

Ternary synergistic structure of Co-TiN@CNTs enables efficient electromagnetic wave absorption through coupling of multi-loss mechanisms

Zhaoying Jia¹, Bingtao Tang¹, Wentao Wang², Shufen Zhang¹, and Yuang Zhang¹✉

¹State Key Laboratory of Fine Chemicals, Frontier Science Center for Smart Materials, Dalian University of Technology, Dalian 116024, China

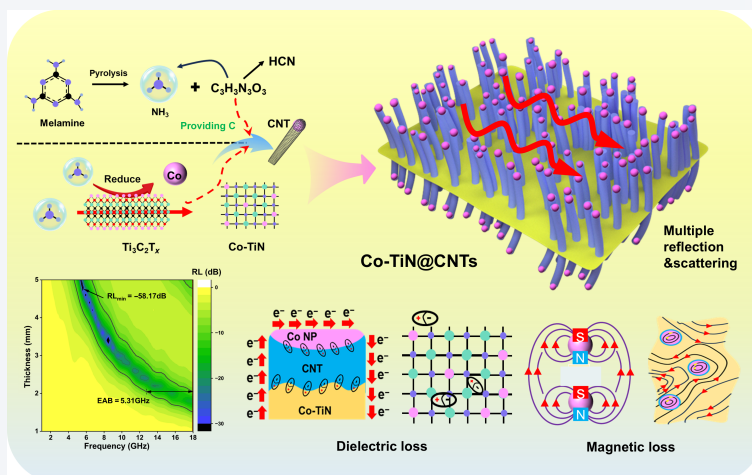
²School of Materials Science and Engineering, Zhejiang Sci-Tech University, Hangzhou 310018, China



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ABSTRACT: The integration of reflection networks into microwave-absorbing (MA) fillers remains a major challenge. While carbon nanotubes (CNTs) possess high conductivity and a unique structure ideal for this purpose, their poor dispersion and difficult alignment limit progress. This work employs a one-step synergistic strategy to enable the concurrent crystallization of titanium nitride (TiN), catalytic activation of cobalt ions (Co^{2+}), and growth of micrometer-long CNTs, through precise tailoring of the reaction ratio. The generated micron-scale bushy CNTs are used to form a reflective network and produce high dielectric loss. The material exhibits exceptional MA performance, with a minimum reflection loss ($\text{RL}_{\min}$) of -58.17 dB at 5.50 GHz and an ultra-wide effective absorption bandwidth (EAB) of 5.31 GHz under mass doping of 5%. This work provides new ideas for the construction of multi-reflection networks.

KEYWORDS: microwave absorbing materials, multiple reflection network, heterointerface, carbon nanotubes



1 Introduction

With the iterative upgrading of electronic devices, the increasingly complex electromagnetic environment poses a serious threat to human health and equipment safety [1, 2]. Addressing the issue of electromagnetic pollution has become an urgent task [3]. In this context, the construction of multidimensional microwave-absorbing (MA) materials with abundant interfacial polarization and multiple reflection loss networks is widely regarded as an effective approach to enhancing MA performance [4, 5].

Compare to zero-dimensional (0D), one-dimensional (1D), two-dimensional (2D) materials, three-dimensional (3D) materials integrate the advantages of various dimensions to enrich loss

mechanisms and further utilize porous or network structures to promote multiple internal reflections and absorption of electromagnetic waves (EMW), thereby improving MA performance [6–9]. Currently, researchers are introducing polymer templates (e.g., aerogels [10, 11], PMMA microspheres [12]) to guide MA fillers in constructing reflection networks, at the expense of some electrical conductivity. How to build intrinsic reflection networks within the fillers remains one of the current challenges [13].

Carbon materials, particularly carbon nanotubes (CNTs), are frequently incorporated into MA systems due to their high electrical conductivity [14]. For instance, Wang et al. fabricated an aligned aerogel using multi-walled CNTs modified with Fe_3O_4 and cellulose nanofibers. The high conductivity of the CNTs contributed to dielectric loss, while the porous structure of the aerogel enhanced the reflection loss, achieving a minimum reflection loss ($\text{RL}_{\min}$) of -45.0 dB and a broad EAB of 6.1 GHz [15]. Theoretically, the unique rod-like structure of CNTs should

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✉Address correspondence to zhangyuang@dlut.edu.cn

enable them to form efficient EMW reflection networks without relying on external structural design [16]. However, their tendency to agglomerate and align disorderly often limits the actual loading amount, making it difficult to establish these reflection networks [17].

Common approaches for achieving aligned CNTs involve the application of magnetic/electric fields, shear forces, or interfacial self-assembly [18, 19]. In contrast, chemical vapor deposition (CVD), which enables the direct, oriented growth of CNTs on a substrate, offers advantages such as simple equipment and a controllable process [20]. For instance, Team Li successfully grew Ni-doped CNTs *in situ* on MXene. This strategy achieved multiple objectives simultaneously: It prevented CNT agglomeration, introduced additional polarization centers to enhance dielectric loss, and constructed nanoscale reflection networks [21]. It is foreseeable that increasing the length of the CNTs will establish continuous and extended conductive pathways, along with a more extensive reflection network. This architecture is anticipated to facilitate multiple reflections and efficient absorption of EMW.

Herein, we propose a strategy involving the introduction of excess Co^{2+} and melamine (MF) into MXene to synergistically regulate the growth of CNTs and the matrix transformation. The excess reactants provide abundant catalytic sites and carbon for growing longer, denser CNTs, while simultaneously releasing nitrogen-containing active species that drive the topochemical conversion of MXene into a titanium nitride (TiN) matrix. Simultaneously, the excess Co^{2+} ions incorporate into the TiN lattice. Through a one-step synergistic strategy with precise ratio control, this work simultaneously crystallizes TiN, deploys Co catalytically, and grows CNTs to micrometer lengths to construct a 3D cobalt-doped TiN/CNTs heterostructure (Co-TiN@CNTs). The material exhibits exceptional MA performance: a RL_{min} of -58.17 dB at 5.50 GHz, and an ultra-wide effective absorption bandwidth of 5.31 GHz ($\text{RL} < -10$ dB) at a matching thickness of 2.00 mm, which is significantly better than that of other materials.

2 Experimental

2.1 Main materials

Cobalt(II) chloride hexahydrate ($\text{CoCl}_2 \cdot 6\text{H}_2\text{O}$, AR), lithium fluoride (LiF, AR), and melamine (AR) were purchased from Aladdin. MAX phase powder (400 mesh) was obtained from Kaixi Powder Co., Ltd., and hydrochloric acid (HCl) was procured from Dalian Bonuo Biochemical Reagent Factory. Polyvinylidene fluoride (PVDF, HSV900) was purchased from Arkema, French. N,N-Dimethylformamide (DMF) was purchased from Aladdin Reagent Shanghai Co., Ltd, China.

2.2 Experimental section

Under ice-water bath, 1.6 g LiF was dissolved in concentrated HCl. The solution was heated to 50 °C, followed by gradual addition of 1 g MAX phase precursor powder with 24 h stirring. The initial centrifugation preserved the bottom sediment, which was purified through acid washing using 1 mol/L HCl solution. Centrifugation at 3500 rpm for 5 min was repeated three times to thoroughly eliminate residual hydrofluoric acid (HF). This was followed by multistage deionized water washing with high-speed centrifugation at $10,000$ rpm for 5 min, repeated three times to effectively remove ionic impurities. The sediment was redispersed in water and sonicated for 2 h to exfoliate the layered structure, and

then the supernatant was collected via low-speed centrifugation (7.5 mg/mL).

0.5 mol/L $\text{CoCl}_2 \cdot 6\text{H}_2\text{O}$ solution was introduced into the MXene dispersion, stirred for 1 h to ensure thorough interaction, and lyophilize to obtain Co-MXene powder. The Co-MXene powder was uniformly mixed with melamine (mass ratio: $1:20$), then heated in a tube furnace at a ramp rate of 5 °C/min to 750 °C, held for 2 h, and cooled to room temperature to yield the final product. After obtaining Co-TiN@CNTs, a coaxial ring sample with an inner diameter of 3.04 mm and an outer diameter of 7 mm was prepared by mixing with paraffin wax at a mass ratio of $5:95$.

2.3 Characterizations

Scanning electron microscopy (SEM5000, CIQTEK, China) and transmission electron microscopy (TEM, JEM-F200, Japan) were used to observe the morphology of all samples. X-ray photoelectron spectroscopy (XPS, ESCALAB XI+, England) was used for elemental composition and valence state analysis of all samples. The degree of graphitization of samples was analyzed using a laser confocal Raman spectrometer (DXR Microscope, USA). X-ray diffraction (XRD, SmartLab 9KW, Rigaku Corporation of Japan) was employed to characterize the crystal structure of the materials in the diffraction angle range of $2\theta = 5^\circ$ – 80° . The thermal stability of the materials was determined by thermogravimetry (TG), with a temperature range of room temperature to 800 °C, at a heating rate of 10 °C/min, in a nitrogen atmosphere. Magnetic hysteresis loop was measured by a vibrating sample magnetometer (VSM, 7404-S, Lake Shore, USA). Evaluation of nitrogen adsorption-desorption curves, porosity, and specific surface area by the Brunauer–Emmett–Teller method (BET, 3FLEX, Micromeritics, USA). Prepare thin films by mixing the samples with PVDF at a mass ratio of $1:1$, the electrical conductivity of the samples were simultaneously determined at 30 °C in a LSR-3 system (Linseis GmbH) under He atmosphere. The Zeta potential of MXene was tested on the Nano-ZS90 (MALVERN, England).

Extended X-ray absorption fine structure (EXAFS) data processing. Transmission XAS measurements were performed on a laboratory device (easyXAFS300+, easyXAFS LLC), which is based on Rowland circle geometries with spherically bent crystal analyzers (SBICA) and a silicon drift detector. Si(5,5,3) was used for Co K-edge measurement. The powder samples were thoroughly ground and mixed with Cellulose using an agate mortar and pestle and pressed into $\phi = 10$ mm pellets. The pressed pellets were then sandwiched by Kapton tapes.

The following EXAFS equation was used

$$\chi(k) = \sum_j \frac{N_j S_0^2 F_j(k)}{k R_j^2} \exp[-2k^2 \sigma_j^2] \exp\left[\frac{-2R_j}{\lambda(k)}\right] \sin[2kR_j + \phi_j(k)] \quad (1)$$

where S_0^2 is the amplitude reduction factor, $F_j(k)$ is the effective curved-wave backscattering amplitude, N_j is the number of neighbours in the j th atomic shell, R_j is the distance between the X-ray absorbing central atom and the atoms in the j th atomic shell (backscatterer), λ is the mean free path in Å, $\phi_j(k)$ is the phase shift (including the phase shift for each shell and the total central atom phase shift), and σ_j is the Debye–Waller parameter of the j th atomic shell (variation in distances around the average R_j). The functions $F_j(k)$, λ and $\phi_j(k)$ were calculated using the *ab initio* code FEFF8.2.

3 Results and discussion

3.1 Construction and characterization of Co-TiN@CNTs

Figure 1(a) illustrates the synthesis pathway of Co-TiN@CNTs composites. The process began with the selective etching of MAX phase Ti_3AlC_2 powder in LiF/HCl solution, followed by the removal of Al atomic layers to obtain few-layered $\text{Ti}_3\text{C}_2\text{T}_x$ MXene nanosheets via ultrasonic exfoliation (Figs. S1 and S2 in the Electronic Supplementary Material (ESM)). The abundant $-\text{F}$, $-\text{OH}$, and $=\text{O}$ terminal groups on MXene surfaces endow them with negative charges (Fig. S1(b) in the ESM) [22], providing effective adsorption sites for Co^{2+} ions (Fig. 1(b)). Based on the EDS (Fig. S3 in the ESM) results of Co-MXene, cobalt was uniformly loaded on the MXene. MXene powder was mixed with melamine under nitrogen protection to yield Co-TiN@CNTs (Fig. 1(c)). Among them, the CNT structure is clearly visible at the scale of 2 μm . During this process, the MXene nanosheets are simultaneously reconstructed into 2D Co-TiN. Figure 1(d) demonstrates the distribution of C, Ti, Co, and N within the Co-

TiN@CNTs system. The substrate effectively anchors CNT growth sites through a pinning effect, ensuring discrete distribution of Co NPs during growth that ultimately become encapsulated within CNTs. Numerous interwoven CNTs (~ 40 nm diameter) vertically anchored on both sides of the Co-TiN sheets form a shrub-like 3D conductive network. High-resolution TEM images (Figs. 1(e)–1(g)) reveal the characteristic lattice spacings of 0.215, 0.24, 0.21, and 0.34 nm, which correspond to the (111) planes of Co, TiN, CoN phase, and (002) plane of graphitic carbon, respectively [23].

XRD analysis (Fig. 2(a)) demonstrates the coexistence of TiN and graphitic carbon, verifying the nitrogenation transformation of MXene during pyrolysis. Comparative experiments indicate that pure MXene directly pyrolyzed in nitrogen fails to generate TiN, suggesting that TiN formation primarily relies on the nitrogen released from the coordination of melamine with Ti atoms in MXene. When MXene is carbonized with melamine, only trace TiN crystals are formed without CNTs growth. The SEM of MXene-750 $^{\circ}\text{C}$ and MXene-MF as shown in Fig. S4 in the ESM. This finding demonstrates the dual functionality of Co^{2+} : catalytically guiding carbon source directional growth into CNTs while

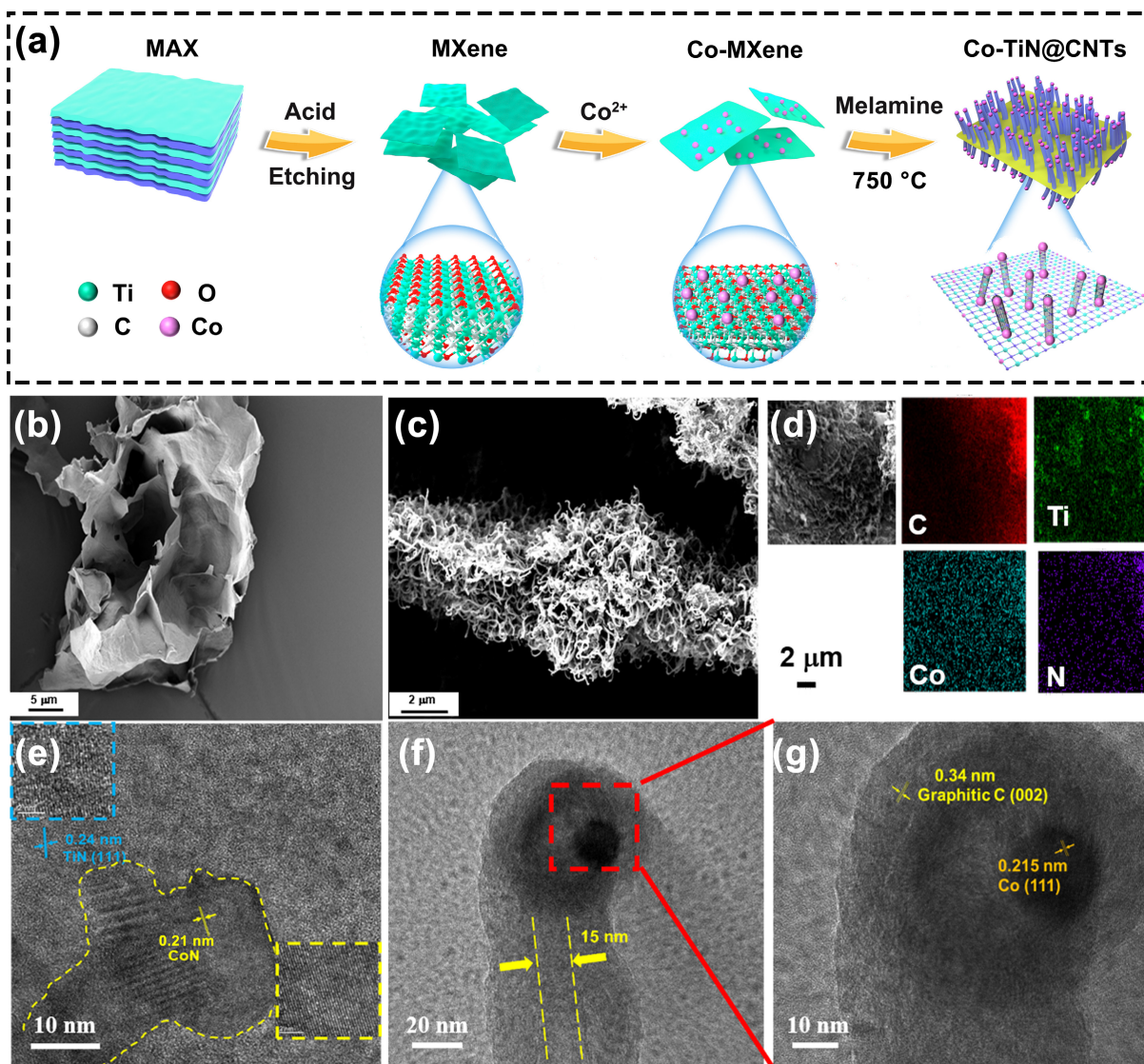


Figure 1 The structure of Co-TiN@CNTs. (a) Schematic for construction of Co-TiN@CNTs. (b) SEM image of Co-MXene. (c) SEM image and (d) elemental mapping images of Co-TiN@CNTs's. TEM images of (e) Co-TiN and (f)–(g) Co-decorated CNTs.

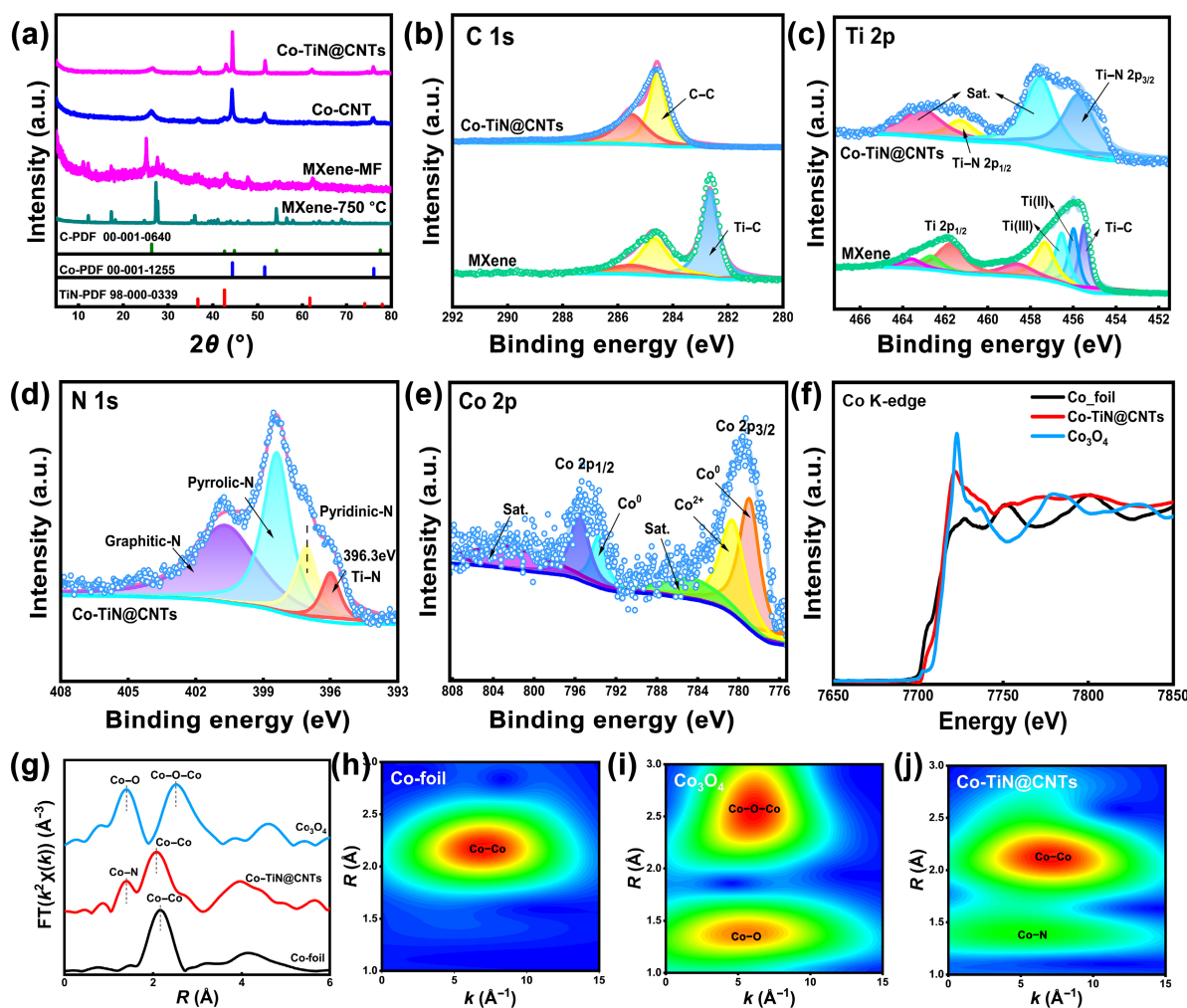


Figure 2 Structure characterization of Co-TiN@CNTs. (a) XRD patterns of different MXene-750 °C, MXene-MF, Co-MXene, Co-TiN@CNTs. XPS spectra of (b) C 1s, (c) Ti 2p, (d) N 1s, and (e) Co 2p. (f) Normalized XANES spectra at the Co K-edge of the Co-foil, Co_3O_4 , and Co-TiN@CNTs. (g) FT-EXAFS spectra of the Co-foil, Co_3O_4 , and Co-TiN@CNTs. Wavelet transform (WT) of the (h) Co foil, (i) Co_3O_4 , and (j) Co-TiN@CNTs.

significantly promoting TiN crystallization through metal-nitrogen synergy [24].

XPS depth profiling reveals the chemical evolution of the composite (Fig. S5 in the ESM). The C 1s spectrum demonstrates the complete disappearance of the original MXene's Ti-C bond (~ 282.1 eV) in the product, confirming that nitrogen fully substitutes MXene's carbon atoms during pyrolysis (Fig. 2(b)) [7]. Ti 2p analysis further shows the transformation of MXene's original Ti-C peaks (455.8/461.5 eV) into Ti-N characteristic doublets (455.5/461.2 eV); the decrease in binding energy indicates electronic structure modification during nitridation (Fig. 2(c)). Deconvolution of the N 1s fine spectrum identifies the following four characteristic peaks (Fig. 2(d)): Ti-N (396.3 eV), pyrrolic N (397.2 eV), pyridinic N (398.4 eV), and graphitic N (400.6 eV) [4]. Co 2p spectra exhibit coexisting Co^{2+} (782.3 eV) and metallic Co^0 (778.6 eV) signals, demonstrating that the partial cobalt ions form zero-valent active sites during reduction (Fig. 2(e)) [25]. The conductive network formed by Co^0 enhances carrier migration, promoting conductive loss; while the defect polarization induced by Co^{2+} , together with interfacial charge accumulation, strengthens polarization relaxation. These two aspects work synergistically to optimize both impedance matching and loss capability [26].

In addition to the localized microstructural information

mentioned above, we further employed X-ray absorption fine structure (XAFS) to reveal the electronic and coordination structures of Co in Co-TiN@CNTs (Fig. 2(f)) [27]. Co K-edge X-ray absorption near edge structure (XANES) spectra demonstrate that the energy absorption threshold of the sample lies between Co-foil and metal oxides, with N atoms stabilizing the positively charged Co^{2+} . FT-EXAFS spectra (Fig. 2(g)) indicate the coexistence of Co-N and Co-Co bonds in the sample, corresponding to the Co dopants in TiN and the metallic Co at CNTs, respectively. WT-EXAFS analysis (Figs. 2(h)–2(j)), reveal enhanced sensitivity in R-space and K-space, further confirming the Co species in Co-TiN@CNTs exclusively exist as Co-N bonds and Co-Co coordination.

The composite's evolution mechanism was revealed through stepwise pyrolysis experiments (400–750 °C). XRD analysis (Fig. S5(a) in the ESM) demonstrates that melamine gradually decomposed to form $\text{g-C}_3\text{N}_4$ within the 400–600 °C range (complete conversion at 600 °C) [28], with product coloration transitioning from gray $\rightarrow$ pale green $\rightarrow$ yellowish-green $\rightarrow$ light yellow (Fig. S5(b) in the ESM). At 700 °C, TiN diffraction peaks emerge and Co^{2+} initiates reduction, though the graphite carbon peak (26.5°) remains absent. The Raman I_D/I_G ratio (0.93) is lower than that of the 750 °C sample (1.13). However, SEM reveals only

sporadic short-range CNT growth on the Co-TiN nanosheets (Figs. S7 and S8 in the ESM), indicating incomplete Co-activated carbon restructuring at this stage. Upon reaching 750 °C with 2 h holding time, metallic Co's catalytic activity intensifies significantly and drives the epitaxial growth of highly graphitized CNTs networks while preserving Co-TiN's 2D architecture. Thermogravimetry-mass spectrometry analysis (Fig. S9 in the ESM) further confirms the following three steps: < 600 °C dominated by melamine pyrolysis (NH_3 and $\text{g-C}_3\text{N}_4$ formation), 600–750 °C featuring concurrent TiN and Co^{2+} reduction, and > 700 °C enabling Co-mediated carbon topological reconstruction for the controlled construction of CNTs/Co-TiN heterojunctions (Fig. 3(a)).

Nitrogen adsorption-desorption analysis (Fig. 3(b)) reveals a typical type IV isotherm with an H_3 hysteresis loop in the P/P_0 range of 0–1, demonstrating hierarchical pore architecture. The composite exhibits a BET surface area of 72.68 m^2/g , a pore volume of 0.27 cm^3/g , and an average pore size of 0.804 nm. This 3D conductive network is constructed via epitaxial CNTs growth. Its high surface area extends the electromagnetic wave propagation paths [29].

3.2 Electromagnetic performance analysis

After thin films were prepared by mixing the samples with PVDF at a mass ratio of 1:1, the electrical conductivities of MXene, Co-MXene, and Co-TiN@CNTs were measured. As shown in Fig. 3(c), MXene demonstrates the highest conductivity of 1003.30 S/m, significantly surpassing those of Co-TiN@CNTs (254.65 S/m) and Co-MXene (53.00 S/m). Vibrating sample magnetometer analysis indicates that high-temperature carbonization reduces Co^{2+} to Co^0 , endowing the composite with magnetic responsiveness. Figure 3(d) shows its low saturation magnetization (M_s) of 1.10 emu/g, attributed to the magnetic moment superposition being suppressed by Co NP dimensions, low loading content, and uniform dispersion [22]. The

composite demonstrates a remarkably high coercivity (H_c) of 352 Oe, originating from the strong coordination between TiN substrate and Co atom. The Co-N coordination field stabilizes the d-orbital electronic configuration of Co, locking spin magnetic moment orientations and significantly enhancing magnetic anisotropy energy barriers [23]. This magnetic anisotropy facilitates substantial hysteresis loss enhancement, effectively converting electromagnetic energy into thermal energy and thereby improving MA efficiency.

Coaxial specimens were prepared by mixing MXene, Co-MXene, and Co-TiN@CNTs powders with paraffin at a mass ratio of 5:95 to validate the effectiveness of the “low-conductivity and weak-magnetism synergy”. The electromagnetic parameters were measured across 2–18 GHz using a vector network analyzer via the coaxial transmission-line method. According to Marksworth's electromagnetic wave transmission theory, an absorber's performance is fundamentally controlled by the frequency-dependent behaviors of its complex relative permittivity (ϵ) and complex relative permeability (μ). In particular, the real components (ϵ' , μ') quantify the electromagnetic energy storage capacity, and the imaginary components (ϵ'' , μ'') reveal the multimodal energy dissipation mechanisms through heat conversion [30].

As illustrated in Figs. 4(a) and 4(b), the ϵ' values of MXene (30.06–14.38), Co-MXene (2.72–2.61), and Co-TiN@CNTs (10.11–6.13) all exhibit typical frequency-dependent attenuation characteristics. This phenomenon aligns with the physical mechanism described in Debye relaxation theory, where electric dipoles demonstrate delayed responses to high-frequency alternating electric fields (2–18 GHz). The ϵ'' of MXene reaches a peak of 29.44 at 2 GHz, significantly surpassing those of Co-TiN@CNTs (4.96) and Co-MXene (0.03), and maintains the same pattern of variation as its conductivity (Eq. (S5) in the ESM).

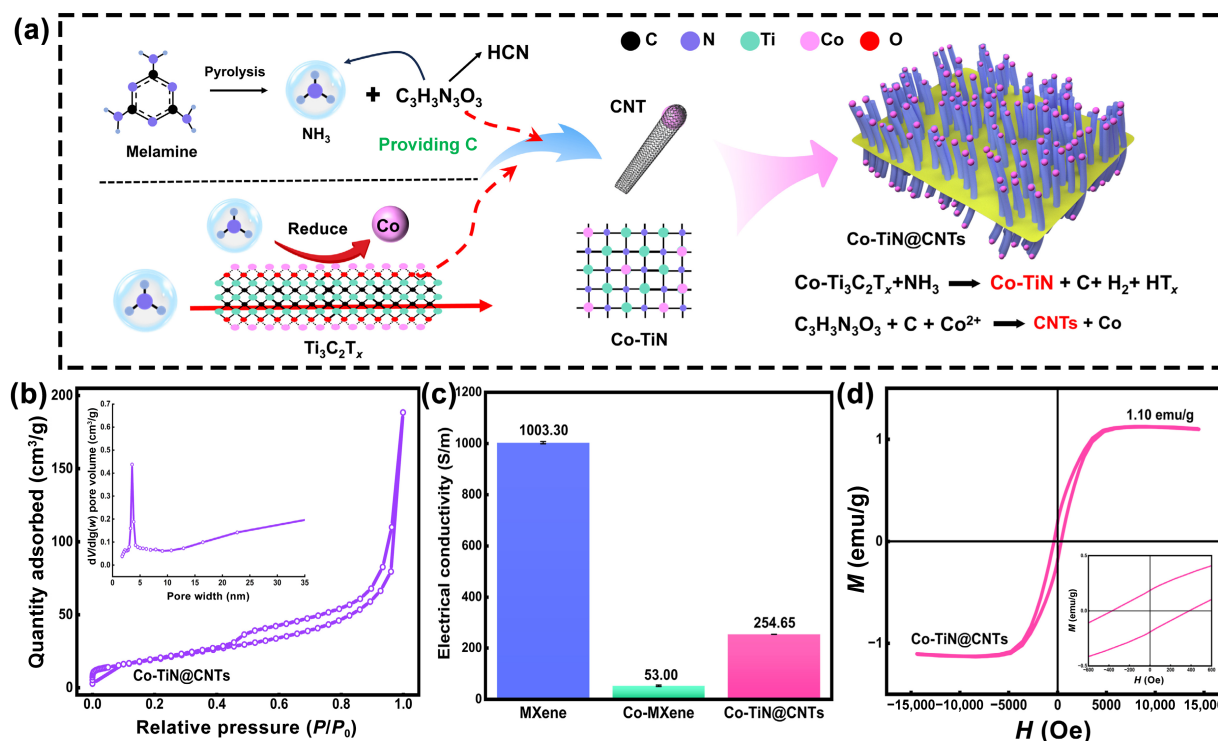


Figure 3 The characterization and evolution mechanism of Co-TiN@CNTs. (a) The evolution mechanism. (b) BET of Co-TiN@CNTs. (c) Electrical conductivities of the MXene, Co-MXene, and Co-TiN@CNTs's. (d) Hysteresis loops.

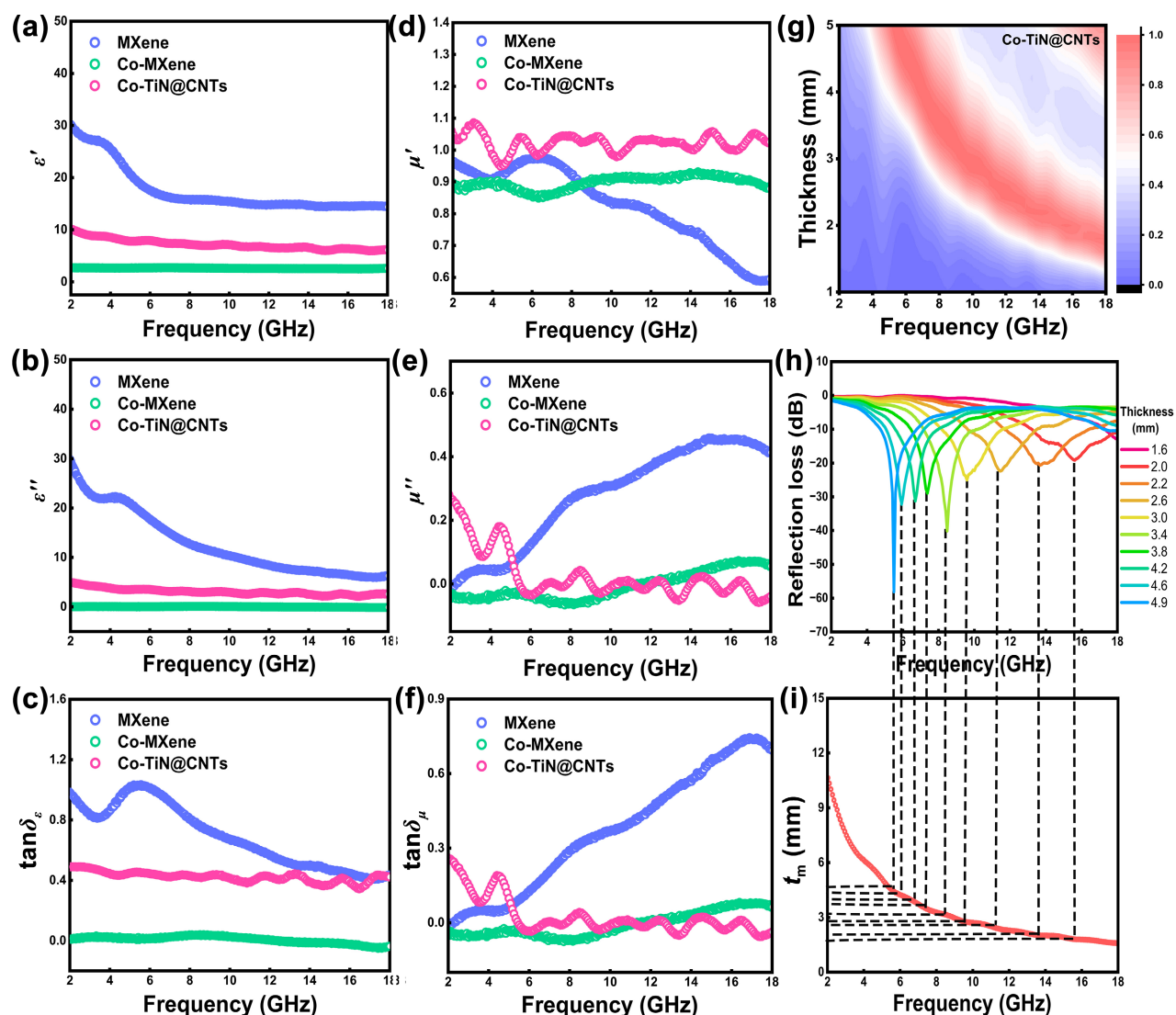


Figure 4 Electromagnetic parameters of Co-TiN@CNTs with a filler loading of 5% in the frequency range of 2–18 GHz: (a) ϵ' , (b) ϵ'' , (c) $\tan\delta_\epsilon$, (d) μ' , (e) μ'' , and (f) $\tan\delta_\mu$. (g) Two-dimensional projections of the (Z') and (Z'') values correlating the thickness and the frequencies. (h) The RL values at various thicknesses. (i) Theoretical matching thickness curves.

Combined with the observed multiple semicirculars in Cole–Cole plots (Fig. S10 in the ESM), the characteristic multippeak resonant response of $\tan\delta_\epsilon$ ($\tan\delta_\epsilon = \epsilon''/\epsilon'$) within 2–18 GHz (Fig. 4(c)) reveals the coexistence of multiple polarization relaxation mechanisms in Co-TiN@CNTs as follows: (1) Maxwell–Wagner interfacial polarization induced at the heterointerface between Co-TiN and surface CNTs, arising from the periodic charge accumulation under alternating electromagnetic fields; and (2) orientation relaxation of localized electric dipoles formed by crystal defects (potential N/Ti vacancies in Co-TiN) under electromagnetic excitation. These synergistic effects significantly enhance the dielectric loss capability of composites.

The μ' and μ'' components characterize a material's magnetic energy storage and dissipation capabilities, respectively. As shown in Figs. 4(d) and 4(e), the μ' values of Co-MXene and Co-TiN@CNTs range within 0.85–0.9 and 0.95–1.08, respectively, and that of MXene shows a quasi-linear decrease from 0.97 to 0.58. The relatively high μ' of Co-TiN@CNTs is due to the high coercivity from its uniformly dispersed small Co particles. Co-TiN@CNTs

manifest an anomalous $\mu'' < 0$ phenomenon in the 12–18 GHz band, arising from the diamagnetic responses triggered by its 3D interpenetrating network structure [31]. According to Lenz's law, this phenomenon occurs when the counter magnetic fields induced by eddy currents surpass the external field intensity. The evolution of magnetic loss tangent ($\tan\delta_\mu = \mu''/\mu'$) remains closely aligned with the μ'' trends, confirming the capability of Co-TiN@CNTs to achieve efficient magnetic energy conversion under low external field excitation (Fig. 4(f)) [13].

Impedance matching and attenuation capability are the core parameters for evaluating the performance of MA materials [32]. The former determines the ease of electromagnetic wave entry into the material, and the latter reflects the material's energy dissipation efficiency. According to impedance matching theory, interface reflection can be minimized when the normalized input impedance Z_{in}/Z_0 approaches 1. Frequency response analysis (Fig. 4(g) and Fig. S11 in the ESM) reveals that Co-TiN@CNTs exhibit excellent broadband impedance matching characteristics across 2–18 GHz and remarkable attenuation capability ($\alpha > 51.11$ dB/cm) due to the

stable convergence of the complex permittivity-to-permeability ratio toward free space impedance under multiple loss mechanisms. By contrast, MXene demonstrates excessive interface reflectivity caused by the severe imbalance between its dielectric and magnetic parameters, which fundamentally obstructs effective electromagnetic wave incidence (Fig. S12 in the ESM). Therefore, the *in situ* conversion of MXene into Co-TiN results in reduced electrical conductivity, and the confinement effect of CNTs enables the uniform dispersion of Co NPs within the system, leading to diminished magnetic properties. Consequently, Co-TiN@CNTs demonstrate exceptional impedance matching performance despite their relatively low M_s and electrical conductivity. As shown in Figs. 4(h) and 4(i), the experimentally measured thickness–frequency correlations (solid dots) closely align with the theoretical curve (dashed line), validating the destructive interference mechanism arising from the 180° phase difference between the incident and reflected waves at the interface and satisfying the $\lambda/4$ matching model [33].

Figures 5(a) and 5(b) systematically display the RL–frequency

response characteristics of MXene and Co-TiN@CNTs across a thickness range of 1.0–5.0 mm. MXene and Co-MXene fillers exhibit suboptimal wave-absorbing performance (Fig. 5(a) and Fig. S13 in the ESM). At a matching thickness of 4.8 mm, Co-TiN@CNTs achieve an $RL_{\min}$ of -58.17 dB at 5.505 GHz. At an ultrathin thickness of 2.0 mm, its EAB reaches 5.31 GHz (12.7–18 GHz), covering 93.6% of the Ku-band (12–18 GHz) and corresponding to an $RL_{\min}$ of -19.15 dB (Fig. 5(c)). Compared with other recently reported MA materials, Co-TiN@CNTs demonstrate unparalleled advantages in terms of filler loading and maximum absorption intensity (Fig. 5(f)).

The engineering application potential of the samples was evaluated through radar cross-section (RCS) simulations based on CST Microwave Studio. Figures 5(d) and 5(e) show the simulation model configured with electromagnetic waves incident along the $-Z$ axis, where θ denotes the incident angle variable. Three-dimensional RCS cloud maps demonstrate that Co-TiN@CNTs exhibit the lowest scattered signal intensity compared with the baseline metal plate (PEC), MXene and Co-MXene. Smith chart

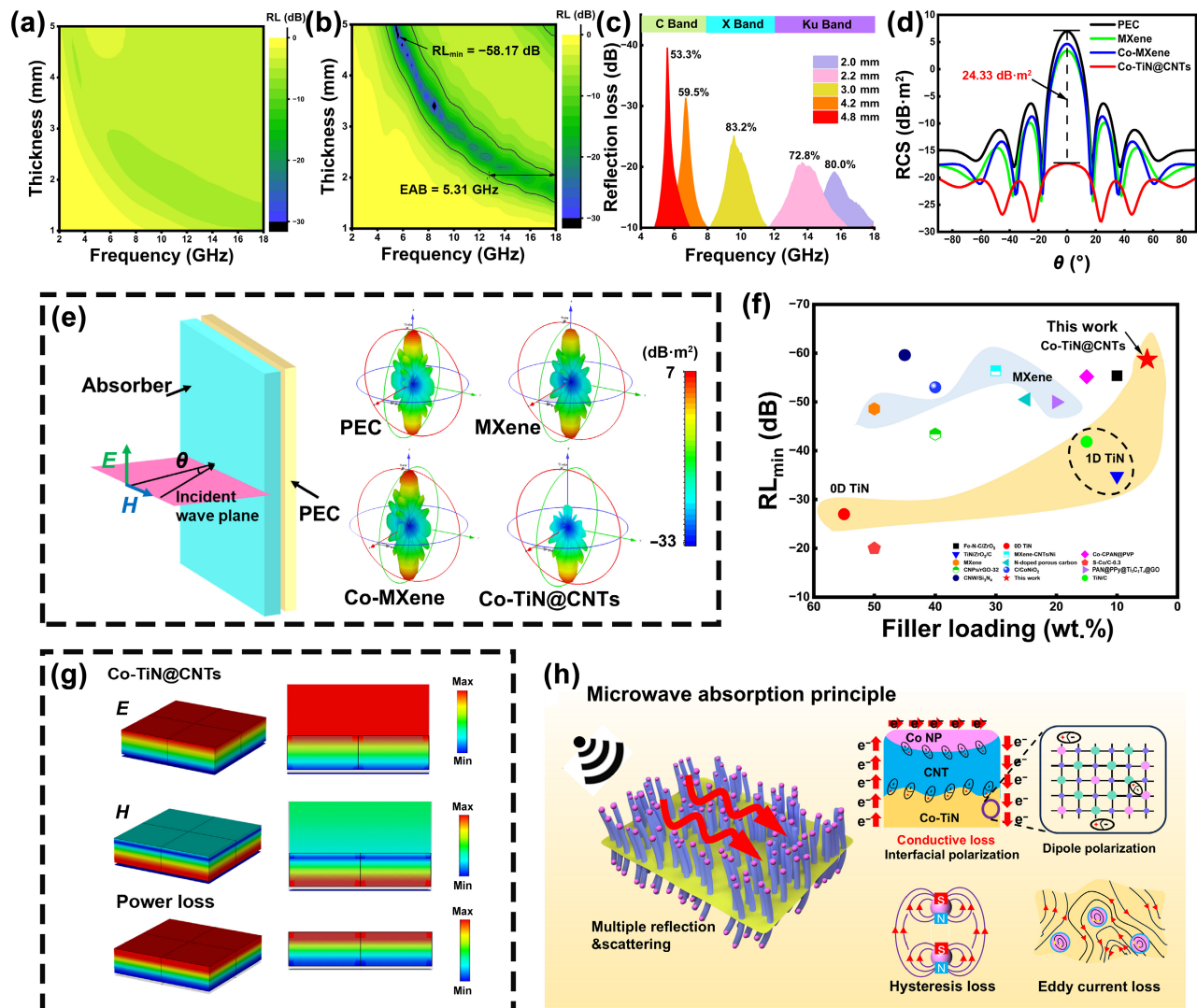


Figure 5 MA performance of Co-TiN@CNTs. The 2D RL values of (a) MXene and (b) Co-TiN@CNTs. (c) Radar cross-section data of the PEC, MXene, Co-MXene, and Co-TiN@CNTs. (d) Smith chart of samples. (e) RCS simulation plots of samples. (f) Comparison of the wave absorption properties of Co-TiN@CNTs with different low-frequency materials [6, 21, 24, 34–43]. (g) Distributions of the electric field, magnetic field, and power loss density of sample. (h) Schematic illustration of the absorption mechanism for Co-TiN@CNTs.

analysis (Fig. S14 in the ESM) confirms that the normalized real/imaginary impedance components of Co-TiN@CNTs converge toward the ideal matching point (1.0), significantly outperforming their MXene and Co-MXene counterparts [44]. In addition, electromagnetic field energy flux density visualization (Fig. 5(g) and Fig. S15 in the ESM) further reveals intense electric field vortices and magnetic field concentration effects within Co-TiN@CNTs, with dielectric loss dominating the electromagnetic energy dissipation mechanism. This finding is consistent with the pattern of the $\tan\delta_e$ and $\tan\delta_m$ curves (Fig. S16 in the ESM). The Co-TiN@CNTs exhibits outstanding microwave absorption performance, primarily due to the synergistic effects of multiple mechanisms (Fig. 5(h)). Within the material, highly conductive CNTs and TiN form a continuous conductive network, which effectively converts electromagnetic wave energy into electrical energy, resulting in significant conductive loss [45]. Simultaneously, the 3D CNTs networks prolong the propagation path of electromagnetic waves, enhancing multiple reflections and scattering, thereby improving dielectric loss [46]. In addition, the magnetic component of Co nanoparticles optimizes the impedance matching of the material, promoting the penetration of electromagnetic waves into its interior [26]. Furthermore, the numerous heterogeneous interfaces and structural defects introduced during the pyrolysis process further strengthen interface polarization and dipole polarization, contributing considerable polarization loss capability [47, 48].

4 Conclusion

By high-temperature topochemical transformation and CVD, we constructed a ternary synergistic structure Co-TiN@CNTs. This strategy not only effectively introduces the magnetic loss, but also synergistically enhances the dielectric loss and significantly improves the interfacial polarization and dipole polarization effects. The optimized absorber achieves exceptional performance: an $RL_{\min}$ of -58.17 dB at 5.505 GHz with 4.8 mm thickness and an ultrawide EAB of 5.31 GHz (12.7 – 18 GHz) at a sub- 2.0 mm thickness, covering 93.6% of the Ku-band. This work provides new ideas for the construction of multi-reflection networks.

Electronic Supplementary Material: Supplementary material (the XRD pattern, Zeta-potential, and TEM image of single layer MXene; XPS spectra, SEM images, XRD patterns, Raman spectra, TG curves, and DSC flow of different Co-TiN@CNTs; the Cole–Cole curve, the (Z') and (Z'') values, RL, Smith chart of MXene, Co-MXene, and Co-TiN@CNTs) is available in the online version of this article at <https://doi.org/10.26599/NR.2026.94908414>.

Data availability

All data needed to support the conclusions in the paper are presented in the manuscript and the Electronic Supplementary Material. Additional data related to this paper may be requested from the corresponding author upon request.

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Declaration of competing interest

All the contributing authors report no conflict of interests in this work.

Author contribution statement

Z. Y. J.: Data curation, writing manuscript, experimental design. B. T. T.: Project administration, validation, funding acquisition. W. T. W.: Project administration, validation. S. F. Z.: Project administration, validation, funding acquisition. Y. A. Z.: Project administration, funding acquisition, writing manuscript. All the authors have approved the final manuscript.

Use of AI statement

None.

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