barrel. With these simplifications, the heat conduction within the barrel wall reduces to a one-dimensional transient problem along the radial direction. At the outer surface, heat is dissipated to the surroundings via natural convection, driven by the temperature difference between the barrel wall and the ambient environment. A schematic representation of the temperature field within the barrel wall is provided in Figure 6.

Basic assumptions:

- (1) The initial temperature of the barrel is uniform and equal to the ambient temperature.
- (2) Frictional and thermal effects due to projectile passage are neglected.
- (3) The temperature field is axially and circumferentially symmetric.
- (4) Heat loss from the outer surface occurs via natural convection and radiation to the environment.
- (5) Axial heat conduction within the barrel wall is negligible.
- (6) The barrel material is assumed to have constant thermophysical properties; i.e., specific heat capacity, thermal conductivity, and density are independent of temperature.

Consider the one-dimensional heat conduction problem in the radius r direction of the barrel; Take the difference between the inner surface temperature and the base temperature of the barrel as $T_s(r,t) - T_f = T(r,t) = u(0,t)$, The barrel $T_f = 25^\circ\text{C}$, and $u(0,t)$ is the surface temperature measurement curve of the barrel in reference [30]. The temperature boundary conditions are fitted based on the measured temperature curve in the reference:

$$\begin{cases} T(0, t) = \begin{cases} T(0, t)_1 = 138.3836 \ln t + 1667.97 & t \leq t_{\max} = 5 \text{ ms} \\ T(0, t)_2 = 264.6374 t^{-0.3} - 268.021 & t > t_{\max} = 5 \text{ ms} \end{cases} \\ T(\infty, t) = 0 \end{cases} \quad (1)$$

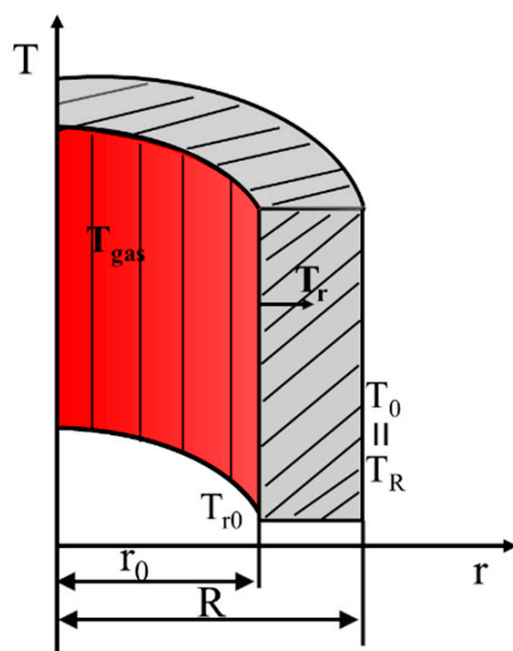


Figure 6. Temperature field within the barrel wall.

The one-dimensional transient heat conduction equation in cylindrical coordinates is given by

$$\partial T / \partial t = \alpha (\partial^2 T) / (\partial r^2) \quad (r_0 < r < R) \quad (2)$$

where T —temperature of the barrel wall ($^{\circ}\text{C}$); t —time (s); r —radial coordinate measured from the bore surface (m); r_0 —inner radius of the barrel (m); R —outer radius of the barrel (m); α —thermal diffusivity of the barrel material ($\text{m}^2 \cdot \text{s}^{-1}$). Reference is made to the properties of the similar gun steel PCrNi3MoVA specified.

Analysis of the heat conduction equation yields the following conclusions:

$$T(r, t) = A + \text{Berf}\left(\frac{xr}{\sqrt{4\alpha t}}\right) \quad (3)$$

Applying boundary condition (1) to Equation (3) gives: $A = T(0, t)$.

Applying boundary condition (2) to Equation (3) yields: $B = -A = -T(0, t)$.

The resulting temperature distribution is thus given by

$$T(r, t) = T(0, t) \left[1 - \text{erf}\left(\frac{xr}{\sqrt{4\alpha t}}\right) \right] \quad (4)$$

The explicit form of the temperature distribution is, therefore,

$$T_s(r, t) = T_f + T(r, t) = T_f + T(0, t) \left[1 - \text{erf}\left(\frac{xr}{\sqrt{4\alpha t}}\right) \right] \quad (5)$$

Figure 7 presents the simulated temperature field within the barrel during a firing cycle. As shown in Figure 7, the inner bore surface temperature rises instantaneously to 960°C under the action of the propellant gases during a single firing cycle. The radial temperature distribution at the instant of peak wall temperature is shown in Figure 7c. At

this instant, the heat absorbed from the propellant gases is concentrated near the inner bore surface (Figure 7a), while the cooling rate decreases progressively with increasing radial distance (Figure 7d). However, owing to the thermal resistance of the barrel material, radial heat conduction is impeded, leading to a steep temperature gradient. At a radial distance of approximately 500 μm from the bore surface, the temperature drops to only 180 $^{\circ}\text{C}$.

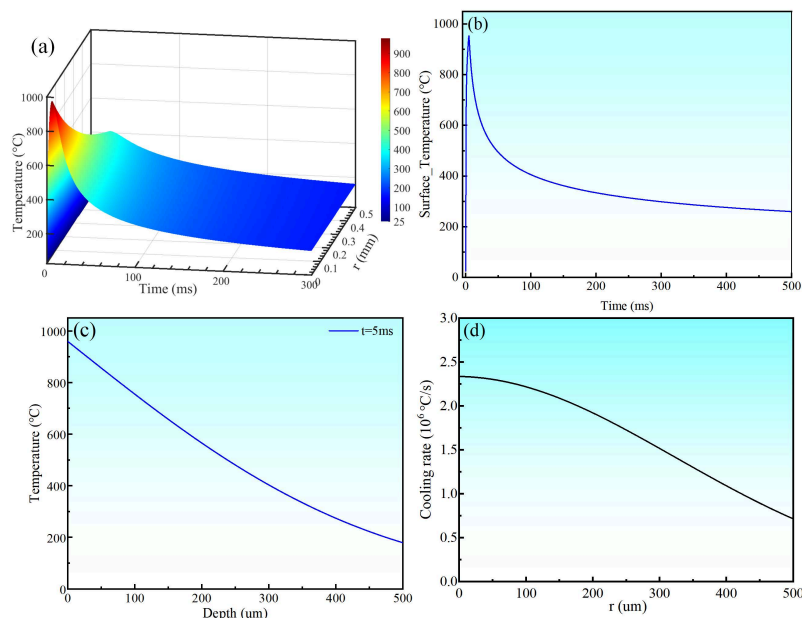


Figure 7. Temperature distribution diagram: (a) barrel temperature time radial depth curve during single shot; (b) the time temperature curve of the surface layer of the inner chamber during a single shot; (c) the radial temperature variation in the surface layer of the bore during single shot ($t = 5$ ms); (d) radial distribution of cooling rate ($t = 5$ ms).

3.2. Modeling the Thickness of the Martensitic Layer Formed Under Thermal Shock

The depth of the self-quenched martensitic layer formed in the barrel during service can be predicted by combining the heating-rate-dependent critical transformation temperatures of the matrix, the estimated thermophysical properties of the martensitic phase, and the solution of the heat-conduction equation. According to reference [31], the thermophysical properties were assumed to be a thermal conductivity of 40 W/(m.K) and a specific heat capacity of 480 J/(kg.K). The density of the martensitic layer was assumed to be identical to that of the original matrix structure.

From the computed temperature field, the heating and cooling rates at various radial positions within the barrel can be derived as follows:

$$T'(r, t) = T'(0, t) \left[1 - \operatorname{erf} \left(\frac{r}{\sqrt{4\alpha t}} \right) \right] - \frac{T(0, t)}{\sqrt{\pi\alpha t}} \exp \left(-\frac{r^2}{4\alpha t} \right) \quad (6)$$

When the temperature at a given radial position reaches the A_{c3} temperature corresponding to the local heating rate, that region will undergo self-quenching upon cooling, leading to the formation of martensite under the prevailing thermal shock conditions. Continuous cooling transformation (CCT) diagrams for 32CrNi3MoV steel were calculated using the JMatPro Version 7.0.0 thermodynamic software (with the composition given in Table 1), and the results are presented in Figure 8. The calculated CCT diagram indicates that at a cooling rate of 100 $^{\circ}\text{C}\cdot\text{s}^{-1}$, the undercooled austenite transforms to martensite when the temperature falls below the martensite start temperature, $M_s = 313$ $^{\circ}\text{C}$.

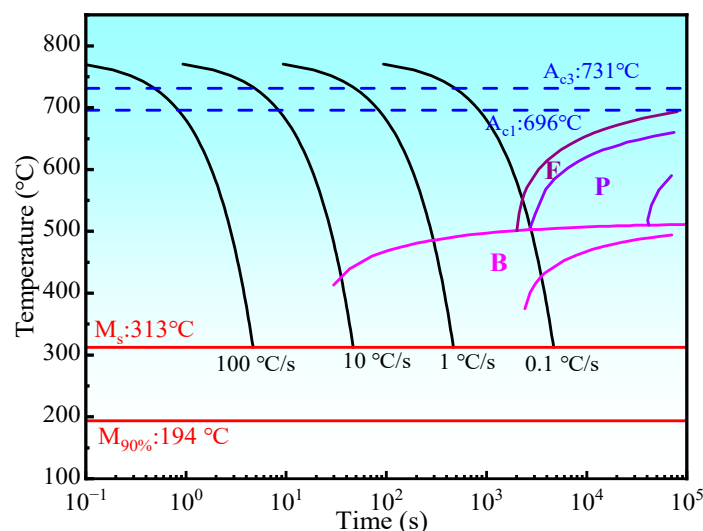


Figure 8. Continuous Cooling Transformation (CCT) diagram of 32CrNi3MoV gun barrel steel.

The simulated temperature field presented in Figure 7 indicates that, during service, the heating rate on the inner bore surface can reach up to $10^4 \text{ }^\circ\text{C}\cdot\text{s}^{-1}$. At such an extreme heating rate, the inner bore surface temperature rises rapidly to $960 \text{ }^\circ\text{C}$ within milliseconds, followed by an instantaneous cooling rate as high as $10^5 \text{ }^\circ\text{C}\cdot\text{s}^{-1}$. Previous studies [32–35] have demonstrated that a substantial increase in heating rate can alter the phase transformation temperatures, thereby influencing the transformation kinetics. The effect of heating rate on the phase transformation temperatures, as reported in the literature [35], is illustrated in Figure 9. By integrating the heating-rate dependence of the A_{c3} and A_{c1} temperatures shown in Figure 9, the thermodynamic calculations in Figure 8, and the data in Figure 7, the effective transformation temperatures under service conditions can be estimated. For the high heating rates characteristic of barrel firing, the dynamic A_{c3}' temperature of the matrix material is approximately $831 \text{ }^\circ\text{C}$, and the A_{c1}' temperature is approximately $760 \text{ }^\circ\text{C}$.

$$831 - T_f = T(0, t) \left[1 - \operatorname{erf} \left(\frac{r}{\sqrt{4at}} \right) \right] \quad (7)$$

By finding the extremum of t with respect to r , the maximum depth at which martensite can form, denoted as $M1r_{\max}$, is obtained.

If the base material has already been transformed to martensite, no elemental diffusion is required during subsequent thermal shock cycles, and austenitization occurs more readily. The critical temperature for martensite formation in this region is then the dynamic A_{c1}' temperature, approximately $760 \text{ }^\circ\text{C}$. The corresponding maximum depth at which martensite can form, denoted as $Mr_{\max}$, is obtained following the same procedure used for $M1r_{\max}$.

$$760 - T_f = T(0, t) \left[1 - \operatorname{erf} \left(\frac{r}{\sqrt{4at}} \right) \right] \quad (8)$$

By finding the extremum of t with respect to r , the maximum depth at which martensite can form, denoted as $Mr_{\max}$, is obtained.

However, once the matrix has been partially transformed to martensite, the underlying untransformed matrix becomes covered by a surface martensitic layer. Consequently, during subsequent thermal shocks, heat must first penetrate this martensitic layer before reaching the underlying matrix. Therefore, for firing cycles subsequent to the initial transformation, the heat-conduction problem within the barrel wall must be treated using a two-layer composite material model.

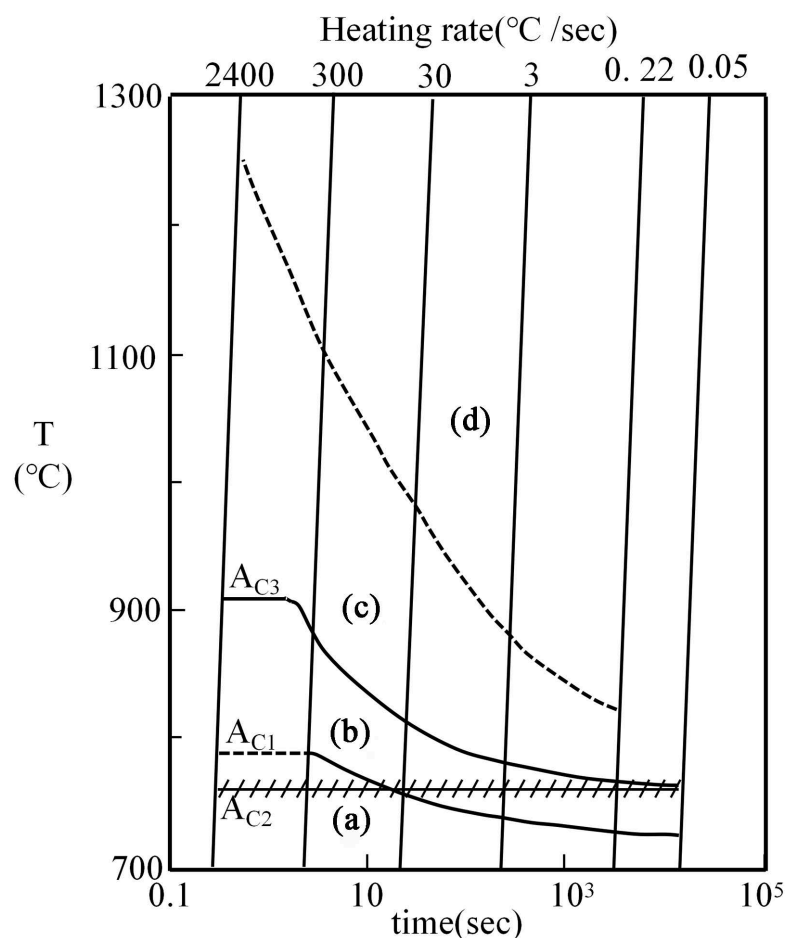


Figure 9. Time–Temperature–Austenitizing (T.T.A.) curve for C45 normalized steel (Reprinted with permission from ref. [35]. 1984 Springer): (a) ferrite and pearlite; (b) ferrite, pearlite, and austenite; (c) inhomogeneous austenite; (d) homogeneous austenite.

3.3. Diffusion Model Formulation

In modeling the thickness of the austenite layer, the first step is to establish the conditions under which austenite can form and be retained at room temperature. Previous analysis of the austenite formation mechanism has shown that the ingress of carbon and nitrogen atoms significantly depresses the martensite start (M_S) temperature of the surface layer to below room temperature [24]. This stabilization effect allows metastable austenite to be retained at ambient temperature. Accordingly, it is assumed that the ingress of C and N lowers the local M_S temperature to approximately room temperature (25 °C), thereby enabling the retention of metastable austenite under ambient conditions. Based on the alloy composition of the austenite layer and the data in Table 2, the average ratio of infiltrated C to N was determined to be $\Delta C:\Delta N = 15.8:1.0$. Using JMatPro software, the critical ingress levels of C and N required to stabilize austenite at room temperature were calculated, starting from the baseline C and N concentrations of the matrix material. By progressively increasing the C and N concentrations until the calculated M_S temperature fell to 25 °C, the critical ingress levels were determined. According to calculations, when the carbon content is 1.05 wt% and 0.044 wt%, the M_S approaches room temperature.

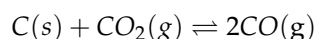
During a single firing event, the inner bore surface temperature rises instantaneously to 960 °C under the action of the propellant gases, while the chamber pressure in conventional gun barrels can reach 250–350 MPa [3,36,37]. Accordingly, using the gas composition reported in the literature [38], whose typical composition is given in Table 3, the theoretical

carbon and nitrogen potentials of the propellant atmosphere at 300 MPa were calculated, corresponding to the peak surface temperature attained during combustion.

Table 3. Typical composition of propellant gas.

Gas	CO	CO ₂	H ₂	H ₂ O	N ₂
Content, wt%	38.03	9.13	10.72	20.85	21.27

Carbon potential is conventionally defined as the α C in austenite that would be in equilibrium with a given gas atmosphere (with graphite taken as the standard state, $\alpha_C = 1$). This is based on the Boudouard reaction [39]:



The equilibrium constant K_P for this reaction is given by the following empirical expression [40]:

$$\ln K_P = -\frac{\Delta G^0}{RT} \quad (9)$$

At 960 °C, $K_P \approx 47.85$ was calculated.

At a pressure of 300 MPa, the gas behavior deviates significantly from ideal behavior. To account for this non-ideality, the fugacity must be calculated. Using the Peng–Robinson (PR) equation of state [41] for theoretical estimation, with parameters $f_i = y_i \cdot P \cdot \phi_i$, the reduced pressure and temperature for CO₂ are $P_r \approx 85.7$ MPa and $T_r \approx 9.2$, respectively; for CO, $P_r \approx 40.6$ MPa and $T_r \approx 4.0$.

According to generalized compressibility factor charts for high-pressure conditions [42], at such elevated temperatures and pressures, repulsive intermolecular forces dominate, and the fugacity coefficient ϕ is typically significantly greater than unity. Based on these correlations, the estimated fugacity coefficients are $\phi_{CO_2} \approx 2.45$ and $\phi_{CO} \approx 2.10$.

According to chemical equilibrium [43],

$$\alpha_C = \frac{1}{K_P} \cdot \frac{f_{CO}^2}{f_{CO_2} \cdot P^0} \quad (10)$$

Calculated $\alpha_C \approx 168.4$.

The relationship between the activity of carbon and its mole fraction in austenite is given by

$$\ln a_C = \ln \left(\frac{x_C}{1 - 5x_C} \right) + \frac{5115.9}{T} - 1.954 \quad (11)$$

Here, α_C denotes the activity of carbon, with graphite taken as the standard state. Using the calculated carbon activity $\alpha_C \approx 168.4$ derived from the gas atmosphere, the corresponding mole fraction x_C was found to be 0.1978. Converting this to mass percent yields a carbon concentration of $C = 5.04$ wt%. Similarly, the theoretical nitrogen potential of the gas atmosphere was determined to be 0.60 wt%.

According to the Fe–C phase diagram, the maximum solubility of carbon in austenite at 960 °C is approximately 1.5 wt% [44]. The carbon potential of the propellant gas under firing conditions far exceeds this solubility limit. Temperature, alloying elements, and pressure are known to influence the solubility of carbon in steel [45–47]. Therefore, for the specific conditions of this study—960 °C and 300 MPa—it is necessary to calculate the maximum solubility of carbon in austenite for 32CrNi3MoV steel.

To account for the effect of pressure on the maximum solubility of carbon, we begin with fundamental thermodynamic principles: the influence of pressure P on the chemical

potential of a component is governed by its partial molar volume. For equilibria involving austenite and cementite or graphite, an increase in pressure induces a shift in the phase boundary. This shift can be described by the Clausius–Clapeyron equation [39]:

$$x_C(P) = x_C(P_0) \cdot \exp\left(-\frac{\Delta V \cdot (P - P_0)}{RT}\right) \quad (12)$$

Here, ΔV represents the molar volume change associated with the reaction. For the dissolution of carbon into austenite, the insertion of carbon atoms into interstitial sites causes lattice expansion, resulting in a positive ΔV . According to the thermodynamic relation, an increase in pressure reduces the exponential term, thereby suppressing carbon dissolution in austenite and decreasing the maximum solubility with rising pressure. P_0 is the standard atmospheric pressure (0.1 MPa), and P denotes the applied high pressure. R is the gas constant ($8.314 \text{ J} \cdot \text{mol}^{-1} \cdot \text{K}^{-1}$), and T is the absolute temperature (K).

According to the formula, when the pressure is 300 MPa, $\Delta C_{\text{pressure}} = -0.063 \text{ wt}\%$.

The influence of alloying elements on the maximum solubility of carbon was evaluated based on the alloy composition given in Table 1, following Gunnarson's empirical formula and thermodynamic principles [39], the saturated carbon content $C_{\text{max},\gamma}$ in alloy steel can be obtained by applying corrections to the baseline solubility C_0 ($C_0 = 1.50 \text{ wt}\%$) of pure iron: $C_{\text{max},\gamma}^{\text{total}} = C_{\text{max},\gamma}(0.1 \text{ MPa}) + \Delta C_{\text{pressure}} = 1.39 \text{ wt}\%$.

According to the calculations presented above, the maximum theoretical solubility of carbon in austenite for 32CrNi3MoV steel under firing conditions is approximately 1.39 wt%. Similarly, the corresponding maximum theoretical solubility of nitrogen is approximately 0.06 wt%. During firing, the maximum solubilities of carbon and nitrogen in austenite are considerably lower than the carbon and nitrogen potentials of the propellant gas. Consequently, the carbon concentration in the inner bore surface layer cannot exceed the maximum solid solubility of carbon under the prevailing conditions. The APT results presented in Table 2 indicate that the carbon concentration in the austenite layer approaches this theoretical maximum solubility, a factor that must be accounted for in subsequent modeling.

During service, propellant combustion within the barrel generates a gas atmosphere rich in carbon-bearing (e.g., C_C) and nitrogen-bearing (e.g., C_N) species. Under the influence of high temperature, this atmosphere promotes carburizing and nitriding along the radial direction. Upon gas evacuation, the high-concentration sources (C_C and C_N) are rapidly removed, and the barrel surface layer subsequently undergoes self-diffusion within the evolving temperature field. The initial concentrations of C and N in the barrel material, denoted as C_{C0} and C_{N0} , are substantially lower than those in the gas atmosphere. According to reference [48], the internal ballistics process for large-caliber artillery—from ignition to projectile exit—lasts approximately 20 ms. This period constitutes the primary diffusion–infiltration stage for C and N. As shown in Figure 7b, the process can be divided into three stages: 0–5 ms (heating stage), 5–20 ms (cooling stage), and 20–500 ms (subsequent cooling stage), the latter corresponding to the homogenization diffusion of C and N within the austenite. Based on the above analysis, the following assumptions are adopted in the model:

- (1) Carburizing and nitriding of the inner bore surface occur independently during service, with no mutual interference under the action of the chemically active gas atmosphere.
- (2) The barrel cools to ambient temperature between successive firing cycles.
- (3) Diffused carbon atoms do not react with other elements to form carbides and are not captured or trapped by crystal defects (e.g., grain boundaries, dislocations).
- (4) Diffusion is considered to occur exclusively in the radial (r) direction.

- (5) The influence of alloying elements on carbon diffusion in austenite is neglected.

After the first firing cycle, a martensitic layer forms on the inner bore surface. During subsequent thermal shocks, this martensitic layer extends into the underlying matrix. Martensite is a supersaturated metastable phase in which carbon atoms are capable only of short-range diffusion; long-range diffusion is effectively inhibited. Therefore, once the martensitic layer has been established, subsequent diffusion must be treated as occurring within a finite domain. The diffusion boundary is defined by the maximum martensitic layer thickness, M_{rmax} , attained during the previous thermal shock.

According to Fick's second law, the carburizing process on the inner bore surface is governed by a non-steady-state diffusion model:

$$\frac{\partial C}{\partial t} = D \frac{\partial^2 C}{\partial r^2} \quad (13)$$

where D —diffusion coefficient ($\text{cm}^2 \cdot \text{s}^{-1}$); C —concentration; t —time (s); r —spatial coordinate.

$$D = D_0 \exp\left(-\frac{Q_D}{RT}\right) \quad (14)$$

where D_0 —diffusion constant ($\text{cm}^2 \cdot \text{s}^{-1}$); Q —diffusion activation energy ($\text{J} \cdot \text{mol}^{-1}$); R —gas constant ($8.314 \text{ J} \cdot \text{mol}^{-1} \cdot \text{K}^{-1}$); T —absolute temperature (K). The diffusion coefficient is influenced by multiple factors [49]. Based on data reported in the literature [50], the diffusion coefficients for carbon and nitrogen in austenite under the relevant conditions were adopted.

Taking carbon diffusion as an illustrative example, the analysis for the first firing cycle proceeds as follows:

Initial conditions:

$$C(r, 0) = C_{C0} \quad (15)$$

$$C(0, t) = \min(C_C, C_\gamma) \quad (16)$$

Boundary conditions:

$$dC/dr = 0 \quad (r = 0) \quad (17)$$

$$C(Nr_{\text{max}}, 0) = C_{C0} \quad (18)$$

For the n th firing cycle ($n \geq 2$):

Initial conditions:

$$C(r, 0) = C_{Cn-1} \quad (19)$$

$$C(0, t) = \min(C_C, C_\gamma) \quad (20)$$

Boundary conditions:

$$dC/dr = 0 \quad (r = 0) \quad (21)$$

$$C(Nr_{\text{max}}, 0) = C_{C0} \quad (22)$$

where C_{C0} —initial carbon concentration of the raw material; C_{Cn-1} —carbon concentration at the inner bore surface after the $(n - 1)$ th firing cycle; C_C —carbon potential of the propellant gas; C_γ —maximum solubility of carbon in austenite; t —carburizing time.

The physical interpretation of Equations (15)–(22) is as follows: during each firing cycle, the inner bore surface undergoes carburization driven by the high carbon potential of

the combustion gases. The carburizing duration corresponds to the period of gas exposure. Upon gas evacuation, the surface no longer receives additional carbon from the atmosphere. Under zero-flux conditions, the concentration profile established at the moment of gas removal begins to evolve via inward diffusion. For the second firing cycle, the concentration profile resulting from the previous diffusion event serves as the initial condition. This procedure is repeated for all subsequent firing cycles.

The formation of the austenite layer on the inner barrel wall under extreme service conditions is essentially the result of phase transformations induced by pulsed thermal impact coupled with the diffusion of C and N atoms. To quantitatively evaluate the synergistic contribution of C and N diffusion to austenite microstructural evolution, this study introduces a carbon equivalent (C_{eq}) model based on thermodynamic sensitivity analysis. The M_s temperature is a multivariable function of the chemical composition. For a fixed matrix composition, the change in M_s can be expressed as a linear expansion using the partial derivatives with respect to each component concentration: $\Delta M_s \approx \left(\frac{\partial M_s}{\partial \omega(C)} \right) \Delta \omega(C) + \left(\frac{\partial M_s}{\partial \omega(N)} \right) \Delta \omega(N)$.

Therefore, the weight coefficient K for the austenite-stabilizing effect of N relative to C can be expressed as: $K = \frac{\partial M_s / \partial \omega(C)}{\partial M_s / \partial \omega(N)}$. This coefficient represents the relative ability of N and C to lower the M_s temperature per unit mass fraction.

Single-variable sensitivity calculations were performed on 32CrNi3MoV steel using JMatPro thermodynamic software. In these calculations, the matrix composition was fixed, while the C and N contents were independently varied to determine the corresponding changes in M_s temperature.

To cover the complete composition range from the matrix baseline to the saturation solubility, the calculation ranges for C and N were set to 0.35–1.39 wt% and 0–0.06 wt%, respectively. The lower limits were taken from the matrix composition of the material (see Table 1), while the upper limits were determined based on the theoretical maximum solubility of C and N in γ -Fe at 960 °C and 300 MPa, namely 1.39 wt% and 0.06 wt%, respectively. Given the high sensitivity of C and N concentrations to the M_s temperature, and especially the significant role of N in the low-concentration range, the step sizes for C and N were set to 0.05 wt% and 0.003 wt%, respectively, to balance fitting accuracy with computational efficiency. Based on the above calculations, K is determined to be 0.68. Consequently, the radial carbon equivalent distribution, denoted as $C_{eq}(r)$, can be expressed as follows: $C_{eq}(r) = \omega(C)_r + 0.68 \times \omega(N)_r$.

Here, $\omega(C)_r$ and $\omega(N)_r$ denote the carbon and nitrogen concentrations (wt%) at the radial position r , as previously determined from the unsteady diffusion model.

The retention of the austenite layer at room temperature requires that the local carbon equivalent be sufficiently high to lower the M_s temperature below 25 °C. The critical carbon equivalent $C_{eq, crit}$ was determined via JMatPro calculations. Based on the matrix alloy composition, an M_s temperature of 20.7 °C corresponds to a carbon content of 1.15 wt%, yielding $C_{eq, crit} = 1.15$ wt%.

This critical value serves as a unified criterion for assessing the austenite layer thickness: any radial position r satisfying $C_{eq}(r) \geq C_{eq, crit}$ will retain its austenitic state upon post-firing cooling to room temperature.

Employing the radial concentration distributions $C(r)$ and $N(r)$ derived from diffusion simulations, along with the carbon equivalent formula and the critical criterion, the radial carbon equivalent distribution was calculated for varying numbers of firing cycles. The theoretical thickness of the austenite layer was subsequently determined.

4. Results and Discussion

Figure 10 schematically illustrates the coupling among the various physical fields considered in this study. Numerical simulations incorporating the multi-field coupling effects were performed using the finite-difference method implemented in MATLAB. The simulations yielded the following results: the transient temperature field within the barrel during a single firing cycle; the thickness of the martensitic layer formed under single-shot thermal shock; the evolution of martensitic layer thickness with the number of firing cycles under repeated thermal shock; the diffusion depth of C and N after a single shot; and the dependence of C and N diffusion depth on the cumulative number of firing cycles.

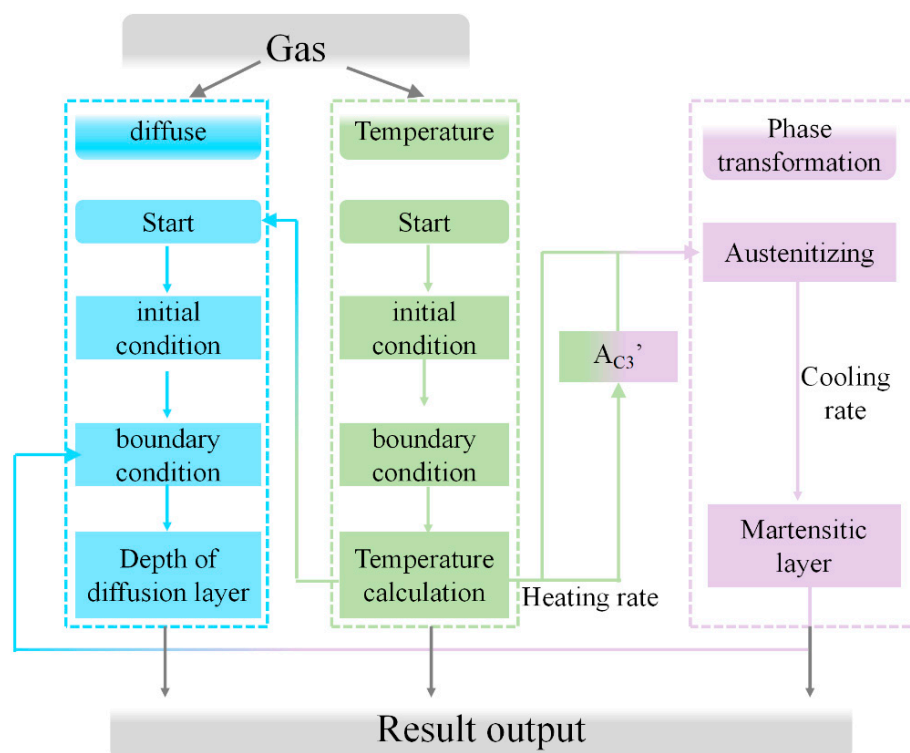


Figure 10. Relation of multi-field coupling.

Figure 11 illustrates the evolution of martensitic layer thickness with the number of firing cycles and presents the corresponding temperature distributions for the first ten thermal shocks. As shown in Figure 11, a martensitic layer approximately 62.51 μm thick forms on the inner bore surface after the first firing cycle. However, with an increasing number of firing cycles, the thickness of this martensitic layer does not continue to grow. This behavior is attributed to the low thermal conductivity and, correspondingly, the low thermal diffusivity of the as-formed martensite. During subsequent heating or cooling, the rate of internal temperature equalization is slow, resulting in steeper temperature gradients and the generation of significant thermal stresses. After the first firing cycle, the presence of a surface martensitic layer means that under subsequent thermal shocks, the newly formed martensite extends from this layer into the underlying matrix. This not only promotes the development of thermal stresses but also facilitates the initiation of thermal cracks at the surface. During the second and subsequent firing cycles, heat conduction along the radial direction occurs within a two-layer composite structure comprising the existing martensitic layer and the untransformed matrix. Consequently, the presence of this martensitic layer impedes radial heat transfer, making it more difficult for the underlying matrix to reach the austenitization temperature under the same surface thermal history. Moreover, because the surface layer already consists of martensite, its complete austenitization temperature

is lower than that of the original matrix. This combination of hindered heat transfer and reduced transformation temperature causes the martensitic layer to thicken gradually under subsequent thermal shocks, eventually stabilizing at a characteristic depth. Thus, it can be inferred that the thickness of the martensitic layer formed during service is not directly governed by the number of firing cycles, but rather by the peak temperature attained under the action of the propellant gases. Higher peak temperatures result in a deeper martensitic layer.

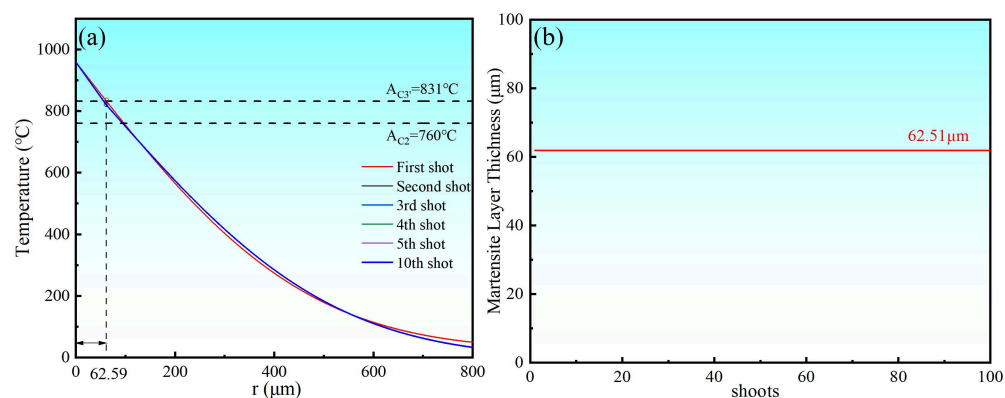


Figure 11. Simulation results: (a) the radial temperature distribution of the first 10 shots; (b) variation curve of martensite layer thickness with the number of rounds.

The resulting martensitic layer thickness was calculated as a function of temperature, and the corresponding curves are presented in Figure 12. As shown in Figures 11b and 12, the martensitic layer thickness increases with the peak temperature attained during firing and exhibits no direct correlation with either the number of firing cycles or the mechanical action of the rotating band. Botstein and Arone [8] previously investigated a 105 mm barrel made of AISI 4340 steel and hypothesized that the thickness of the white layer is independent of the number of firing cycles, depending instead primarily on the barrel's service environment, pressure, and temperature distribution. The simulation results presented in this study are consistent with this hypothesis.

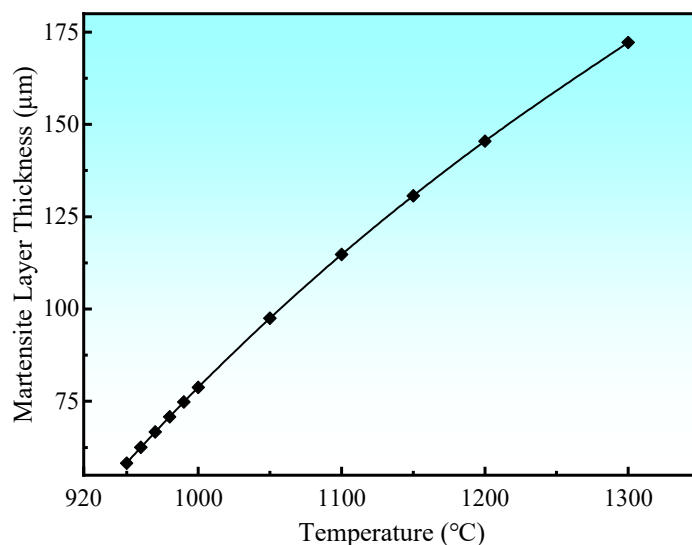


Figure 12. Relationship between martensite thickness and temperature.

Using the computed temperature field as input, the diffusion equation was solved numerically via the finite difference method. The simulations, implemented in MATLAB, yielded the radial diffusion depths and concentration profiles of C and N after 1, 2, 3, 4, 5,

10, 50, 100, 500, and 1000 firing cycles. The results are presented in Figure 13. By converting the synergistic interactions of C and N into an equivalent carbon content, the relationship between the carbon equivalent and the diffusion layer thickness is established, as illustrated in Figure 14.

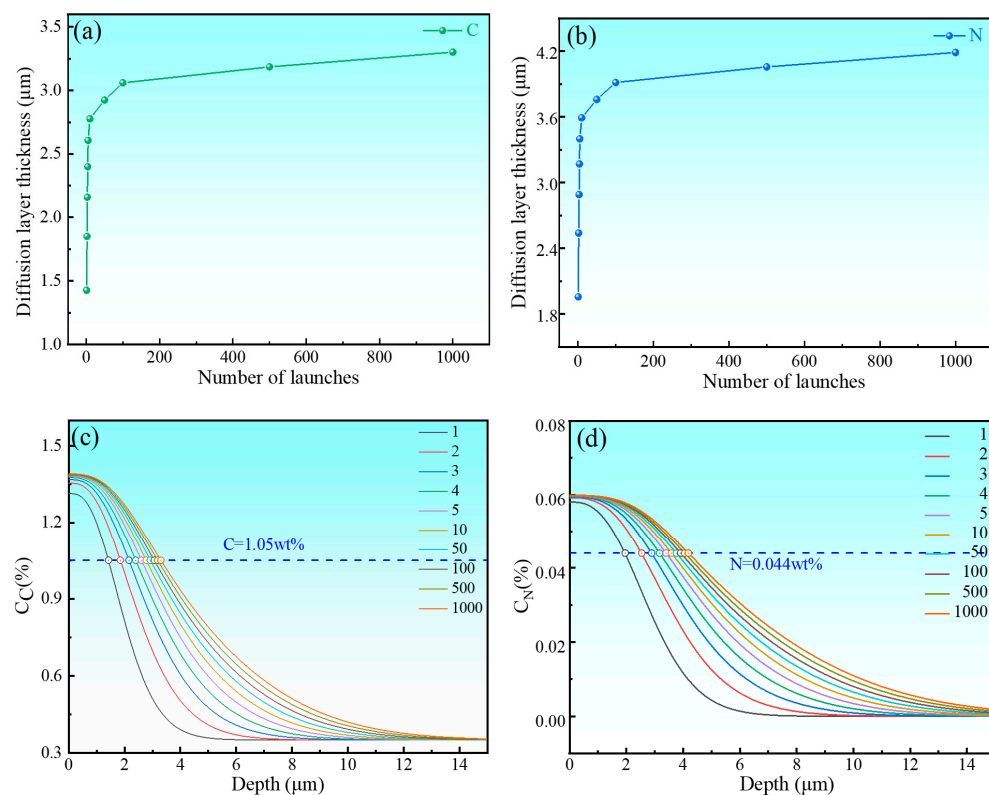


Figure 13. The diffusion calculation results of C, N: (a,b) the relationship between the thickness of the C, N diffusion layer and the number of emissions; (c,d) radial variation curves of C, N after different firing times.

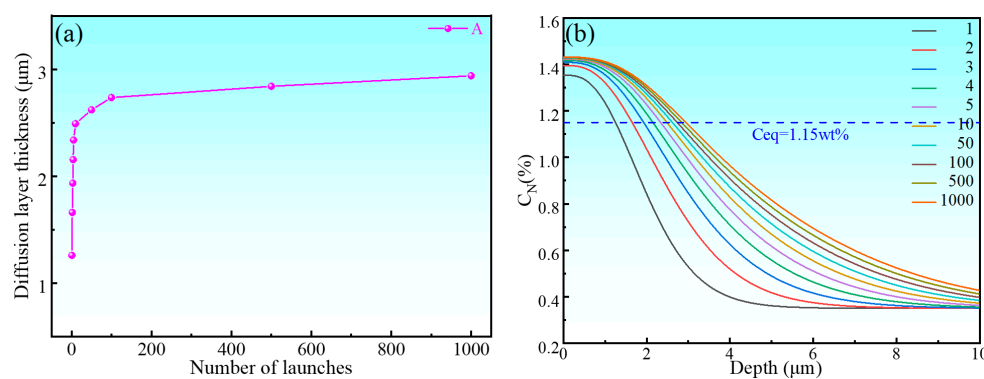


Figure 14. Fitting carbon equivalent calculation results: (a) Relationship between austenite layer thickness and emission frequency; (b) Ceq radial variation curve after different launch times.

As shown in Figure 13a,b, after the first firing cycle, the diffusion depths of C and N are 1.42 μm and 1.95 μm, respectively. With an increasing number of firing cycles, the diffusion depths continue to increase. The increase is rapid during the initial cycles, but the incremental growth per cycle gradually diminishes, with the growth factor approaching unity. After 1000 firing cycles, the C and N diffusion depths reach 3.30 μm and 4.19 μm, respectively. By converting the C and N interactions into carbon equivalents, the relationships between the number of firing cycles and the austenite layer thickness, along with the radial

variation of C_{eq} after different firing cycles, were obtained, as shown in Figure 14. The thickness of the austenite layer initially increases with an increasing number of firing cycles and then grows more slowly. After the first firing cycle, an austenite layer approximately 1.27 μm thick is formed. After 1000 firing cycles, the thickness reaches approximately 2.94 μm . As shown in Figure 14a, once the number of firing cycles reaches 100, the austenite layer thickness continues to increase gradually with further firing cycles but only increases by approximately 0.2 μm between 100 and 1000 cycles. Comparing these values with the experimentally measured austenite layer thickness (1.52–4.16 μm) shown in Figures 3b and 4, it can be concluded that the austenite layer thickness increases with the number of firing cycles and, moreover, that this thickness corresponds directly to the diffusion depth of C and N.

Figure 15 schematically illustrates the diffusion-induced phase transformation in the inner bore surface layer under coupled thermochemical effects, while Figure 16 depicts the corresponding elemental diffusion behavior during firing. Based on Figures 15 and 16 and consistent with actual service conditions, it can be inferred that during firing, the inner bore surface layer undergoes cyclic, pulsed thermochemical actions. Upon propellant combustion, the inner bore surface layer rapidly heats up to 960 °C (Figure 10). Simultaneously, reactive species—including carbon (C) and nitrogen (N) atoms—generated by the decomposition of the propellant gases are adsorbed onto the inner bore surface, establishing a concentration gradient in which the surface layer is enriched in C and N relative to the underlying substrate. This creates a significant driving force for inward diffusion. Furthermore, the rapid temperature rise enhances atomic mobility, promoting the migration of C and N from the surface-enriched layer toward the interior. However, the duration of exposure to the propellant atmosphere is extremely brief (on the order of 5 ms). Upon gas evacuation, the inner bore surface loses both the high-temperature heat source and the chemical driving force from the atmosphere, profoundly altering the diffusion thermodynamics. During the gas exposure period, a steep carbon concentration gradient develops in the near-surface region. After gas evacuation, the barrel body continues to extract heat from the inner wall, leading to rapid quenching of the high-temperature austenite layer. Driven by the concentration gradient, atoms diffuse from the high-concentration surface region toward the lower-concentration interior to reduce the system's free energy. During the initial stage of cooling, atoms retain sufficient thermal energy to continue diffusing, albeit at a diminishing rate. Because the surface carbon concentration remains elevated after heating and some inward diffusion has already occurred, carbon atoms continue to migrate inward during cooling, although the diffusion rate progressively decreases. Under rapid-cooling conditions, atomic diffusion may be kinetically suppressed, as the high cooling rate allows insufficient time for complete redistribution, potentially leading to compositional inhomogeneities within the steel. When high-temperature austenite undergoes martensitic transformation during quenching, the available diffusion paths for carbon atoms become severely restricted, causing the diffusion rate to plummet. Long-range diffusion of carbon is effectively inhibited within the martensitic region. Consequently, carbon redistribution is confined to the high-carbon surface layer, where local concentration homogenization occurs, as illustrated in Figure 16. During each firing cycle, the brief exposure to the high-concentration atmosphere limits carbon diffusion to short distances, resulting in a steep carbon concentration gradient near the surface. Following gas evacuation, the time available for homogenization diffusion is considerably longer than the high-concentration exposure period, allowing for more extensive redistribution within the surface layer. As illustrated in Figure 15, from the first firing cycle to eventual retirement, the inner bore surface layer undergoes repeated rapid heating and cooling cycles. This induces grain refinement and increases the dislocation density within the surface layer. These defects act

as short-circuit diffusion paths for carbon, facilitating more rapid homogenization and promoting the approach to local equilibrium. Ultimately, this results in a uniform distribution of C and N within the austenite layer. During each firing cycle, a limited inward migration of C and N within the austenite layer. Upon cooling, the martensite that forms adjacent to the austenite layer acts as a diffusion barrier, effectively blocking further inward migration of C and N. Consequently, homogenization is confined to the austenite layer, and the diffusion depth achievable per cycle is limited. Over multiple cycles, the concentration front gradually advances inward, resulting in a stepped concentration profile at the austenite/martensite (A/M) interface. This stepped distribution is consistent with the APT results presented in Figure 4 and Table 2.

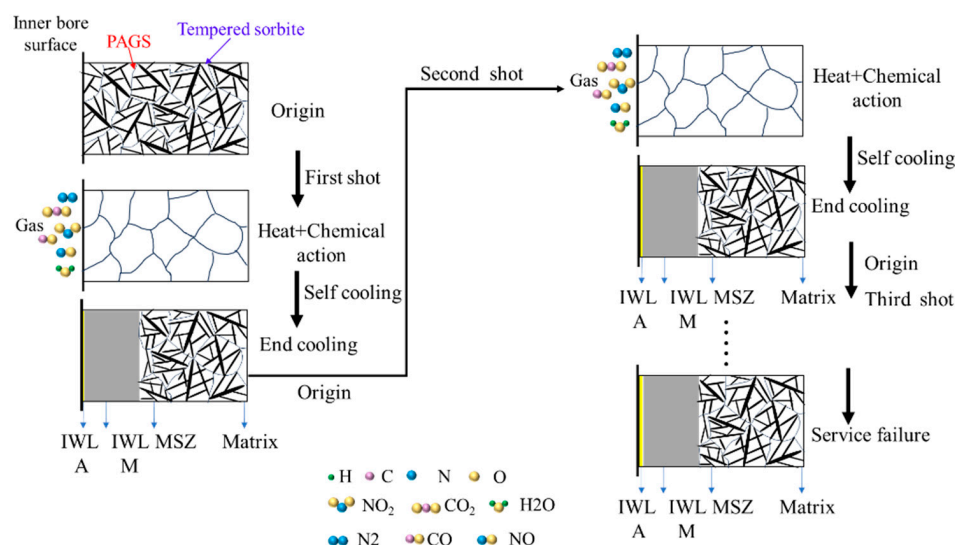


Figure 15. Schematic diagram of diffusion phase transition under thermal chemical action on the surface of the barrel. Note: IWL: inner white layer; MSZ: matrix softening zone; PAGS: primary austenite grain boundary.

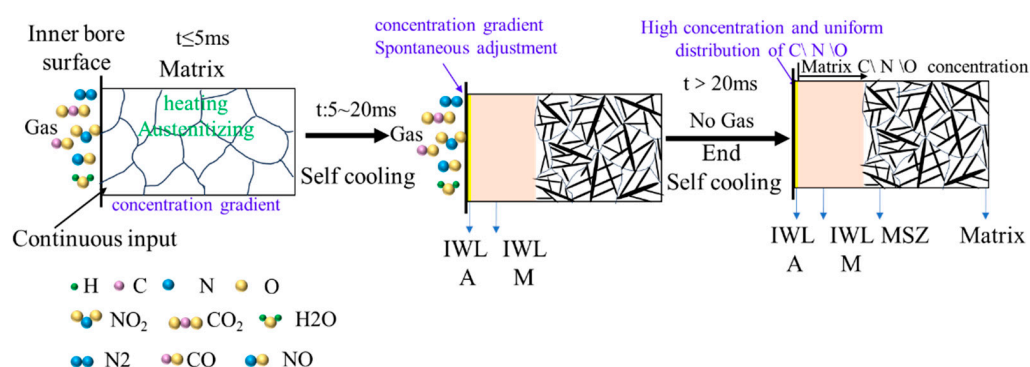


Figure 16. Schematic diagram of element diffusion on the surface of barrel bore under thermal chemical action.

5. Conclusions

The purpose of this study is to elucidate the formation and evolution mechanisms of the austenite layer in the barrel bore under cyclic thermal shock and thermochemical coupling, and to establish a quantitative relationship between the austenite layer thickness and the number of firing cycles for large-caliber barrel life assessment. The main findings are summarized as follows:

- (1) A one-dimensional heat conduction model was established, which simulates that the instantaneous temperature of the inner bore surface reaches 960 °C during a

single firing cycle. After the first firing cycle, a martensitic layer approximately 62.51 μm thick is formed. The thickness of this layer is governed primarily by the peak temperature attained during firing and exhibits no direct correlation with the number of firing cycles.

- (2) The martensitic layer formed after the first firing cycle exhibits low thermal conductivity, which impedes radial heat transfer, making it difficult for the underlying matrix to reach the austenitization temperature during subsequent firing cycles. Furthermore, the presence of this martensitic layer lowers the complete austenitization temperature of the surface region. Consequently, the martensitic layer thickness stabilizes at a characteristic depth, providing a key theoretical basis for barrel service life assessment.
- (3) The austenite layer results from the synergistic interplay between thermal-shock-induced phase transformation and elemental diffusion over multiple firing cycles. The numerical simulation results show that the thickness of the austenite diffusion layer increases with the number of launches. After the first launch, an austenite layer of approximately 1.27 μm forms. When the number of launches reaches 100, the thickness of the austenite layer continues to increase gradually with further launches. From 100 to 1000 launches, the thickness increases by only 0.2 μm . At 1000 launches, the thickness reaches 2.94 μm , which falls within the experimentally observed range of 1.52–4.16 μm for the austenite layer. This layer is therefore speculated to be a carbon–nitrogen diffusion layer.

Author Contributions: J.C.: conceptualization, methodology, validation, formal analysis, writing—original draft, visualization; Y.L. (Yiming Liu): investigation, data curation, visualization; M.Z.: investigation, data curation; J.F.: resources, supervision, funding acquisition; Y.J.: conceptualization, investigation, resources, supervision, project administration; Z.L.: conceptualization, investigation, resources, supervision, project administration; Y.L. (Ying Liu): conceptualization, investigation, resources, supervision; J.W.: Conceptualization, methodology, resources, writing—review and editing, visualization, supervision, project administration, funding acquisition. All authors have read and agreed to the published version of the manuscript.

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