

Rare-earth MOF composite amorphous-nanocrystalline FeSiB soft magnetic material for enhanced microwave absorption performance

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ABSTRACT: Using soft magnetic alloys as the matrix, the structural diversity of rare-earth metal–organic frameworks (RE-MOFs) provide a versatile platform for developing advanced electromagnetic wave (EMW) absorbing materials. In this study, an annealing-induced amorphous-nanocrystalline transformation in FeSiB alloys effectively relieves internal stress and stabilizes the magnetic matrix. Subsequently, a low-conductivity, wrinkled RE-MOF layer is constructed on the alloy surface to regulate dielectric behavior and optimize impedance matching. The abundant heterogeneous interfaces and defect sites introduced by the RE-MOF architecture facilitate interfacial polarization by providing charge accumulation centers at phase boundaries. Meanwhile, the hierarchical and rigid framework of the RE-MOF layer contributes to multiple internal reflections and scattering of incident electromagnetic waves, thereby extending the propagation path and enhancing energy dissipation. At a La-MOF loading of 5 wt.%, the composite exhibits a minimum reflection loss ($RL_{\min}$) of -63.72 dB (13.58 GHz, 1.82 mm), with a bandwidth of 2.93 GHz at $RL \leq -20$ dB for a thickness of 1.8 mm. Increasing the La-MOF content to 10 wt.% resulted in $RL_{\min}$ of -56.16 dB at 8.565 GHz with a matching thickness of 2.63 mm, while the corresponding absorption bandwidth decreases to 1.93 GHz (1.8 mm). These results indicate that the incorporation of rare-earth MOF into the amorphous-nanocrystalline FeSiB matrix effectively enhances microwave attenuation, which is closely associated with the folded MOF microstructure that facilitates repeated scattering and prolongs the propagation path of incident electromagnetic waves.

KEYWORDS: FeSiB amorphous-nanocrystalline, fancy ruffle-shaped La metal–organic framework (La-MOF), interfacial polarization and multiple reflection/scattering, impedance matching

1 Introduction

The rapid growth of high-frequency electronics and wireless communications has driven demand for lightweight, efficient electromagnetic wave (EMW) absorbers. Iron-based soft magnetic alloys are promising due to their high saturation magnetization and inherent magnetic loss [1, 2]. However, crystalline Fe alloys often suffer from imbalanced electromagnetic parameters, resulting in poor impedance matching and limited absorption [3]. Although

amorphous alloys reduce coercivity, their structural disorder often limits magnetic–dielectric synergy [4]. Therefore, simultaneously optimizing attenuation capability and impedance matching in Fe-based systems remains a fundamental challenge.

Notably, iron-based amorphous-nanocrystalline alloys have attracted increasing attention because they integrate the low coercivity of the amorphous phase with the high saturation magnetization of the crystalline phase [5–10]. Amorphous-nanocrystalline alloys represent a promising platform by integrating the low magnetic anisotropy characteristic of amorphous phases with the elevated saturation magnetization derived from nanocrystalline domains. Through controlled annealing, internal stresses can be effectively relieved, while enhanced exchange coupling between nanocrystals and the amorphous matrix leads to improved magnetic performance [4, 10]. Despite improved

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magnetic performance, broadband microwave absorption cannot be realized without coordinated dielectric regulation. The scarcity of heterogeneous interfaces and effective polarization centers restricts interfacial polarization processes, thereby compromising the balance between intrinsic attenuation capability and impedance matching [11]. Existing surface modification and carbonization strategies typically strengthen dielectric loss by elevating conductivity, yet such enhancement commonly disrupts impedance compatibility [12, 13].

In recent years, considerable attention has been focused on metal–organic frameworks (MOFs) in the field of EMW absorbing materials, owing to their distinctive highly ordered periodic structure, substantial specific surface area, and porosity [14–16]. In our previous work, we have prepared and investigated several transition metal (3d)-based MOFs and their derivative composites with some useful findings [17–23]. Compared with transition metals, rare earth (RE) elements are contracted by the lanthanide system, which makes them chemically similar, but the larger ionic radius can give them high coordination ability, which can form a series of structurally stable and novel RE-MOFs crystals by combining with organic ligands [24]. Li et al. [25] prepared straw-bundled RE-MOF-derived REO/C (REO = CeO₂, La₂O₃, Nd₂O₃) via co-precipitation and carbonization. The CeO₂/C-800 °C sample exhibited a minimum reflection loss (RL_{min}) of −58.7 dB and an effective absorption bandwidth (EAB) of 5.0 GHz at a thickness of 1.7 mm, which was attributed to multiple absorption and scattering mechanisms.

Despite recent progress in employing rare-earth elements to enhance electromagnetic wave absorption, effectively integrating RE-MOFs with magnetic alloys to achieve synergistic regulation of magnetic and dielectric responses remains challenging. Conventional coating or carbonization strategies typically increase dielectric loss by enhancing conductivity, but often deteriorate impedance matching and thus limit overall absorption performance. Moreover, rational and controllable interfacial designs capable of introducing multiphase interfaces and defect structures to strengthen interfacial polarization and multiploidization losses while preserving the intrinsic magnetic properties of soft magnetic alloys are still lacking. Therefore, developing an effective interface engineering strategy to simultaneously regulate magnetic and dielectric losses is crucial for high-performance electromagnetic wave absorbers. In this study, La-MOF was introduced onto FeSiB amorphous-nanocrystalline alloys to construct rare-earth-modified Fe-based soft magnetic composites. The resulting RE-MOF interfacial architecture provides abundant heterogeneous interfaces that promote interfacial polarization and electromagnetic wave scattering, thereby enhancing electromagnetic attenuation while maintaining favorable impedance matching.

2 Experimental

As shown in Fig. 1(a), the preparation route of FeSiB@La-MOF composites comprises the thermal treatment of amorphous FeSiB, La-MOF synthesis, and subsequent fabrication steps, which are described in detail below.

2.1 Heat treatment of FeSiB amorphous alloy powder

FeSiB amorphous alloy powder with a composition of Fe:Si:B:Nb:Cu = 73.5:16.5:6.3:1 (mass ratio) was prepared via the water atomization method. Subsequently, controlled heat treatment

(from room temperature to 550 °C at a heating rate of 5 °C/min, followed by a 2 h soak) was performed, yielding FeSiB amorphous soft magnetic alloy powder after coarse grinding [23]. The FeSiB amorphous material was placed in a tube furnace, and oxygen (1.0 L/min, 1 atm, 99.2%) was introduced. The furnace chamber temperature was raised from room temperature to 450 °C (heating rate 5 °C/min). After holding at this temperature for 2 h, the mixture was then cooled to room temperature to obtain the reddish-brown FeSiB powder, labeled as L₀.

2.2 Preparation of FeSiB@La-MOF

Heat-treated FeSiB amorphous alloy powder, La(NO₃)₃·6H₂O (99%), and pyromellitic acid (PMA, 98%) were uniformly dispersed in 40 mL of N,N-dimethylformamide (DMF, AR) by mechanical stirring for 30 min. The resulting suspension was transferred to a 100 mL polytetrafluoroethylene (PTFE)-lined autoclave and maintained at 120 °C for 24 h. After cooling to ambient temperature, the obtained product was collected by centrifugation, washed twice with DMF, and subsequently dried under vacuum at 60 °C for 24 h. The dried solid was ground to obtain gray-white FeSiB@La-MOF powders. By adjusting the relative amounts of the starting materials, FeSiB@La-MOF composites with FeSiB loadings of 5 wt.%, 10 wt.%, 15 wt.%, and 20 wt.% were prepared and denoted as L₁, L₂, L₃, and L₄, respectively. For comparison, pure La-MOF was synthesized following the same procedure.

2.3 Characterization

The crystalline structures were analyzed by X-ray diffraction (XRD, Cu K α , Panalytical Empyrean), and the carbon graphitization degree was assessed via Raman spectroscopy (Renishaw InVia). Morphology and microstructure were examined by scanning electron microscopy (SEM, MLA650F) and transmission electron microscopy (TEM, FEI Talos F200X), while surface composition and chemical states were determined by X-ray photoelectron spectroscopy (XPS, Thermo Fisher Nexsa). Specific surface area and pore size distribution were obtained from N₂ adsorption–desorption isotherms using the Brunauer–Emmett–Teller (BET) method (ASAP 2460). Complex permittivity and permeability (2–18 GHz) were measured with a vector network analyzer (CETC 3672B-S). For electromagnetic tests, samples were mixed with paraffin wax (mass ratio 3:2) and pressed into toroidal specimens (outer/inner diameters: 7.01/3.00 mm).

3 Results and discussion

Figures 1(c) and 1(d) present SEM images of pure La-MOF and sample L₂, illustrating their structural features. As seen in Fig. 1(c), the synthesized pure La-MOF exhibits a distinctive ruffle-like morphology. Figure 1(d) shows that the surface of sample L₂ contains more material compared to L₁, and the number of floral-like structures on the surfaces of samples L₃ and L₄ increases with the proportion of composite material (Figs. S1(a)–S1(d) in the Electronic Supplementary Material (ESM)). Notably, not all material grown on the alloy surface assumes the ruffle-like shape; some regions exhibit irregular spherical structures, likely due to surface stress. The surface stress interferes with the growth process, giving rise to local morphological nonuniformity. As evidenced by the line-scan profile collected from the surface region of sample L₂ (Fig. 1(e)), the concentrations of C, O, and La increase markedly at the outer layer and decrease sharply toward the interior. Consistent

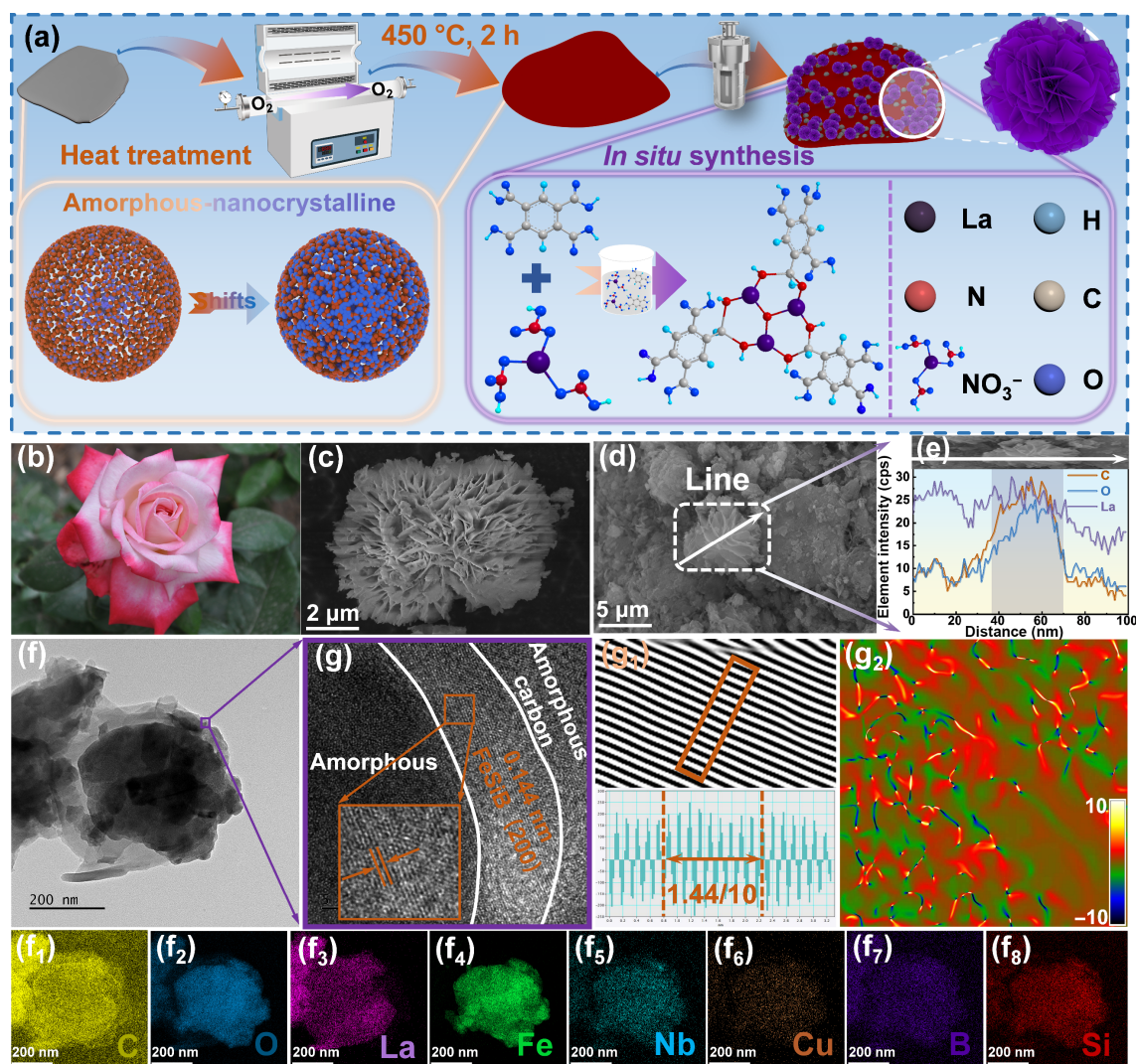


Figure 1 (a) Schematic of the synthesis process. (b) Optical image of moon-flower. SEM images of (c) pure La-MOF and (d) sample L_2 . (e) Line-scan profile of (d). (f) TEM image, (g) HRTEM image, and (g₁) corresponding fast Fourier transform (FFT) pattern of L_2 . (g₂) Strain map from HRTEM, with positive and negative values indicating tensile and compressive strain. (f₁)–(f₈) Elemental mapping images in the TEM field shown in (f).

with this observation, SEM image (Fig. 1(d)) shows that C, O, and La are predominantly localized within the surface layer, whereas the distributions of the remaining elements remain essentially unchanged throughout the matrix. These results collectively confirm the successful *in situ* growth of La-MOF on the alloy surface (Figs. S1(e)–S1(e₆) in the ESM).

To gain more insight into the surface microstructure, sample L_2 was examined using TEM. From the TEM image in Fig. 1(f), it is observed that there exists a black shaded region peripherally wrapped by a white fuzzy shaped structure. Elemental distribution analysis was carried out in the field of view of Fig. 1(f). As shown in the mapping images (Figs. 1(f₁)–1(f₈)), the distributions of C, O, and La elements align with the TEM morphology in Fig. 1(f), while Fe and Si elements correspond to the internal dark regions, indicating that La-MOF has been deposited on the surface of the FeSiB alloy. High-resolution TEM (HRTEM) analysis of its purple edge portion is shown in Fig. 1(g). The dark contrast region corresponds to the amorphous alloy, while lattice fringes are discernible at its periphery. Fourier transform analysis (Fig. 1(g₁)) identifies an interplanar spacing of 0.144 nm, which can be indexed to the (200) plane of FeSiB. In addition, a region of disordered carbon is present

in the upper-right area. To clarify the stress state of the alloy surface after La-MOF growth, a strain distribution map was constructed (Fig. 1(g₂)). The simulated map reveals pronounced spatial variations in strain caused by structural heterogeneity. In particular, the lower-right and central regions exhibit alternating red and green features, suggesting the formation of ordered nanocrystals within the amorphous matrix during heat treatment. The emergence of these nanocrystals contributes to stress relaxation by mitigating localized strain accumulation. In contrast, the lower-left region, which remains largely amorphous, exhibits extensive blue areas, reflecting sustained high local stresses. Elevated stress is observed in the amorphous carbon region at the upper-right side, which can be attributed to interfacial imperfections between the surface-deposited La-MOF and the substrate. These imperfections not only lead to localized stress accumulation but also act as effective polarization centers at the defective interfaces.

The samples were characterized by XRD. As observed from the XRD patterns in Fig. 2(a), the diffraction peaks of samples L_0 – L_4 at $2\theta = 45.194^\circ$, 64.827° , and 83.449° correspond to the (220), (400), and (422) crystallographic planes of the main phase FeSiB (JPCS No. 47-1114). Two characteristic diffraction peaks at $2\theta = 9^\circ$ – 11°

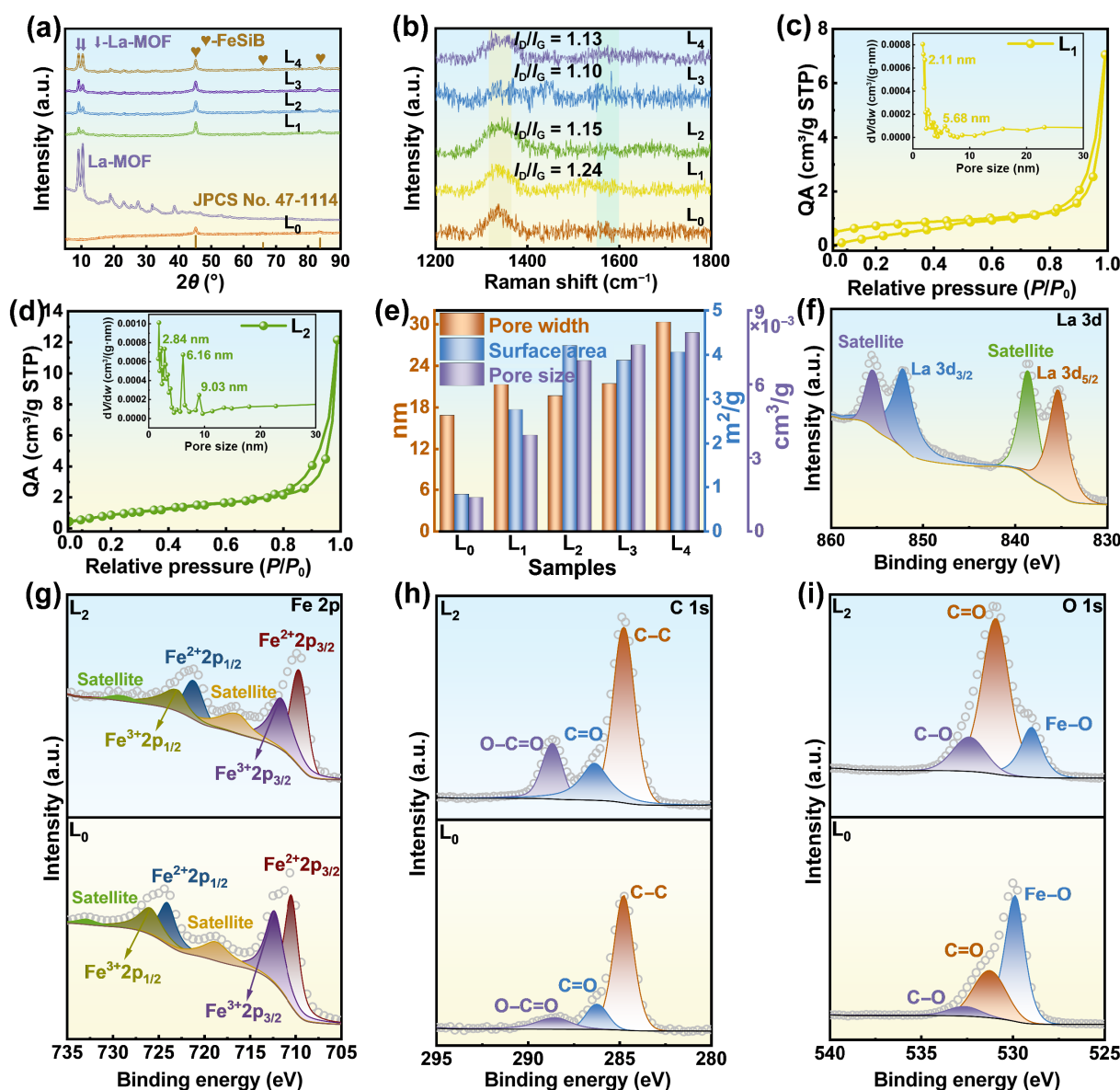


Figure 2 (a) XRD patterns and (b) Raman spectra of samples L_0 – L_4 . (c) Adsorption and (d) desorption curves of samples L_1 and L_2 for N_2 . (e) Histogram of average pore size, specific surface area, and porosity. (f) Fine La 3d spectrum of L_2 . Fine spectra of (g) Fe 2p, (h) C 1s, and (i) O 1s of samples L_0 and L_2 .

appeared in samples L_1 – L_4 after La-MOF incorporation, and their intensities increased gradually with higher La-MOF content. In order to prove that the diffraction peak is a characteristic peak of rare-earth La-MOF, pure La-MOF was prepared for comparison in this study. Pure La-MOF was found to coincide with the characteristic peaks appearing at $2\theta = 9^\circ$ – 11° for L_1 – L_4 . Raman spectroscopy was performed on all samples to assess the surface graphitization, as presented in Fig. 2(b). The peaks observed at 1350 and 1580 cm^{-1} correspond to the D and G bands, representing sp^3 -hybridized disordered carbon and sp^2 -hybridized graphitic carbon, respectively [26, 27]. The intensity ratio of the D band to the G band (I_D/I_G) is a reliable indicator of the degree of carbon ordering. It has been established that a smaller ratio indicates a higher degree of carbon ordering [28, 29]. Samples L_1 – L_3 have smaller I_D/I_G as the La-MOF content increases, indicating an increase in graphitic carbon and an increase in the ordered carbon content of the samples. The increase from L_3 to L_4 is attributed to the enhanced complexity of abundant heterogeneous interfaces and the presence

of more defects resulting from the increased MOF. It is worth noting that the G band is not obvious in Fig. 2(b), which is due to the fact that the sp^2 hybridization is present in the C–C bond, which generally occurs after graphitization, whereas the synthesized La-MOF has not been carbonized by heat treatment, and the C is still mainly in the organic framework/organic ligand, so the G band is not obvious [30, 31]. In Fig. 2(c) and Fig. S2(a) in the ESM, the increasing trend observed in samples L_1 and L_0 at low pressure indicates the presence of micropores, while the hysteresis curves at high pressure suggest the coexistence of macropores. Furthermore, as the composite content increases, the hysteresis phenomenon becomes increasingly pronounced across the relative pressure range for sample L_2 in Fig. 2(d) and samples L_3 and L_4 in Figs. S2(b) and S2(c) in the ESM [32, 33]. In addition, as observed from the insets of the Barrett–Joyner–Halenda (BJH) pore size distributions in Figs. 2(c) and 2(d), the pore size distribution of L_2 is mainly in the range of 2–10 nm, and the pore size distribution gradually increases with the gradual increase in the amount of the composite.

Consequently, sample L_4 was significantly larger than the other samples (Fig. S2(c) in the ESM). The histogram in Fig. 2(e) shows that the average pore size, specific surface area, and porosity show an increasing trend with the increase of La-MOF. This synergistic improvement of the structural features not only increases the interfacial polarizability but also extends the propagation path of electromagnetic waves within the material, thereby enhancing multiple reflection and scattering effects.

XPS analysis was conducted to investigate the surface chemical states of samples L_0 and L_2 . Notably, the full spectrum of L_2 displayed a La 3d signal at 850 eV, which did not appear in L_0 (Fig. S3(a) in the ESM). From the La 3d high-resolution spectrum of sample L_2 (Fig. 2(f)), the peaks at 835.3 and 852.1 eV can be assigned to La 3d_{5/2} and La 3d_{3/2}, respectively, while the adjacent higher-binding-energy peaks correspond to the satellite features [34]. As illustrated in Fig. 2(g), six peaks of Fe 2p are evident in the samples L_0 and L_2 , corresponding to Fe²⁺ 2p_{3/2} (710.4 eV), Fe³⁺ 2p_{3/2} (712.5 eV), Fe²⁺ 2p_{1/2} (724.1 eV), Fe³⁺ 2p_{1/2} (726.1 eV), and two satellite peaks [35, 36]. This suggests that Fe in the composite exists as metal oxides. Notably, no corresponding Fe oxide peaks are detected in the XRD patterns, likely because only a thin surface layer of oxides is present, insufficient to produce detectable diffraction intensity [37]. In addition, the composites exhibit characteristic C 1s peaks at 284.7 (C–C), 286.4 (C=O), and 288.5 eV (O–C=O) (Fig. 2(h)), indicating abundant carbon-containing functional groups that contribute to dipole polarization under the incident electromagnetic field [24, 38]. In Fig. 2(i), the three O 1s peaks in the 525–535 eV range correspond to iron oxides, C=O,

and C–O bonds [39, 40]. In addition, fine spectra of Si 2p, B 1s, Nb 3d, and Cu 2p were also plotted in this study (Figs. S3(b)–S3(e) in the ESM).

The EMW absorption performance depends on the electromagnetic properties. As Fig. 3(a) shows, the samples L_0 to L_3 exhibit decreasing ϵ' values with increasing frequency, in which ϵ' is the real part of the complex permittivity, representing the electric energy storage capability of the material. This reflects frequency dispersion caused by energy dissipation due to dipole polarization [41, 42]. Samples L_0 – L_4 showed a gradual decrease in ϵ' values as the amount of composite is increased. This is due to the fact that La-MOF is a metal–organic framework, which is a low-conductivity material compared to metal alloys, and the overall conductivity decreases as the content of low-conductivity materials increases [43]. Based on the free electron theory, higher conductivity leads to an increased dielectric constant [44], which explains the reduction in ϵ' observed for samples L_1 – L_4 relative to L_0 . In addition, the conductivity curves (Fig. S4 in the ESM) reveal a decreasing trend in conductivity for samples L_0 to L_4 , indicating a gradual reduction in their dielectric loss. Figure 3(b) displays the imaginary component, while Fig. 3(c) shows the tangent plots, both revealing multiple resonance peaks. This indicates the existence of several resonance behaviors within the composite materials. Among them, the resonance peaks of samples L_1 and L_3 are the most obvious, which is attributed to the shape anisotropy after the composite fancy ruffle-shaped RE-MOFs facilitate the charge accumulation effect, forming abundant dipoles, causing dipole polarization, and increasing the EMW dissipation [45]. The dipole polarization

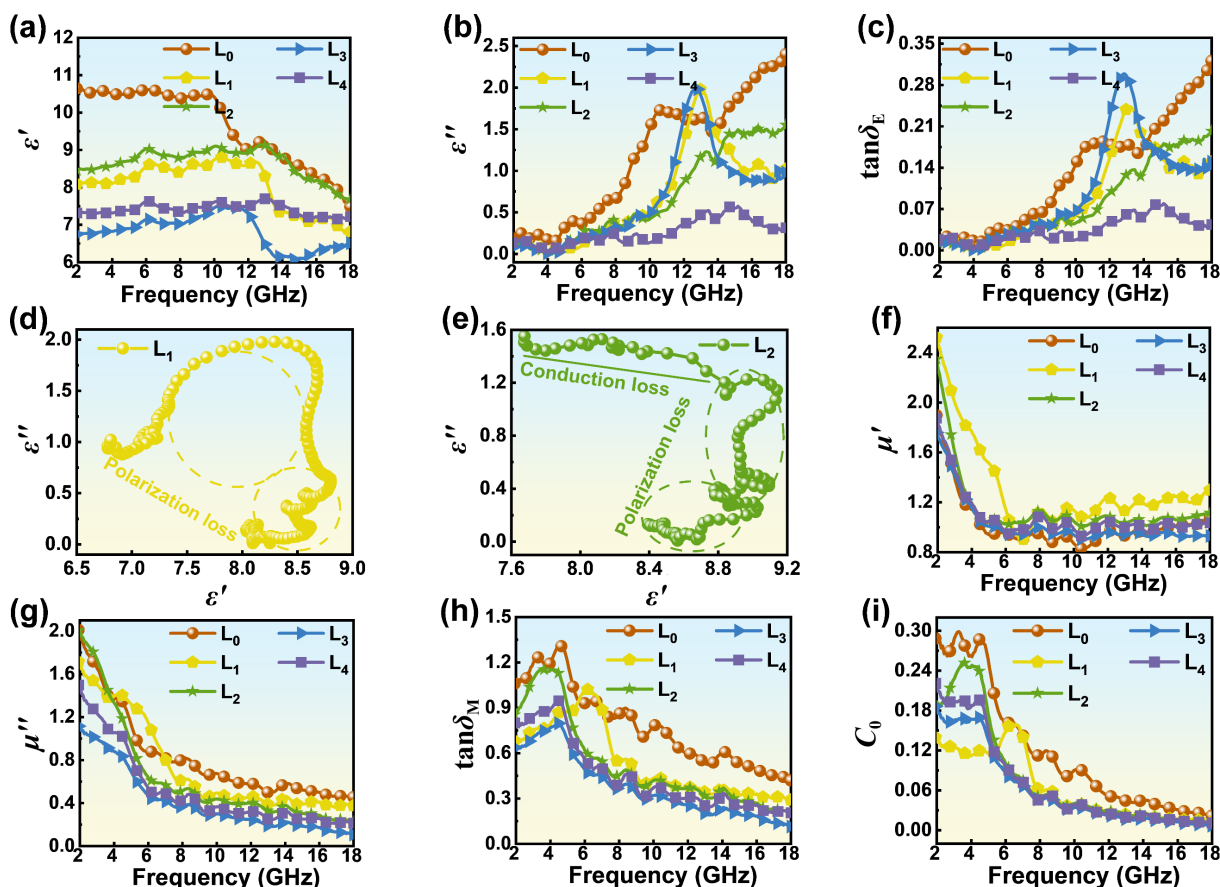


Figure 3 (a) Real part, (b) imaginary part, and (c) dielectric loss tangent of samples L_0 – L_4 . Cole–Cole plots for samples (d) L_1 and (e) L_2 . (f) Real part, (g) imaginary part, and (h) magnetic loss tangent of samples L_0 – L_4 . (i) Eddy current loss.

originates from the orientational polarization of the intrinsic functional groups within the materials with defect-induced local charge asymmetry. These structural non-uniformities can provide additional carriers to facilitate the polarization relaxation process by lowering the energy barriers; and interfacial polarization is essentially a space charge accumulation phenomenon triggered by the difference in conductivity/dielectric constants between the components at heterogeneous interfaces, which is in line with the Maxwell–Wagner–Sillars polarization model [46, 47]. To elucidate the wave-absorption mechanism of the composites, the Cole–Cole plot was employed to describe the polarization relaxation process, as expressed in Eqs. (1)–(3) [48]

$$\varepsilon' = \varepsilon_{\infty} + \frac{\varepsilon_s - \varepsilon_{\infty}}{1 + (\omega\tau)^2} \quad (1)$$

$$\varepsilon'' = \frac{\omega\tau(\varepsilon_s - \varepsilon_{\infty})}{1 + (\omega\tau)^2} \quad (2)$$

$$\left(\varepsilon' - \frac{\varepsilon_s + \varepsilon_{\infty}}{2}\right)^2 + (\varepsilon'')^2 = \left(\frac{\varepsilon_s - \varepsilon_{\infty}}{2}\right)^2 \quad (3)$$

where ε' is the real part of the complex permittivity, representing the electric energy storage capability of the material; ε'' is the imaginary part of the complex permittivity, representing the dielectric loss capability, i.e., the ability to dissipate electromagnetic energy into heat; ε_s is static permittivity, corresponding to the dielectric constant at low frequency or static electric field; ε_{∞} is relative permittivity at infinite frequency, representing the dielectric response at very high frequency; ω is the angular frequency of the electromagnetic wave, $\omega = 2\pi f$, in which f is the frequency of the electromagnetic wave; τ is relaxation time, indicating the response time required for dipoles to reorient under an alternating electromagnetic field.

As illustrated in Figs. 3(d) and 3(e), the dielectric responses of samples L_1 and L_2 exhibit characteristic Cole–Cole features, suggesting the coexistence of multiple polarization relaxation mechanisms under an alternating electric field. These relaxation processes, mainly associated with interfacial polarization and dipolar polarization, provide effective channels for electromagnetic energy dissipation and thereby contribute substantially to microwave attenuation [49]. Notably, the multiple distorted Debye semicircles observed in samples L_1 and L_2 indicate the occurrence of a Maxwell–Wagner relaxation process [50]. This is due to the introduction of fancy ruffle-shaped La-MOF, which causes multiple heterogeneous interfaces to be created in the composite materials, and the charge accumulates at the interfaces to form micro-network currents, causing interface-dipole polarization [51]. Figure 3(e) and Fig. S5(a) in the ESM reveal linear tails in samples L_0 and L_2 , suggesting that electromagnetic energy is partially dissipated through conduction losses caused by currents induced under the alternating field. With increasing La-MOF content, the polarization relaxation process of samples L_3 and L_4 is relatively weakened (Figs. S5(b) and S5(c) in the ESM).

As shown in Figs. 3(f) and 3(g), the μ' and μ'' values of samples L_0 – L_4 decrease with increasing frequency, which can be attributed to dispersion behavior arising from the natural resonance of ferromagnetic materials [52]. μ' represents a material's responsiveness to a magnetic field, that is, its ability to store magnetic energy. μ'' represents a material's ability to dissipate

magnetic energy as heat. It reflects the energy loss during the process of magnetic energy decay. The initial permeability of the samples decreases with the increase of La-MOF composite, which is due to the fact that the antimagnetic closed-shell layer 4f electrons of La^{3+} generate a weak antimagnetic moment making the La-MOF permeability lower than that of the FeSi-based alloy materials [53]; and the porous structure of La-MOF composite introduces residual stress at the interface, pinning down the wall motion of the magnetic domains [54]. In summary, it is observed from the magnetic loss tangent plot in Fig. 3(h) that the magnetic loss capability of the sample decreases after compounding La-MOF. Furthermore, magnetic losses in the 2–18 GHz range arise from exchange resonance, natural resonance, and eddy current effects. When magnetic loss is solely due to eddy currents, the eddy current loss parameter remains constant with frequency [55]. From the eddy current loss plot in Fig. 3(i), it is observed that the magnetic loss of the sample not only originates from the eddy current loss caused by C_0 , but also shows significant fluctuation with increasing frequency. C_0 is a criterion parameter used to determine whether eddy current losses are the primary source of magnetic losses. Since the value of C_0 remains nearly constant, this indicates that magnetic losses are primarily due to eddy current effects. The C_0 peaks generated within 2–6 GHz for all samples can be attributed to natural resonance since it usually occurs at lower frequencies. It is worth noting that the sample L_0 decreases gradually with increasing frequency, which is typical of magnetic natural resonance domination. For samples L_0 – L_4 , the magnetic loss shifts from being dominated by natural resonance to eddy currents in the 8–18 GHz range, indicating the predominance of eddy current loss at higher frequencies. Moreover, the rise in eddy current loss reduces the magnetic properties of the materials, so the incorporation of La-MOF leads to a decrease in their magnetic loss performance.

The RL and the EAB of the composite materials can be calculated based on line transmission theory and electromagnetic parameters [56, 57]

$$Z_{in} = \sqrt{\mu_r/\varepsilon_r} \tanh \left[j \left(\frac{2\pi f d}{c} \right) \sqrt{\mu_r \varepsilon_r} \right] \quad (4)$$

$$\text{RL (dB)} = 20 \lg \left| \frac{Z_{in} - 1}{Z_{in} + 1} \right| \quad (5)$$

in which Z_{in} is normalized input impedance, where μ_r ($\mu_r = \mu' - j\mu''$) and ε_r ($\varepsilon_r = \varepsilon' - j\varepsilon''$) are the complex permittivity and permeability of the materials; c is the light velocity in the free space; f is the frequency; and d is the thickness of the absorber. $\tanh$ is the hyperbolic tangent function, and j is the imaginary unit, representing the propagation characteristics of complex numbers. From Figs. 4(a) and 4(b), it is observed that for L_1 , when the d is 1.82 mm and f is 13.58 GHz, the $\text{RL}_{\min}$ reaches up to -63.72 dB; for L_2 , when the d is 2.63 mm and f is 8.565 GHz, the $\text{RL}_{\min}$ reaches up to -56.16 dB. However, the reflection losses of L_0 , L_3 , and L_4 became progressively inferior with the increase of the composite amount (Figs. S6(a)–S6(c) in the ESM). To further highlight the advantages of the introduced La-MOF, this study also compares the absorptive properties of the composite La-MOF after ball milling (Figs. S7(a)–S7(e) in the ESM). In comparison with amorphous-nanocrystalline composites, the amorphous composites exhibit inferior absorption performance. Moreover, they require thicker coating layers to achieve optimal reflection loss and demonstrate narrower absorption bandwidths.

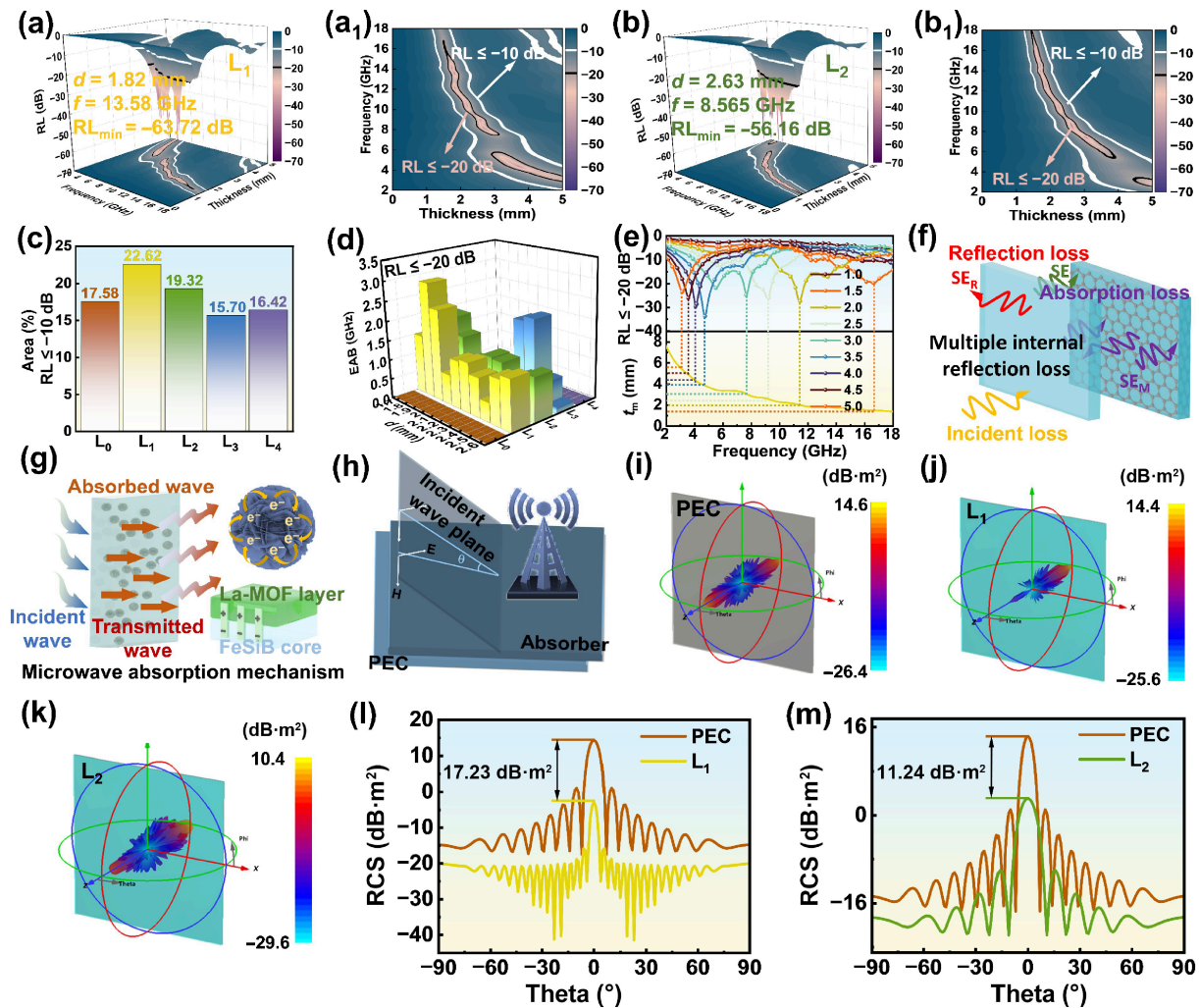


Figure 4 (a) 3D reflection loss plot and (a₁) top view of sample L₁. (b) 3D reflection loss plot and (b₁) top view of sample L₂. (c) Area share histogram for all samples with RL ≤ -10 dB in the range of 1–5 mm. (d) Width of the EAB with RL ≤ -20 dB in the range of $d = 1.7\text{--}2.6$ mm. (e) Plots of the relationship between RL and $\lambda/4$. (f) and (g) Illustrations of electromagnetic wave absorption mechanisms. (h) Simulated radar scattering pattern. Three-dimensional radar scattering patterns of (i) the PEC plate and samples (j) L₁ and (k) L₂. Simulated RCS curves of samples (l) L₁ and (m) L₂.

Based on the top views of RL in Figs. 4(a₁) and 4(b₁), we calculated the area share of all samples for RL ≤ -10 dB in the range of 1–5 mm (Fig. 4(c)) and the width for RL ≤ -20 dB in the range of $d = 1.7\text{--}2.6$ mm (Fig. 4(d)). For practical purposes, RL ≤ -10 dB is adopted as the criterion for EAB, corresponding to 90% electromagnetic wave attenuation, while RL ≤ -20 dB represents a stronger absorption level with approximately 99% attenuation [58, 59]. From Fig. 4(c), it can be observed that samples L₁ and L₂ have the largest percentage of RL ≤ -10 dB throughout the band and thickness, which are 22.62% and 19.32%, respectively. From Fig. 4(d), it can also be observed that the width of samples L₁ and L₂ are widest in the low thickness range for RL ≤ -20 dB, with bands of 2.93 and 1.93 GHz at $d = 1.8$ mm. The correlation between the matching thickness and the RL peak frequency can be interpreted using the $\lambda/4$ model [60]. The relationship between absorber thickness and matching frequency can be interpreted in terms of the quarter-wavelength condition. As shown in Fig. 4(e), the reflection-loss minima of sample L₁ shift toward higher frequencies with decreasing thickness, consistent with the expected inverse thickness–frequency trend. The experimentally determined matching thicknesses, denoted by inverted triangles, closely follow

the calculated $\lambda/4$ curves, indicating that the microwave absorption behavior of the composites is governed by the quarter-wavelength matching mechanism.

Schematic illustrations of the EMW absorption mechanism are presented in Figs. 4(f) and 4(g). As shown in Fig. 4(f), incident EMWs undergo repeated internal reflections and scattering within the material, during which part of the wave energy is dissipated. Figure 4(g) further details the absorption process in the FeSiB@La-MOF composite, where interfacial polarization, dipole relaxation, and magnetic resonance convert electromagnetic energy into heat, resulting in diminished reflection and enhanced absorption efficiency. The electronic stealth performance of the composites was assessed through radar cross-section (RCS) simulations of L₁ and L₂ samples coated on perfect electric conductor (PEC) substrates (Figs. 4(h)–4(k)). The 3D RCS maps (Figs. 4(i)–4(k)) indicate that FeSiB@La-MOF coatings significantly reduce reflected signals compared with bare PEC, demonstrating superior EMW attenuation. The corresponding 2D RCS distributions across angles from -90° to 90° are presented in Figs. 4(l) and 4(m), confirming the composites' effective electromagnetic shielding capability. At normal incidence, L₁ and L₂ show RCS values of 17.23 and

11.24 dB·m², respectively. The L₁ coating offers superior RCS reduction across all angles, with values consistently below −20 dB·m², indicating the potential of this composite for wide-angle stealth applications.

In electromagnetic wave absorption studies, α usually represents the attenuation constant, which describes the ability of a material to attenuate electromagnetic waves during propagation. In order to further represent the overall loss capability of the materials to electromagnetic waves, the α is introduced [61]

$$\alpha = \frac{\sqrt{2}\pi f}{c} \sqrt{(\mu''\epsilon'' - \mu'\epsilon') + \sqrt{(\mu''\epsilon'' - \mu'\epsilon')^2 + (\mu''\epsilon'' + \mu'\epsilon')^2}} \quad (6)$$

From the graph of Fig. 5(a), which reflects the variation of α with frequency, it is observed that the value of α gradually decreases for samples L₀–L₄, indicating that the overall attenuation ability of the materials for electromagnetic waves decreases. Notably, it is observed from the reflection loss plots in Fig. 4 that samples L₁ and L₂ have the strongest wave-absorbing properties and sample L₀ has the worst wave-absorbing properties, which is opposite to the variation of α values. To further elucidate the superior performance of samples L₁ and L₂, impedance matching is introduced, which can be expressed using a delta ($|\Delta|$) function [62, 63]

$$|\Delta| = \sin h^2(Kfd) - M \quad (7)$$

$$K = \frac{4\pi\sqrt{\mu'\epsilon'}\sin[(\delta_E + \delta_M)/2]}{c \times \cos \delta_E \cos \delta_M} \quad (8)$$

$M =$

$$\frac{4\mu'\cos\delta_E\epsilon'\cos\delta_M}{(\mu'\cos\delta_E - \epsilon'\cos\delta_M)^2 + \left[\tan\left(\frac{\delta_E - \delta_M}{2}\right)\right]^2} (\mu'\cos\delta_E - \epsilon'\cos\delta_M)^2 \quad (9)$$

where $\delta_E = \arctan(\epsilon''/\epsilon')$ and $\delta_M = \arctan(\mu''/\mu')$ are the dielectric and magnetic loss angles, and $|\Delta|$ represents the degree of impedance matching. A smaller $|\Delta|$ value indicates better adaptation of the material to the electromagnetic wave. In practice, the $|\Delta| \leq 0.4$ area is considered to be the effective impedance matching. The impedance matching plots of samples L₀–L₂ are calculated in Figs. 5(b)–5(d). It is observed from the percentage of area bar graphs of Fig. 5(e) that the percentage of area of samples L₀–L₄ progressively increased by 32.9% and 32.7% for samples L₃ and L₄ (Figs. S8(a) and S8(b) in the ESM). Additionally, impedance matching was analyzed based on normalized impedance for samples at $d = 1.82$ mm (Fig. S9 in the ESM). When $|Z_{in}/Z_0|$ approaches 1, effective impedance matching is generally considered achieved. In the theory of electromagnetic wave absorption and transmission lines, Z_0 typically represents the characteristic impedance of free space, also known as the air impedance. Results indicate that samples L₁ and L₂ exhibit the broadest frequency coverage within the range $|Z_{in}/Z_0| \approx 0.85$ –1.25, demonstrating optimal impedance matching performance at this thickness.

Although sample L₀ exhibits the highest attenuation constant and the strongest dielectric and magnetic losses, its effective impedance matching area is only 6.1%, which is insufficient to allow

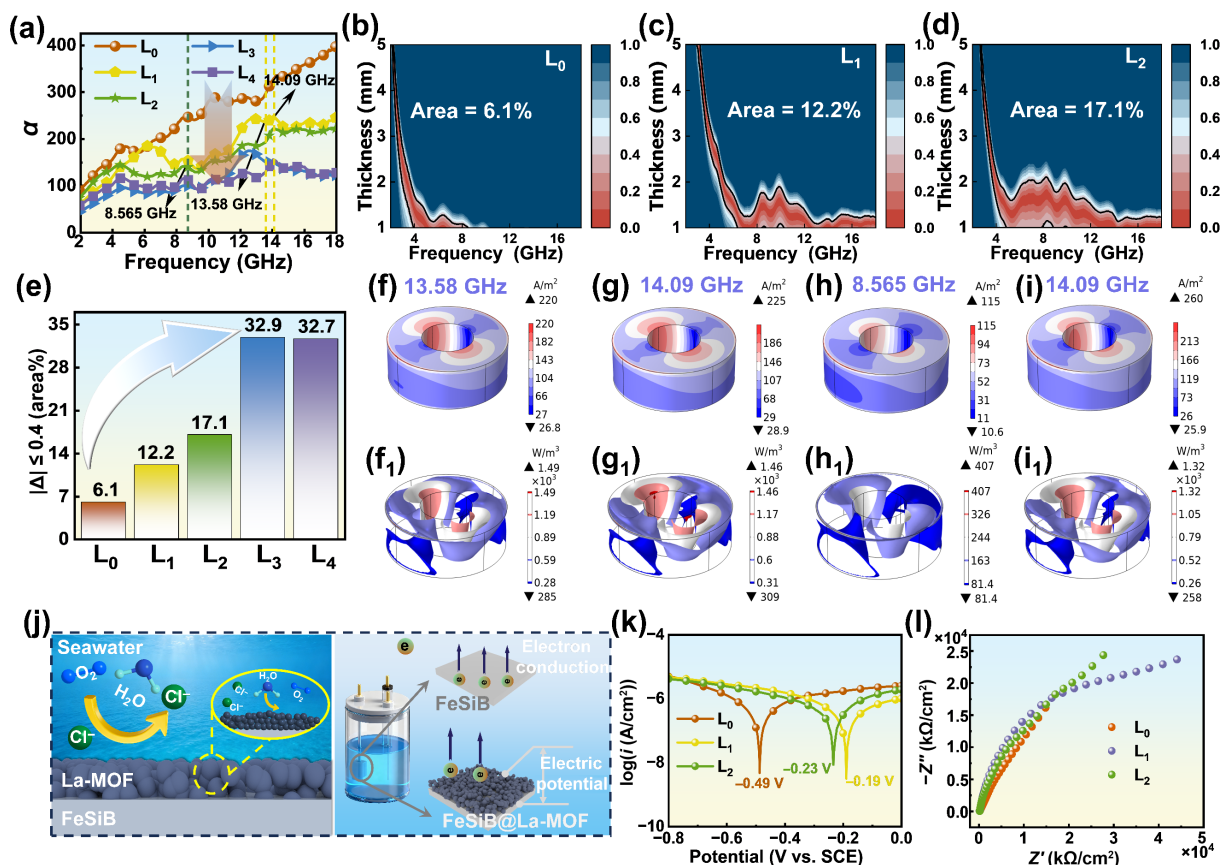


Figure 5 (a) Attenuation constant diagram for L₀–L₄. Impedance matching diagrams for samples (b) L₀, (c) L₁, and (d) L₂. (e) Percentage bar graph of the area of $|\Delta| \leq 0.4$ for L₀–L₄. (f)–(i) SCD. (f₁)–(i₁) VLD. (j) Simulated diagram of composite material corrosion in seawater. (k) and (l) Tafel and EIS curves for L₀–L₂.

electromagnetic waves to penetrate the material for effective dissipation, resulting in poor absorption performance. Samples L_3 and L_4 have the largest effective impedance areas (32.9% and 32.7%), but the smallest attenuation constants as well as the weakest dielectric and magnetic losses, which do not allow effective attenuation of the electromagnetic waves when they enter the interior of the materials. This eventually results in the poor wave-absorbing properties of L_3 and L_4 as well (Figs. S6(b) and S6(c) in the ESM). It is worth noting that attenuation constants and loss capabilities for the samples L_1 and L_2 are not the strongest, nor is the effective impedance area (12.2% and 17.1%) optimal. However, the optimal wave absorption results from a compromise between impedance matching and attenuation, allowing electromagnetic waves to enter the material and be efficiently dissipated. Accordingly, both attenuation capability and impedance matching must be simultaneously optimized to achieve the best absorption performance.

Using the optimal RL from Fig. 4 and the corresponding attenuation constant frequencies in Fig. 5(a), the surface current density (SCD) and volume loss density (VLD) of samples L_1 and L_2 were modeled at various frequencies via a coaxial circular geometry (Figs. 5(f)–5(i)). The SCD reflects the impedance matching behavior: Smaller SCD values are associated with expanded blue regions, signifying better impedance matching characteristics. The VLD is related to the loss capability per unit volume of wave-absorbing materials: The larger the VLD value, the larger the area of the pink region, and the stronger the loss [64, 65]. Figures 5(f)–5(i) show the SCD values of sample L_1 at $f = 13.58$ GHz, 14.09 GHz and sample L_2 at $f = 8.565$ GHz, 14.09 GHz, respectively. It can be observed in Fig. 5(f) that sample L_1 has the smallest SCD value of 26.8 A/m^2 at $f = 13.58$ GHz compared to Fig. 5(g) (14.09 GHz) where the area of the blue region is also more. In Fig. 5(h), the SCD value of sample L_2 at $f = 8.565$ GHz is even smaller, and is only 10.6 A/m^2 compared to Fig. 5(i) (14.09 GHz) where the area of the blue region is more. It is shown that samples L_1 and L_2 have optimal impedance matching characteristics at this frequency. From the VLD plots in Figs. 5(f)–5(i), it is observed that samples L_1 and L_2 have the largest area of pink region at the frequency of 14.09 GHz, where the loss is stronger, contrary to the SCD. Therefore, when regulating the electromagnetic parameters to enhance the wave-absorbing performance, impedance matching should be considered first and balanced with the attenuation constant. Additionally, finite element simulations were conducted on samples L_1 , L_3 , and L_4 (Figs. S10(a)–S10(f) in the ESM). Comparative analysis reveals that L_1 and L_2 achieve a superior synergistic balance between impedance matching characteristics and attenuation capabilities, exhibiting more balanced and stable electromagnetic response behavior.

To further evaluate the electrochemical performance of the composite materials in a seawater environment, a conventional three-electrode configuration (Fig. 5(j)) was employed to investigate both the amorphous-nanocrystalline FeSiB matrix and the corresponding composites. The corrosion potential indicates the onset of the corrosion process, while the corrosion current represents the rate of corrosion. Thus, materials exhibiting higher corrosion potentials and lower corrosion currents typically demonstrate superior anti-corrosion performance. As shown in Fig. 5(k), the Tafel curves were measured at a scan rate of 0.01 V/s , within a potential range of -0.8 to 0.0 V . Compared to the FeSiB amorphous-nanocrystalline materials, the 5 wt.% and 10 wt.% La-

MOF-loaded FeSiB composites display more positive corrosion potentials and lower corrosion rates [66, 67]. AC impedance testing was performed at an AC voltage of 5 mV within a frequency range of 0.1 to 10^5 Hz , as shown in Fig. 5(l). The arc length in an electrochemical impedance spectrum is a crucial indicator for evaluating corrosion resistance. A larger impedance arc radius corresponds to lower charging capacity and reduced susceptibility to corrosion, thereby serving as an effective corrosion barrier. It is evident that the La-doped FeSiB amorphous-nanocrystalline composite presents a higher impedance modulus, suggesting an increased charge-transfer resistance and consequently improved environmental adaptability.

Figure 6(a) and Table S1 in the ESM compare the reflection loss of different rare-earth ion composites, demonstrating that the MOF-based composite developed in this work achieves outstanding reflection loss at relatively low thicknesses [68–75]. The wave absorption mechanism of the composite material is shown in Fig. 6(b). The fancy ruffle-shaped La-MOF grown on the surface of the transition state alloys and the lamellar structure of the interlayer stacks promote multiple reflection/scattering of electromagnetic waves, leading to EMW attenuation. At the same time, the introduction of low-conductivity La-MOF improves the dielectric loss, and reasonable conductivity contributes to the microwave performance [76]. From the histogram of Fig. 2(e) it can also be concluded that the samples have large porosity and specific surface area which makes a large number of non-uniform interfaces between the MOFs and the alloys, increasing the interfacial polarization. At the same time, the dipole polarization generated by defects and functional groups in the composite material, together with the multipolar relaxation process, leads to an enhancement in the dielectric loss of the composite. In addition, the magnetic losses caused by the magnetic components further dissipate the EMW in the form of heat or other form of energy. The introduction of fancy ruffle-shaped MOFs material improves the multicomponent synergistic effect between the composites and optimizes the impedance matching, so that the incident wave enters into the interior of the materials as much as possible for attenuation.

4 Conclusions

In summary, FeSiB@La-MOF composites were successfully synthesized through a combined annealing and hydrothermal approach. The annealing of amorphous alloys drives the formation of amorphous-nanocrystalline structures, effectively relieving internal stress and concurrently improving soft magnetic performance. The introduction of pleated La-MOF materials effectively optimizes the balance between dielectric and magnetic losses, thereby improving impedance matching. Concurrently, La-MOF enhances interfacial polarization and multicomponent synergistic effects, while multiple reflections and scattering further amplify electromagnetic wave attenuation. At a loading of 5 wt.%, the $RL_{\min}$ reached -63.72 dB at 13.58 GHz with a matching thickness of 1.82 mm , while the EAB ($RL \leq -10 \text{ dB}$) reached 22.62% . At a loading of 10 wt.%, the $RL_{\min}$ reached -56.16 dB at 8.565 GHz with a matching thickness of 2.63 mm , and the EAB reached 19.32% . The $RL \leq -20 \text{ dB}$ bands were 2.93 and 1.93 GHz (1.8 mm). Additionally, Tafel and electrochemical impedance spectroscopy (EIS) results revealed that the incorporation of the rare-earth MOF shifted the corrosion potential to more positive values and enhanced electron transfer resistance in seawater,

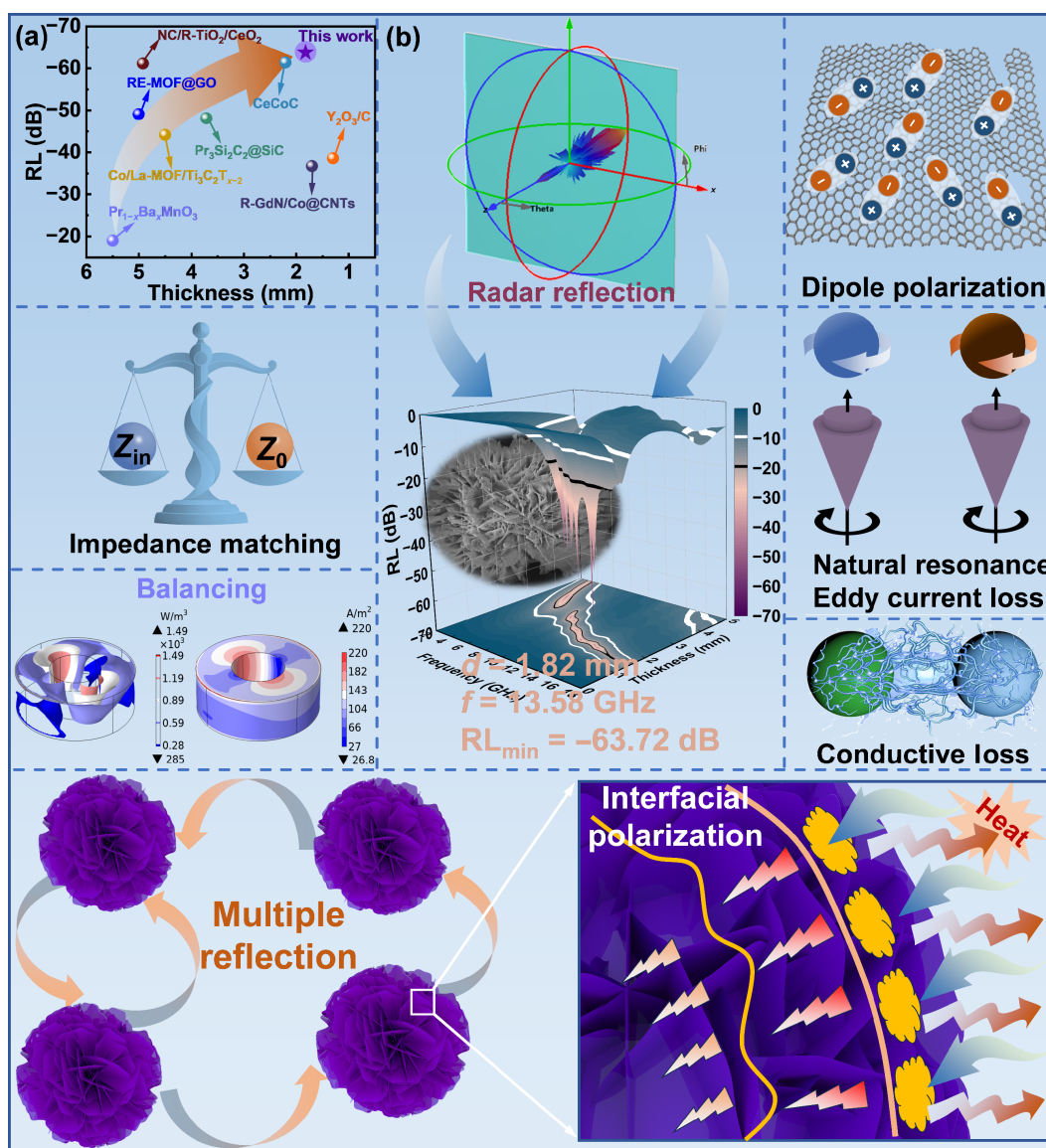


Figure 6 (a) Comparison of the properties of the rare earth ion composites. (b) Wave-absorbing mechanism of the composites.

thereby enhancing the environmental adaptability of composite materials. Nevertheless, challenges remain in terms of large-scale fabrication, long-term structural stability of the MOF layer, and precise regulation of magnetic–dielectric balance at higher MOF loadings. Future research should focus on interface engineering, pore structure optimization, and scalable synthesis approaches to further improve performance and application potential.

Electronic Supplementary Material: Supplementary material (morphological and elemental analysis, pore structure analysis, XPS results, electromagnetic parameter analysis, data on signal absorption performance, and a comparative table of electromagnetic wave absorption for samples with different concentrations) is available in the online version of this article at <https://doi.org/10.26599/NR.2026.94908646>.

Data availability

The data used to support the findings of this study are available 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

W. W. D.: Methodology, experiments, investigation, writing – original draft preparation. L. W.: Conceptualization, supervision, writing – review & editing, funding acquisition. S. H. T.: Experiments, data acquisition. H. X. and T. X. L.: Supervision.

Use of AI statement

None.

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