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Numerical Simulation of the Effects of Temperature and Porosity on Corrosion Behaviors of β -Li Phase in Mg-8Li Alloy

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Received: 14 April 2026; Accepted: 26 June 2026; Published: 23 July 2026

ABSTRACT: In this study, the effects of temperature and corrosion product porosity on the micro-galvanic corrosion behavior of the β -Li phase in Mg-8Li alloy are systematically investigated using COMSOL Multiphysics numerical simulations. A two-dimensional micro-galvanic corrosion model incorporating mass transport, electrochemical reactions, and level set-based interface tracking is established to simulate the corrosion evolution over 72 h under varying temperature and porosity levels. The results indicate that temperature can significantly accelerate the corrosion process and the exchange current density increases exponentially. As the temperature increases from 35°C to 55°C, the electrolyte potential shifts negatively, and the maximum electrode thickness change rises from 8.2 to 34.0 mm, indicating that the localized corrosion approximately doubles when the temperature increases by 10°C, and the peak local current density increases from 250 to 1100 A/m². When the corrosion product porosity increases from 3% to 4.5%, the corrosion current density increases from 60.2 to 162 A/m², and the thickness of the corrosion product layer increases from 0.5 to 2.0 mm. Simultaneously, under high porosity conditions, the current distribution becomes more uniform, and corrosion products form a more evenly distributed deposition layer, mitigating excessive localized corrosion. The combined effects of temperature and porosity significantly alter the interfacial ion transport and current density distribution, thereby governing the corrosion evolution path and interface morphology of the β -Li phase.

KEYWORDS: Mg-8Li alloy; temperature; porosity; numerical simulation

1 Introduction

Since the onset of the 20th century, rapid industrial development worldwide has resulted in increasingly severe environmental pollution. Particularly over the past two decades, the automotive industry has experienced rapid expansion, accompanied by a sustained increase in the global vehicle population. The air pollution caused by vehicle exhaust emissions has become a major environmental issue of global concern. The substitution of conventional components with lightweight materials to achieve vehicle lightweighting not only contributes to the reduction of exhaust emissions but also offers significant energy-saving potential. Consequently, the research and application of lightweight materials have garnered considerable interest from scientific institutions globally [1]. Magnesium is one of the lightest metal materials, with a density of 1.74 g/cm³, which is only two-thirds that of aluminum and 60%–75% that of conventional magnesium alloys [2]. Statistical analyses indicate that replacing conventional steel components with magnesium alloy

parts can achieve a weight reduction of over 100 kg by utilizing 40 kg of magnesium alloy components and reduce exhaust emissions by 10%, which is of great significance for environmental pollution control and energy conservation.

As the lightest metallic engineering material, magnesium-lithium alloys have attracted widespread attention. With a density of only 1.30–1.65 g/cm³, Mg-Li alloys are ideal lightweight materials [3–6], and have broad application prospects in military, aerospace, automotive and other fields, as well as continued development in civilian areas such as the 3C industry, biomedicine, and electrochemical power sources.

As shown in Fig. 1, when the lithium content is below 5.7 wt%, the alloy is a single-phase α -Mg with a hexagonal close-packed structure, when the lithium content exceeds 10.3 wt%, it transforms into a single-phase β -Li with a body-centered cubic structure, in the range of 5.7–10.3 wt%, a dual-phase $\alpha + \beta$ structure is formed [7–10]. In the dual-phase Mg-8Li alloy, the α -Mg phase is a Mg-rich solid solution with a hexagonal close-packed (hcp) structure, in which the Li content is typically below 5.7 wt%. The β -Li phase is a Li-rich solid solution with a body-centered cubic (bcc) structure, containing more than 10.3 wt.% Li. The α -Mg phase serves as the cathodic and the β -Li phase as the anodic in the micro-galvanic corrosion system. The continuous advancements in smelting technology and surface treatment processes, the 21st century has witnessed accelerated progress in the research and application of Mg-Li alloys. Although research on Mg-Li alloys in China began relatively late, substantial progress has been achieved in recent years [11–18].

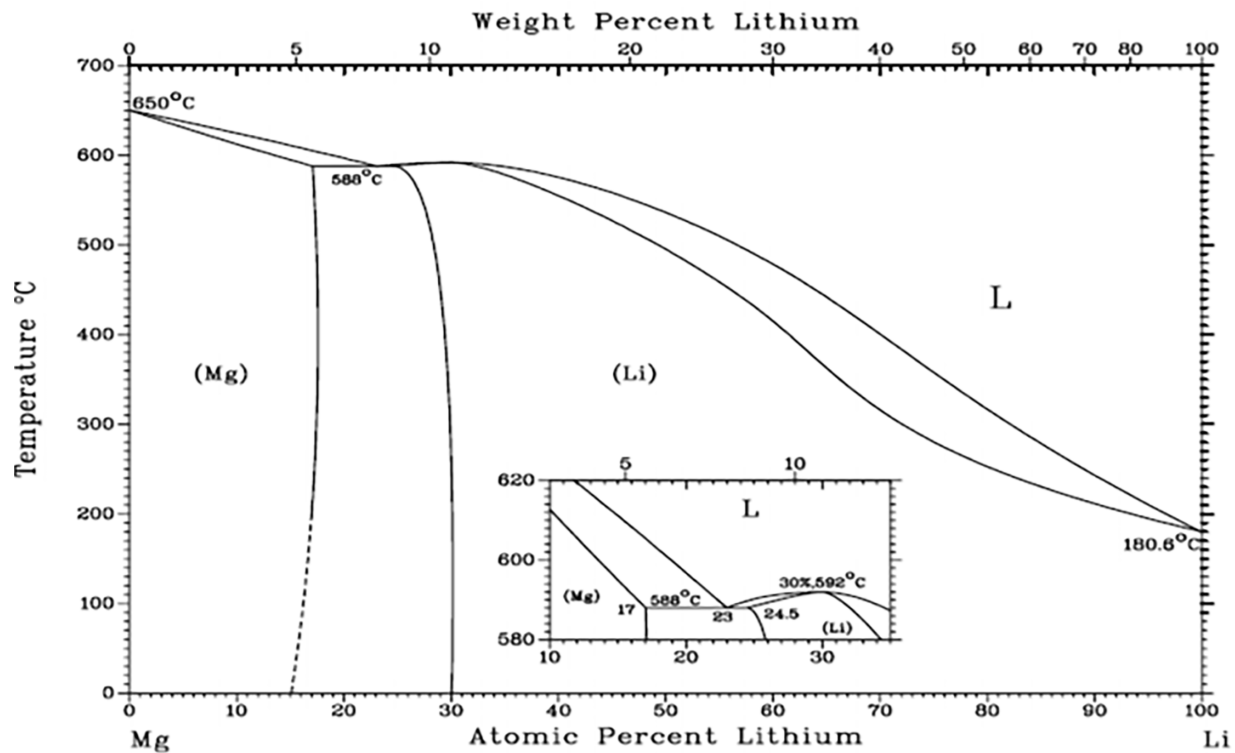


Figure 1: Binary phase diagram of magnesium-lithium alloys [19].

The corrosion mechanism of Mg-Li alloys is governed by their high chemical activity and microstructural characteristics. Although the addition of Li makes the alloy the lightest structural material, its standard electrode potential is extremely low (−3.04 V), making it more electrochemically active than magnesium (−2.37 V), rendering the alloy highly susceptible to corrosion in humid environments or chloride-containing

media [20]. Generally, the corrosion behaviors of Mg-Li alloys depend on various factors, including the alloy's chemical composition and microstructure, the physicochemical properties of the oxide film, and environmental conditions [21,22]. As an alloying element, Li content has a significant effect on the corrosion resistance of Mg-Li alloys. Huang et al. [23] found that the corrosion rate of the Mg-Li alloy was lower than that of pure Mg in 3.5 wt.% NaCl solution. Moreover, although pure Mg exhibited superior corrosion resistance than the Mg-Li alloy in the initial stage of corrosion, there also had smaller and shallower corrosion pits on its surface. Song et al. [24] investigated the electrochemical corrosion behaviors of Mg-8.8Li and AZ91D in 3.5 wt.% NaCl solution and found that the AZ91D alloy had superior corrosion resistance than the Mg-8.8Li alloy. However, as the immersion time extended (after 48 h), the mass loss rate of the Mg-8.8Li alloy decreased and became lower compared with that of the AZ91D alloy. This is attributed to the high activity of Li, which led to the rapid corrosion of the Mg-8.8Li alloy in the initial immersion stage. As corrosion progressed, the protective surface film on AZ91D was destroyed, leading to severe galvanic corrosion between the magnesium matrix and the cathodic secondary phases. In contrast, the Mg-8.8Li alloy gradually transformed from pitting and filiform corrosion to uniform corrosion. Song et al. [25] described the occurrence and characteristics of filiform corrosion in Mg-8Li alloys. Corrosion mainly occurs at the interface between the α -Mg phase and the β -Li phase and then extends into the β -Li phase. During filiform corrosion, the tip of the filament is acidic, promoting the dissolution of Mg, while the tail of the filament is passivated. The addition of Li promotes the hydrogen evolution and correspondingly accelerates the propagation of the filaments. Dong et al. found that the filiform corrosion first propagates into the α -Mg phase with filiform characteristics, whereas it manifests as intergranular corrosion in the β -Li phase [26].

Galvanic corrosion, also known as dissimilar corrosion or contact corrosion, refers to the phenomenon where two metal materials with different corrosion potentials form a circuit with the surrounding environment, causing the metal with a more positive potential to experience a reduced corrosion rate and become protected (cathodic protection effect), while the metal with a more negative potential undergoes an accelerated corrosion rate. Liu et al. [27] investigated the effect of heat treatment on corrosion behaviors of LA103Z Mg-Li alloy and found that after solution treatment, the galvanic corrosion tendency can be significantly reduced due to the disappearance of Al-Li phase. Corrosion behavior mainly manifested as filiform attack at the interface between the α -Mg and β -Li phases. In contrast, for the samples subjected only to high-temperature aging or solution treatment followed by aging, the Al-Li phase particles coarsened, and the number of precipitates increased, leading to a remarkably aggravation of localized corrosion.

With the continuous development and widespread adoption of computer technology, numerical simulation methods have become increasingly prevalent in solving engineering problems. Deshpande [28] used COMSOL Multiphysics to investigate the galvanic corrosion behavior of two phases in magnesium alloys. Assuming no concentration gradient and electroneutrality in the electrolyte solution, he employed the secondary current distribution and deformed geometry modules in COMSOL, utilizing a level set function to represent the cathode and anode, and simulated the corrosion behavior of magnesium alloy in a 1.6% NaCl solution. The study distinguished between two scenarios: continuous and discontinuous distribution of the β phase. The simulation results showed that as the proportion of the β phase increased, the peak anodic current density also increased, the average anodic current density for discontinuously distributed β phase was lower than that for continuously distributed cases, but the corrosion duration was longer. Deshpande [29] simulated the galvanic corrosion behavior of AE44 magnesium alloy coupled with mild steel and AA6063 aluminum alloy in 1.6% NaCl solution and found that the numerical simulation results were in good agreement with the full immersion results. Lee [30] conducted a numerical simulation of galvanic corrosion between zinc and iron materials under a thin electrolyte film. He assumed that both oxygen reduction and iron oxidation occurred on the iron surface, while only zinc oxidation occurred on

the zinc surface. The results showed that the current distribution on the iron surface was influenced by the thickness and conductivity of the electrolyte film. Yin et al. [31] used COMSOL Multiphysics to study the effect of intermetallic compounds on anodic and cathodic reactions in aluminum alloys. The results indicated that the blocking effect of corrosion products led to a gradual decrease in the corrosion rate.

Hu et al. [32] employed COMSOL to numerically simulate the corrosion behavior of metallic materials in a marine environment and revealed the underlying corrosion mechanisms. The results showed that cathodic regions can be protected and showed relatively low corrosion rates. These phenomena are mainly attributed to significant differences in dissolved oxygen content and temperature at various depths within the thermocline.

Wang et al. [33] developed an enhanced finite-element model to illustrate the micro-galvanic corrosion behaviors of aluminum alloy induced by intermetallic particles (IMPs). The results indicate that the $\text{Al}(\text{OH})_3$ deposited on the electrode surface suppresses further corrosion by lowering both the solution conductivity and the available electrochemically active area.

In Mg-based alloys, corrosion largely stems from the formation of galvanic couples between the different constituent phases. Thekkepat et al. [34] introduced a simple method based on the Born-Haber cycle for calculating the electrochemical potentials of intermetallic compounds and alloys. These calculated potentials can then be used to foresee and control the formation of galvanic cells, thus minimizing corrosion. The authors validated the model's effectiveness by adjusting the Zn composition to reduce galvanic corrosion in the Mg-3wt%Sr-xZn alloy. The proposed strategy is applicable to a wide range of alloy systems and is expected to facilitate the rational design of corrosion-resistant alloys.

As an ultra-light metal structural material, Mg-8Li alloy holds significant application prospects in aerospace, military equipment, and other fields. However, its severe micro-galvanic corrosion has consistently been a critical bottleneck restricting its service reliability. There is a significant electrochemical difference between the hcp-structured α -Mg phase and the bcc-structured β -Li phase in this alloy, forming a micro-galvanic corrosion system in corrosive media, in which the β -Li phase acts as the anode and preferentially dissolves, while the α -Mg phase serves as the cathode driving the corrosion process. The corrosion behavior of magnesium-lithium alloys is closely related to their unique dual-phase microstructure. Taking Mg-8Li alloy as an example, its microstructure consists of the hcp-structured α -Mg phase and the bcc-structured β -Li phase. Due to the significant potential difference between the two phases, a micro-galvanic corrosion system is formed in corrosive media. To deeply investigate the dynamic behavior of the β -Li phase in micro-galvanic corrosion, it is essential to clarify the coupling mechanism between environmental factors and the interfacial product layer. On one hand, temperature, as an environmental factor, directly regulates the current density and polarity reversal of the α -Mg/ β -Li micro-galvanic couple by altering the physicochemical properties of the electrolyte (such as ion diffusion rate and dissolved oxygen content) and the intrinsic activation energy of the electrochemical reactions. Under non-room temperature conditions, the dissolution rate of the β -Li phase may undergo order-of-magnitude changes, leading to a transition in corrosion morphology from uniform pitting to deep intergranular corrosion or exfoliation. During the corrosion process of the β -Li phase in Mg-8Li alloy, a surface product film mainly composed of Li_2CO_3 and possibly LiOH is formed. The porosity of this product determines its blocking efficiency for ion transport. A dense product layer with low porosity may inhibit the cathodic oxygen reduction reaction to a certain extent, thereby weakening the micro-galvanic effect and delaying the dissolution of the β -Li phase. In contrast, a loose product layer with high porosity not only fails to provide effective protection but can also form local "micro-concentration cells," continuously accelerating the anodic dissolution of the β -Li phase by maintaining high ionic conductivity, and may even lead to autocatalytic acidification beneath the

product layer. Currently, research on the corrosion behavior of magnesium-lithium alloys has mostly focused on composition optimization and surface modification, while a systematic understanding of the coupling mechanism of the two key variables—temperature and corrosion product porosity—on the β -Li phase in micro-galvanic corrosion is still lacking. Based on this, this paper focuses on Mg-8Li alloy to study the evolution of corrosion product porosity under different ambient temperatures and its influence mechanism on the micro-galvanic corrosion behavior of the β -Li phase.

2 Establishment of a Corrosion Simulation Model

2.1 Geometric Model

The simulation is performed on a two-dimensional domain of $0.01 \text{ m} \times 0.01 \text{ m}$, which represents the electrolyte solution region. The metal substrate (Mg-8Li alloy) is positioned at the bottom boundary ($y = 0$), consisting of the α -Mg (cathode) and β -Li (anode) phases. The geometry of the two-dimensional model is illustrated in Fig. 2. This model analyzes the micro-galvanic corrosion occurring between the α -Mg and β -Li phases, along with the surface deformation of the β -Li phase caused by dissolution reactions. It accounts for the mass transport of Li^+ , CO_3^{2-} and carbon dioxide, the precipitation reaction forming lithium carbonate (Li_2CO_3) corrosion products, and their subsequent effects on transport properties. Furthermore, the model employs a Level Set interface to track the corrosion product front, which separately defines the transport properties in the electrolyte and the corrosion product deposition regions [8]. To more accurately simulate the corrosion morphology and depth, a free mesh was separately created for the electrolyte and the electrode reaction interface. The mesh in the electrolyte region was predefined as “finer” while the mesh at the electrode reaction interface was predefined as “extremely fine”.

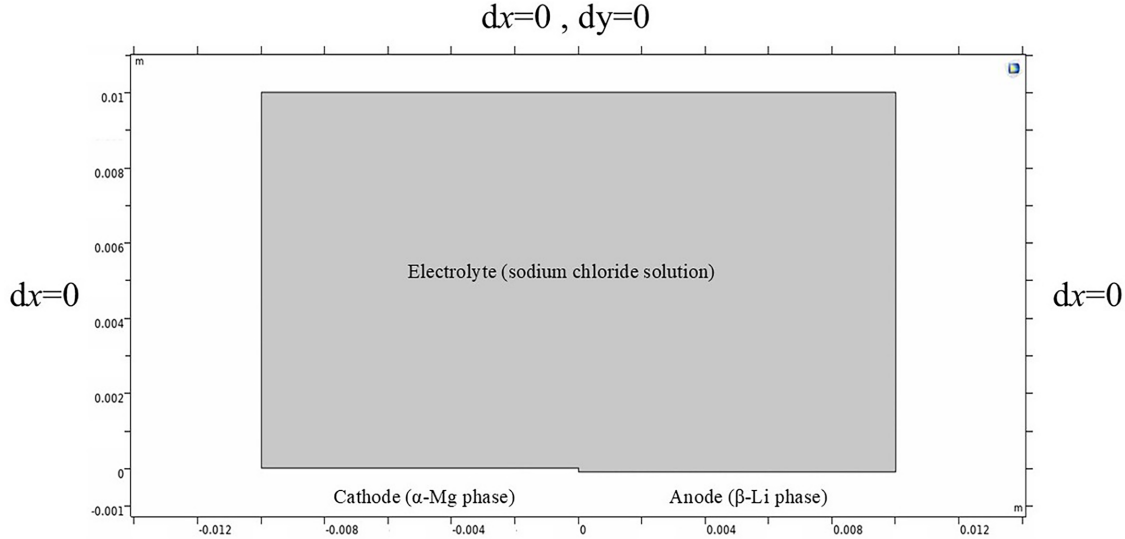


Figure 2: Schematic diagram of the 2D model.

The meshing procedure was as follows: (1) According to the simulation requirements, free meshes were created separately for the electrolyte and the electrode reaction interface. A free triangular mesh method was used for the solution domain. (2) Since the accuracy of the results is influenced by mesh density, the mesh resolution in the electrolyte region was defined as “finer.” (3) Balancing accuracy with computational cost, only the mesh at the electrode reaction interface was predefined as “extremely fine.” The specific parameters included a maximum element size of $6.77 \text{ m} \times 10^{-5} \text{ m}$, a minimum element size of $2.02 \text{ m} \times 10^{-7} \text{ m}$, a

maximum element growth rate of 1.05, a curvature factor of 0.2, and a narrow region resolution of 1. Mesh independence verification was performed by refining the mesh further, the changes in key outputs (e.g., maximum current density, electrode thickness change) were less than 5%, confirming that the current mesh density is sufficient to ensure reliable simulation results.

2.2 Mathematical Model

2.2.1 Electrochemical Reaction

The fundamental mechanisms of galvanic corrosion were investigated by simulating electrode kinetics, mass transfer, and the resulting geometric deformation. During the simulation, the overall mechanism of localized corrosion was assumed as follows.

The β -Li phase acts as the anode in the galvanic couple, and Li^+ ions are produced according to the following equation:



The α -Mg phase serves as the cathode of this galvanic couple, in which the oxygen reduction reaction is shown below:



2.2.2 Mass Transfer

Assuming diluted species in a supporting electrolyte, the mass transport of ionic species is described using the Transport of Diluted Species interface according to the following equation.

$$N_i = -D_{i,eff} \nabla c_i - z_i u_{i,eff} F c_i \nabla \phi_l \quad (3)$$

Here N_i represents the transport vector ($\text{mol} \cdot \text{m}^{-2} \cdot \text{s}$), $D_{i,eff}$ denotes the effective diffusion coefficient of the ionic species, c_i is the concentration in the electrolyte ($\text{mol} \cdot \text{m}^{-3}$), z_i is the charge number of the ionic species, $u_{i,eff}$ represents the effective mobility of the charged species ($\text{m}^2 \cdot \text{s}^{-1} \cdot \text{J}^{-1} \cdot \text{mole}^{-1}$), F is Faraday's constant ($\text{A} \cdot \text{s} \cdot \text{mole}^{-1}$), and ϕ_l is the potential in the electrolyte (V).

The material balance equation is

$$\frac{\partial c_i}{\partial t} + \nabla \cdot N_i = R_i \quad (4)$$

The transfer property is defined using the electrolyte volume fraction

$$D_{i,eff} = \varepsilon_l D_i + (1 - \varepsilon_l) D_{i,cp} \quad (5)$$

where ε_l is the level set variable used in Eq. (5), taking values of 1 in the electrolyte and 0 in the corrosion product region. $\varepsilon_l D_i$ is the diffusion coefficient of the ionic species in the electrolyte domain, while $D_{i,cp}$ is the diffusion coefficient in the corrosion product deposition region. The parameter ε_p in Eq. (6) is the porosity of the corrosion product layer, which directly affects the effective diffusivity within the porous deposit. Thus, the porosity effect is introduced through ε_p in Eq. (6).

Due to the porous nature of corrosion product deposits, the diffusion coefficient $D_{i,cp}$ needs to be updated to analyze the effect of porosity. It is defined based on the MacMullin number N_M and the porosity

ε_p of the corrosion product region. The effective transport properties within the deposition zone are defined by the following expression.

$$D_{i,cp} = D_i (1 - \varepsilon_p) \frac{D_i}{N_M} \varepsilon_p \quad (6)$$

where ε_p has a value of 4%, and N_M is defined by the following expression.

$$N_M = \frac{\tau}{\varepsilon_p} \quad (7)$$

Li^+ and OH^- Lithium ions react with hydroxide ions to form lithium hydroxide as shown below



The reaction producing lithium hydroxide is considered irreversible and proceeds when the ionic product $c_{\text{Li}^+} c_{\text{OH}^-}$ exceeds the solubility product constant k_{sp} ($\text{mol}^2 \cdot \text{m}^{-6}$) of lithium hydroxide. The reaction rate for the consumption of Li^+ and OH^- ions ($\text{mol} \cdot \text{m}^{-2} \cdot \text{s}^{-1}$) is described by the following expression.

$$R_{s,\text{Li}^+} = k (c_{\text{Li}^+} c_{\text{OH}^-} - k_{sp}) \times H(\zeta) \quad (9)$$

$$R_{s,\text{OH}^-} = k (c_{\text{Li}^+} c_{\text{OH}^-} - k_{sp}) \times H(\zeta) \quad (10)$$

The generated lithium hydroxide then reacts with carbon dioxide to form lithium carbonate, as shown below



Similarly, the reaction rate for the consumption of LiOH and CO_2 is described by the following expression.

$$R_{s,\text{LiOH}} = -2k (c_{\text{LiOH}}^2 c_{\text{CO}_2} - k_{eff}) \times H(\sigma) \quad (12)$$

$$R_{s,\text{CO}_2} = -k (c_{\text{LiOH}}^2 c_{\text{CO}_2} - k_{eff}) \times H(\sigma) \quad (13)$$

where is the reaction rate constant for the precipitation reaction ($\text{m}^4 \cdot \text{mol}^{-1} \cdot \text{s}^{-1}$), k_{eff} is a constant related to the solubility product of lithium carbonate, the equilibrium constant of carbon dioxide in water, and other relevant factors. $H(\zeta)$, $H(\sigma)$ is the Heaviside step function. ζ , σ is a step function variable defined by the following expression.

$$\zeta = \frac{c_{\text{Li}^+} \cdot c_{\text{OH}^-}}{k_{sp}} - 1 \quad (14)$$

$$\sigma = \frac{c_{\text{LiOH}}^2 \cdot c_{\text{CO}_2}}{k_{eff}} - 1 \quad (15)$$

The reaction source term is defined based on the consumption rate of LiOH and CO_2 as shown below.

$$R_i = -R_{s,i} \delta \quad (16)$$

Here, δ is the level set function used to specify the deposition reactions at the corrosion product interface. The local current density i_{loc} (A/m^2) on the electrode surface is determined by the electrochemical

reaction kinetics and the local overpotential, as described in the secondary current distribution module. The boundary conditions at the anode surface are given by:

$$-N_{\text{Li}^+} \cdot n = -\frac{i_{\text{loc}}}{F} \quad (17)$$

where n denotes the normal vector to the boundary. The boundary conditions on the cathode surface are:

$$-N_{\text{OH}^-} \cdot n = \frac{i_{\text{loc}}}{F} \quad (18)$$

The initial conditions and concentration at the upper boundary are set according to the following expression

$$c_{\text{Li}^+} = 0 \quad (19)$$

$$c_{\text{OH}^-} = c_0 \quad (20)$$

$$c_{\text{CO}_2} = c_i \quad (21)$$

Other boundaries are treated as insulating boundaries, with the general stretch operator enabling it

$$-N_i \cdot n = 0 \quad (22)$$

2.2.3 Corrosion Product Interface Tracking

The Level Set interface is used to track the deformation caused by corrosion product deposition. This interface automatically establishes the governing equation for the motion of the interface between the liquid electrolyte and the porous corrosion product. The interface is represented by the 0.5 contour of the level set variable. This variable varies from 1 in the electrolyte domain to 0 in the deposition region [35]. Therefore, the level set variable can be regarded as the electrolyte volume fraction, and its transport is governed by the following equation:

$$\frac{\partial \phi}{\partial t} + u \cdot \nabla \phi = \gamma \nabla \cdot \left(\varepsilon \nabla \phi - \phi (1 - \phi) \frac{\nabla \phi}{|\nabla \phi|} \right) \quad (23)$$

ε determines the thickness of the interface defined as $\varepsilon = h_{\text{max}}/4$, where h_{max} is the maximum mesh element size in the domain. γ determines the intensity of reinitialization, and a suitable value for it should be the maximum velocity magnitude present in the model.

The level set delta function is approximated as:

$$\delta = 6 |\phi (1 - \phi)| |\nabla \phi| \quad (24)$$

As lithium carbonate corrosion products deposit on the electrode surface, the lithium surface also deforms due to the dissolution reaction. The velocity field used in the level set variable transport equation must account for the following two components: the velocity representing the deposition of lithium carbonate corrosion products u_{dep} ($\text{m} \cdot \text{s}^{-1}$) and the velocity representing magnesium dissolution u_{corr} ($\text{m} \cdot \text{s}^{-1}$). Eq. (23) is the conservative level set equation, and Eq. (24) is the smoothed delta function used to localize reactions at the interface.

The deposition velocity of lithium carbonate corrosion products in the normal direction is defined as follows:

$$u_{\text{dep}} = \frac{1}{2} R_{s, \text{LiOH}} \frac{M_{\text{Li}_2\text{CO}_3}}{\rho_{\text{Li}_2\text{CO}_3}} \quad (25)$$

where $M_{\text{Li}_2\text{CO}_3}$ is the molar mass of 73.89 ($\text{g}\cdot\text{mol}^{-1}$), $\rho_{\text{Li}_2\text{CO}_3}$ is the density of 2110 ($\text{kg}\cdot\text{m}^{-3}$).

The dissolution velocity of lithium in the normal direction is defined as follows:

$$u_{\text{corr}} = \frac{i_{\text{loc}} \cdot M_{\text{Li}}}{F \cdot \rho_{\text{Li}}} \quad (26)$$

where M_{Li} is the molar mass of 6.941 ($\text{g}\cdot\text{mol}^{-1}$), ρ_{Li} is the density of 534 ($\text{kg}\cdot\text{m}^{-3}$). Due to u_{corr} is only used at the deforming boundary, therefore, we employ a generalized stretch operator to make it available across the entire domain.

If the deposits form a parallel vertical rod-like porous structure, only the y-component of the dissolution velocity is used in the level set transport equation. Therefore, the velocity field employed in the transport equation for the level set variable is defined according to the following expression.

$$u = u_{\text{dep}} \times n_{\text{LS},x} \times H(0.6 - \phi) \quad (27)$$

$$v = (u_{\text{dep}} \times n_{\text{LS},y} + u_{\text{corr}} \times n_y) \times H(0.6 - \phi) \quad (28)$$

Here, $n_{\text{LS},x}$ and $n_{\text{LS},y}$ are respectively the x-component and y-component of the normal vector to the interface, which is defined by the level set variable as follows:

$$n_{\text{LS}} = \frac{\nabla \phi}{|\nabla \phi|} \quad (29)$$

n_y is the y-component of the unit normal vector, pointing outward from the electrolyte domain. Heaviside step function $H(0.6 - \phi)$ ensures that the level set variable is advected only near the corrosion product region. A region with a level set value of zero enters the domain from the bottommost boundary, which is specified using an inlet boundary condition, while the remaining boundaries are specified using outlet boundary conditions.

2.2.4 Electrolyte Properties

Finally, the effective electrolyte conductivity is defined separately for the electrolyte domain and the porous corrosion product region using the electrolyte volume fraction, which is defined based on the level set variable as follows:

$$\sigma_{\text{eff}} = \varepsilon_l \sigma + (1 - \varepsilon_l) \sigma_{\text{cp}} \quad (30)$$

Here, σ is the electrolyte conductivity in the electrolyte domain, which is equal to 2.5 ($\text{S}\cdot\text{m}^{-1}$). Electrolyte conductivity σ_{cp} in the corrosion product deposition region is described by the following expression.

$$\sigma_{\text{cp}} = s_L^m \varepsilon_p^n \sigma \quad (31)$$

where s_L is the fluid saturation with a value of 1, m is the cementation exponent with a value of 1, and n is the saturation exponent, which has a value of 2 in this model. The model is solved in a transient study, simulating corrosion after three days of immersion in a NaCl solution. In the present model, the electric

double layer (EDL) at each electrode/electrolyte interface is not explicitly resolved because its characteristic thickness (Debye length, typically <100 nm) is much smaller than the computational mesh size (~ 10 μm). Instead, the kinetic effects of the EDL are incorporated into the phenomenological parameters of the electrochemical reactions, such as the exchange current density and anodic/cathodic Tafel slopes, following the Butler-Volmer formulation.

3 Results and Discussion

3.1 Effect of Temperature on the Corrosion of β -Li Phase

3.1.1 Effect of Temperature on Corrosion Potential

Fig. 3 shows the electrolyte potential, current density distribution, total electrode thickness change, corrosion product interface, and local current density at 72 h. The electrolyte potential increases uniformly from the cathode toward the anode. Since no additional external potential is applied to the electrode, the electrode potential is given by $\varphi = -\varphi_1$. From $x = 0$ to $x = 0.1$, the potential in the anodic region ranges from -2.16 to -2.24 V, while from $x = -0.01$ to $x = 0$, the potential in the cathodic region ranges from -2.09 to -2.16 V. Because the current density is highest at the metal contact point, metal dissolution also reaches its maximum there. The electrolyte current density spreads out in a fan shape at the interface between the cathode and anode. The closer to the contact point, the higher the electrode current density, and the electrode thickness change is also concentrated at the metal contact point. Currently, the total change in electrode thickness is approximately 4 mm. In the corrosion product interface diagram, regions where the volume fraction of fluid 1 equals 1 indicate the deposition of corrosion products, while regions where it equals 0 represent the electrolyte. The contour line for a fluid 1 volume fraction of 0.5 represents the position of the corrosion product interface at the end of the simulation. Corrosion product deposition is greatest at the metal contact point and decreases with increasing distance along the metal surface from the contact point. It can also be observed from the local current density diagram that the current density at the interface between the cathodic and anodic regions is significantly higher than in other areas and exhibits a radial distribution, which is consistent with the pattern shown in the electrolyte current density diagram.

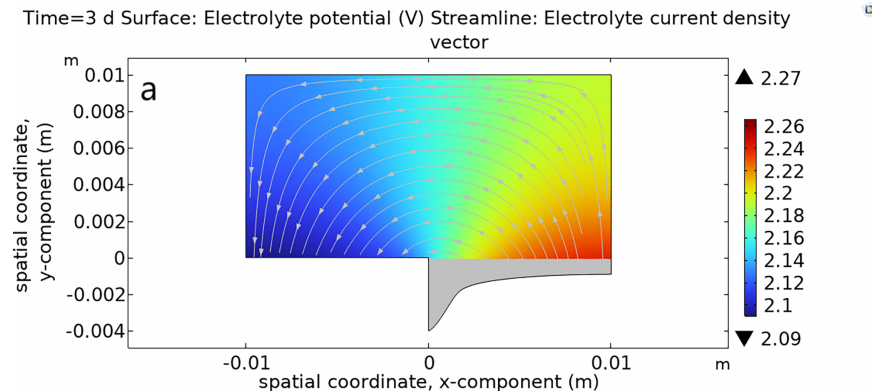


Figure 3: (Continued)

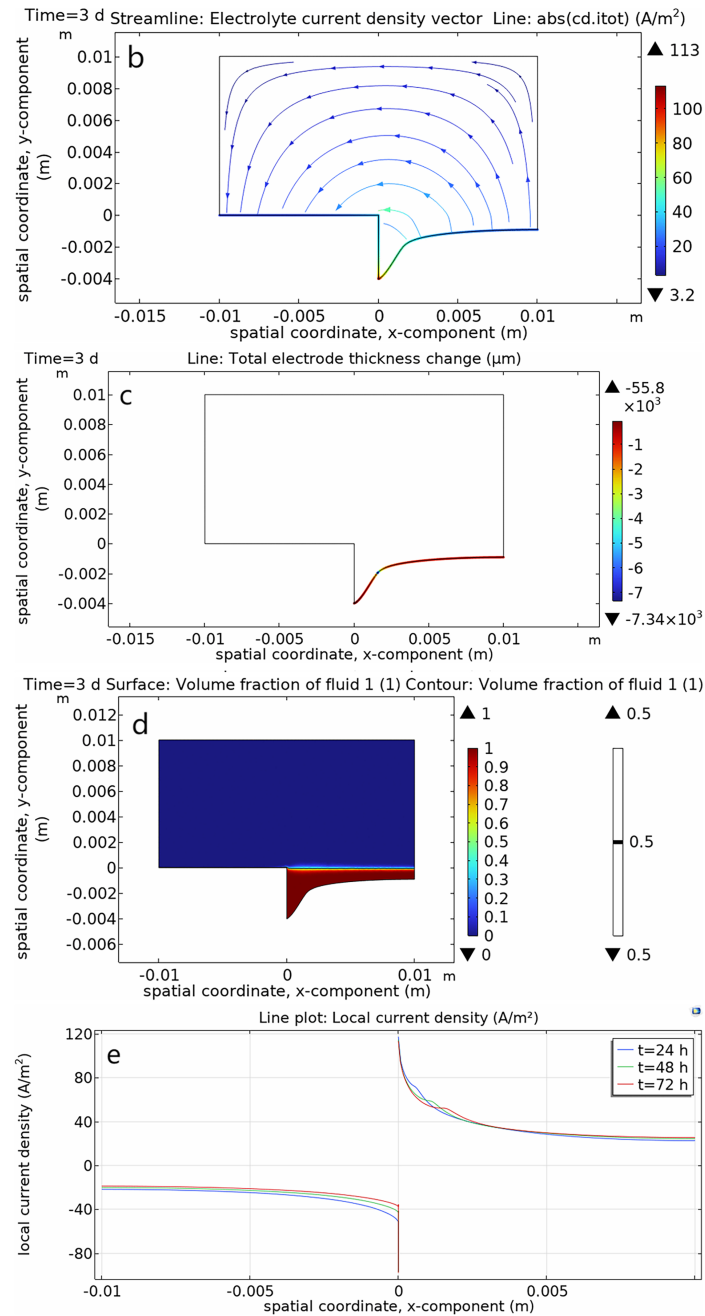


Figure 3: Simulation results at 25°C after 72 h (corrosion product porosity 4%): (a) electrolyte potential (V), (b) electrolyte current density (A/m^2), (c) total electrode thickness change (mm), (d) corrosion product interface represented by the volume fraction of fluid 1, where “fluid 1” denotes the liquid electrolyte, the value is 1 in the electrolyte domain, 0 in the porous corrosion product region, and the 0.5 contour marks the moving interface, (e) local current density (A/m^2).

As shown in Fig. 4, under the condition of 35°C, the minimum central potential after 72 h of operation is approximately 2.04 V, the potential distribution is relatively uniform, and the current density streamlines are sparse and widely dispersed. When the temperature increases to 45°C, the central potential decreases to about 1.98 V, the curvature of the equipotential lines becomes more pronounced, and the streamlines

become clearly concentrated toward the center. As the temperature further rises to 55°C, the central potential further drops to approximately 1.93 V, the potential depression becomes most significant, and the streamlines are highly concentrated, indicating a notable increase in local current density. These observations suggest that the increase in temperature, together with the rise in exchange current density, collectively intensifies the degree of polarization in the system. According to the Nernst equation (Eq. (32)) and the temperature correction for the standard potential (Eq. (33)), when the temperature increases from 35°C to 55°C ($\Delta T = 20$ K), the change in equilibrium potential is approximately ± 2 to ± 20 mV, which is far smaller than the actual change in the central potential (approximately 90 mV).

$$E_{eq} = E^0 + \frac{RT}{nF} \ln Q \quad (32)$$

$$E^0(T) = E^0(298.15) + \frac{\Delta S^0}{nF} (T - 298.15) \quad (33)$$

E_{eq} —Balanced electrode potential (V), E^0 —Standard electrode potential, (V) R —Ideal gas constant, T —temperature (K), n —Number of electron transfers, F —Faraday constant, Q —Reaction entropy, ΔS^0 —Standard reaction entropy change

Therefore, the temperature effect on the equilibrium potential can be neglected, and the change in the actual potential distribution mainly originates from the alteration of the polarization overpotential. Temperature exponentially increases the exchange current density through the Arrhenius relationship, leading to an amplification of local reaction rate differences, current concentration, and a significant enhancement of activation polarization. Under the condition of 55°C, an obvious potential depression already appears at 24 h, and the depth of the depression at 72 h is far greater than that under the 35°C condition. This indicates that the elevated temperature accelerates the development of polarization, which is consistent with the exponential increase in exchange current density.

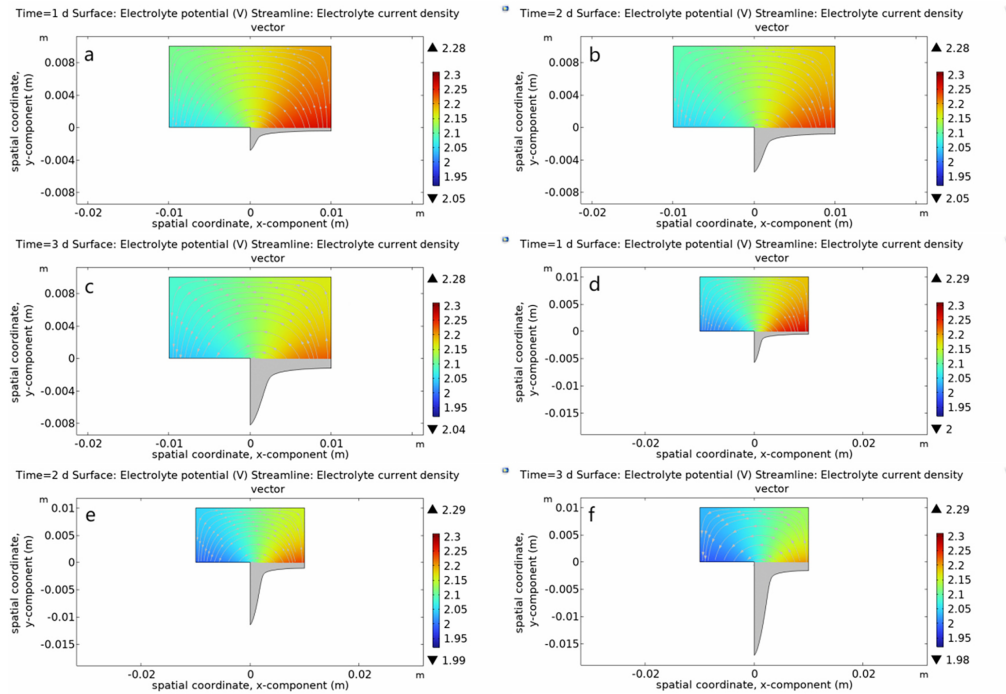


Figure 4: (Continued)

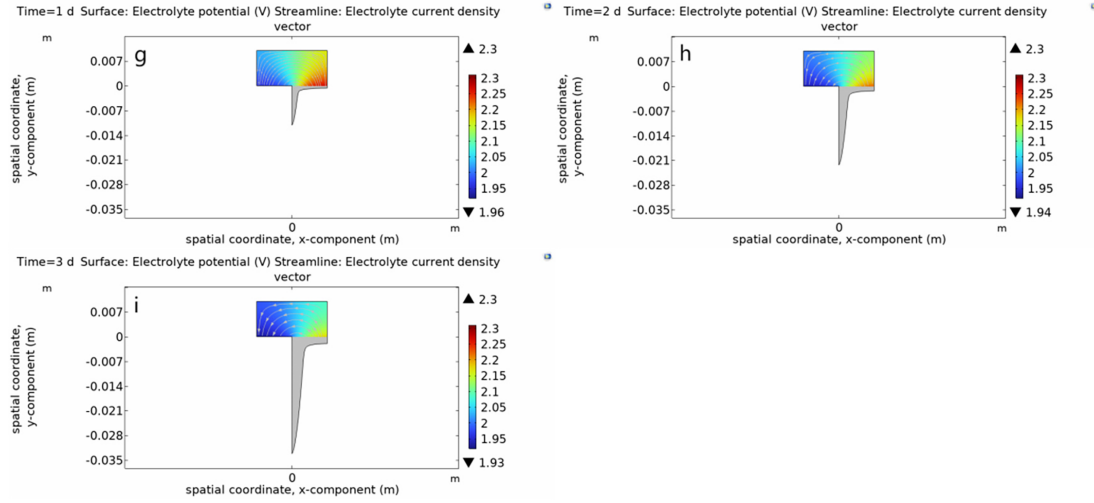


Figure 4: Electrolyte potential distribution at different temperatures and times. (a–c) 35°C, (d–f) 45°C, (g–i) 55°C. Time correspondence: 24 h (a,d,g), 48 h (b,e,h), 72 h (c,f,i). Porosity = 4%.

3.1.2 Effect of Temperature on Electrolyte Current Density

As shown in Fig. 5, at a temperature of 35°C, the current density amplitude is relatively low, with a maximum current density of approximately 250 A/m². The streamline distribution is relatively sparse, and the current density contours are gentle, indicating a relatively uniform current distribution. As time extends from 24 to 72 h, the maximum current density decreases slightly, from about 250 A/m² to about 248 A/m², with little change in the distribution pattern, suggesting that the system operates relatively stable under low temperature and low exchange current density conditions. At a temperature of 45°C, the current density amplitude increases significantly, with a maximum current density of approximately 532–549 A/m², about twice that at 35°C. The streamlines are noticeably concentrated toward the central region, and the curvature of the contours increases, indicating intensified non-uniformity in the current distribution. At a temperature of 55°C, the current density amplitude is the largest, with a maximum current density of approximately 1080–1100 A/m², about four times that at 35°C. The streamlines are highly concentrated in the central region, the contours are extremely curved, and the current distribution is highly non-uniform. From 24 to 72 h, the maximum current density remains essentially at 1100 A/m², but the area with high current density expands, indicating that current concentration intensifies over time. According to (Eq. (34)), it is found that as the temperature increases from 35°C to 55°C, the exchange current density increases exponentially, which is the fundamental reason for the significant increase in current density amplitude.

$$i_0 = i_0^{\text{ref}} \exp \left[-\frac{E_a}{R} \left(\frac{1}{T} - \frac{1}{T_{\text{ref}}} \right) \right] \quad (34)$$

i_0 —Exchange current density, i_0^{ref} —Exchange current density at reference temperature, E_a —Activation energy, T_{ref} —Temperature

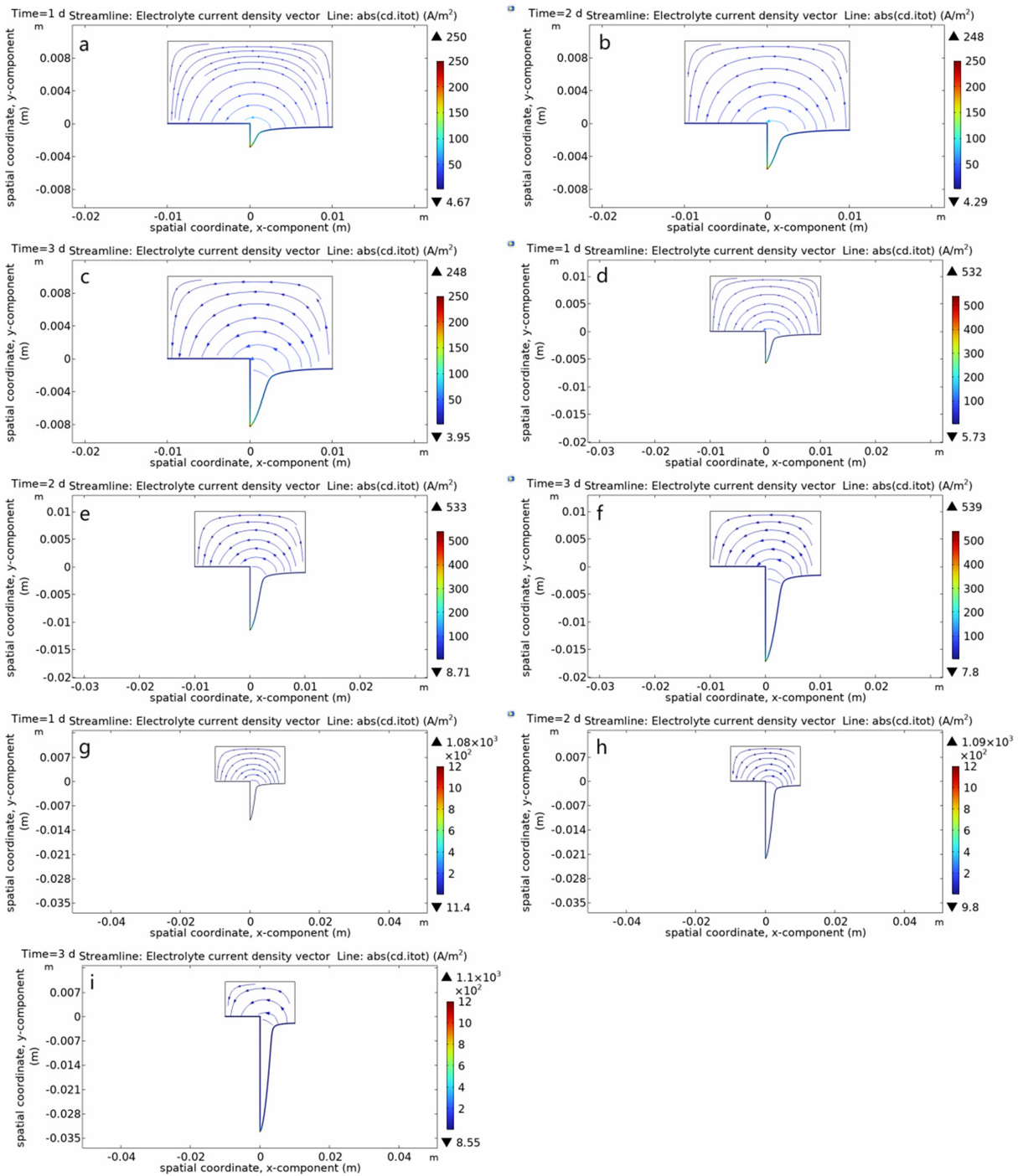


Figure 5: Electrolyte current density distribution at 35°C (a–c), 45°C (d–f), and 55°C (g–i). Time points: 24 h (a,d,g), 48 h (b,e,h), 72 h (c,f,i). Porosity = 4%.

3.1.3 Effects of Temperature on Electrode Thickness Variation and the Interface of Corrosion Products

The electrode thickness change refers to the maximum dissolution depth of the β -Li phase surface relative to its initial position, with a positive value indicating corrosion thinning. As shown in Figs. 6 and 7, under temperatures of 35°C, 45°C, and 55°C, the maximum changes in electrode thickness after 72 h

of operation are 8.2, 17.4, and 34.0 mm, respectively. Further analysis of the corrosion product thickness distribution reveals that under 35°C, 45°C, and 55°C, the maximum thickness of the corrosion product layer (defined as the farthest distance from the original β -Li surface to the 0.5 level set contour in Fig. 7) is approximately 1.8, 2.0, and 2.5 mm, respectively. The overall results indicate that as the temperature increases, the magnitude of electrode thickness increases significantly, and the non-uniformity of the distribution intensifies. For every 10°C increase in temperature, the maximum electrode thickness changes approximately doubles, exhibiting a clear temperature-accelerating effect. This trend is consistent with the exponential increase in exchange current density with temperature, aligning with the reaction kinetics described by the Arrhenius relationship. In summary, the increase in temperature significantly exacerbates the degree of localized corrosion on the electrode while also altering the distribution morphology and stability of the corrosion products. Under high-temperature conditions, the non-uniformity of current density distribution intensifies, and the differences in local reaction rates are amplified, causing the electrode thickness change to increase rapidly in current-concentrated regions. Meanwhile, the corrosion products show a tendency to migrate from the center toward the edges.

3.1.4 Effect of Temperature on Local Current Density

As shown in Fig. 8, when the temperature is 35°C, the maximum local current density at the anode is approximately 250 A/m², at 45°C, it is about 532 A/m², and at 55°C, it is about 1100 A/m². The current density at the cathode gradually decreases over time due to the continuous accumulation of corrosion products on the cathode surface. At 35°C, the current density near $x = 0.002$ m changes significantly over time, indicating that the amount of accumulated corrosion products at this location increases the most with time. After 24 h, the corrosion products are mainly concentrated in this region. As the temperature rises, this accumulation range gradually shifts toward the anode region. This trend is consistent with the thickness distribution of the corrosion product interface, indicating that the local current density has a direct impact on the formation and evolution of the corrosion product layer.

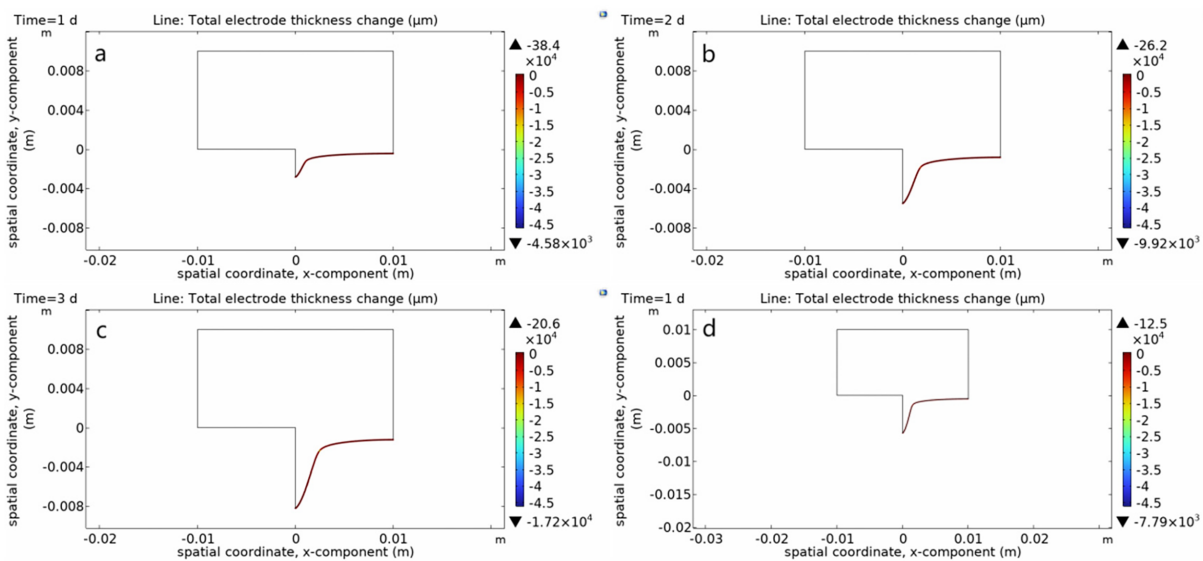


Figure 6: (Continued)

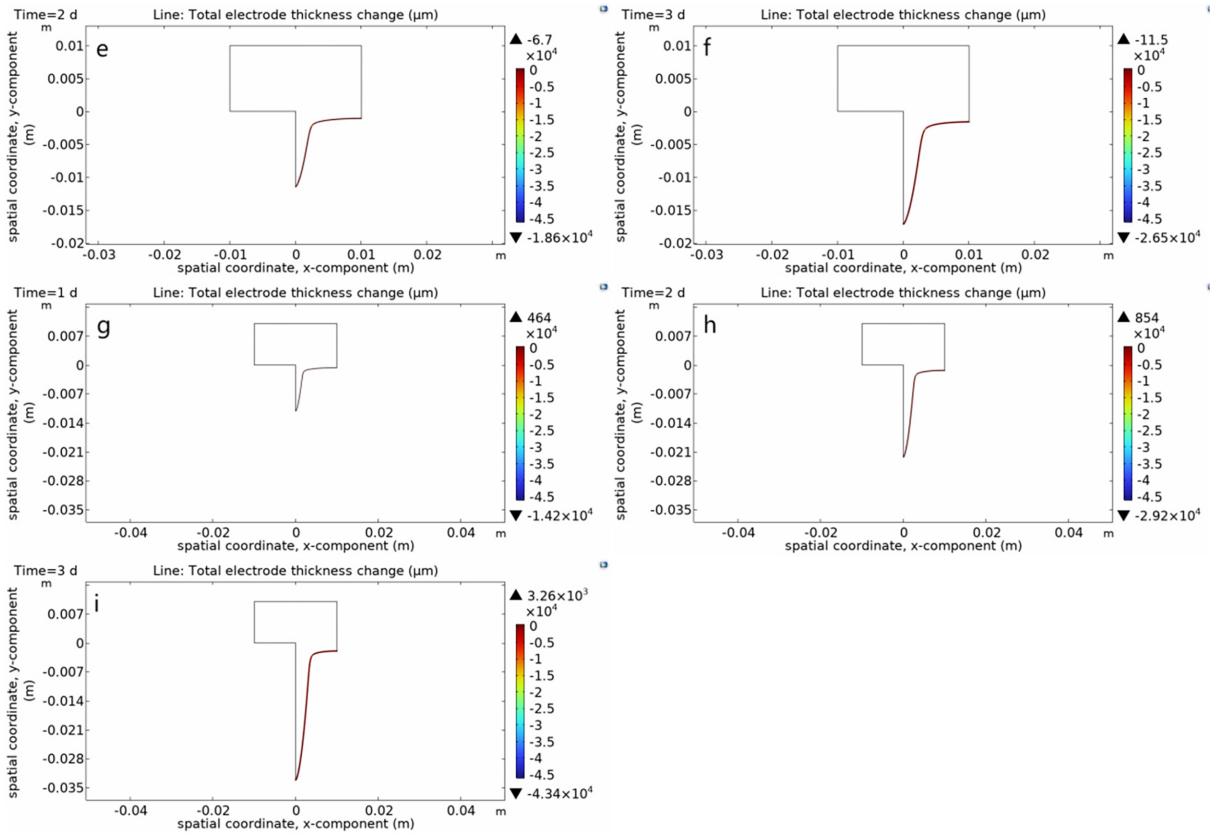


Figure 6: Electrode thickness change at different temperatures and times. (a–c) 35°C, (d–f) 45°C, (g–i) 55°C. Time correspondence: 24 h (a,d,g), 48 h (b,e,h) 72 h (c,f,i). Porosity = 4%.

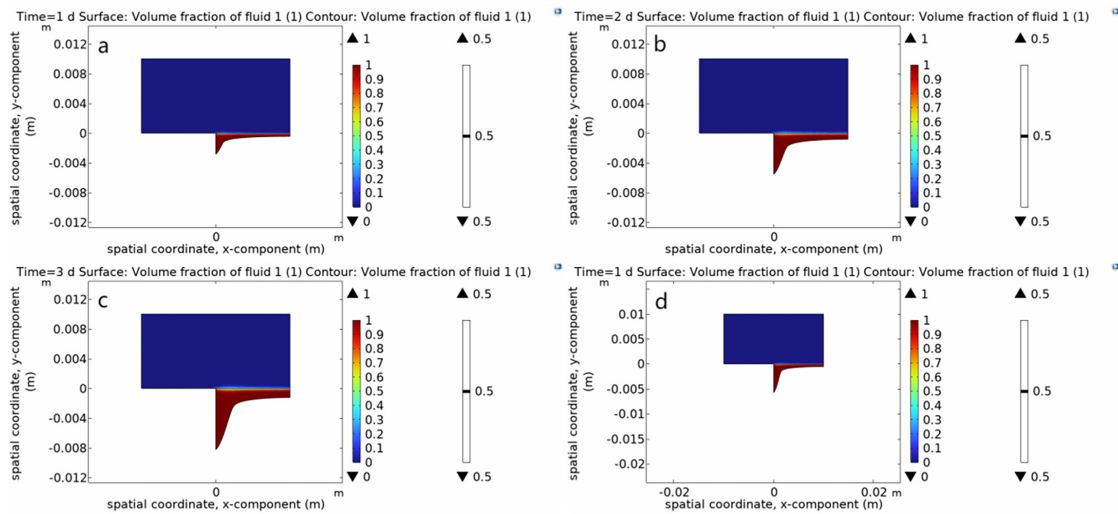


Figure 7: (Continued)

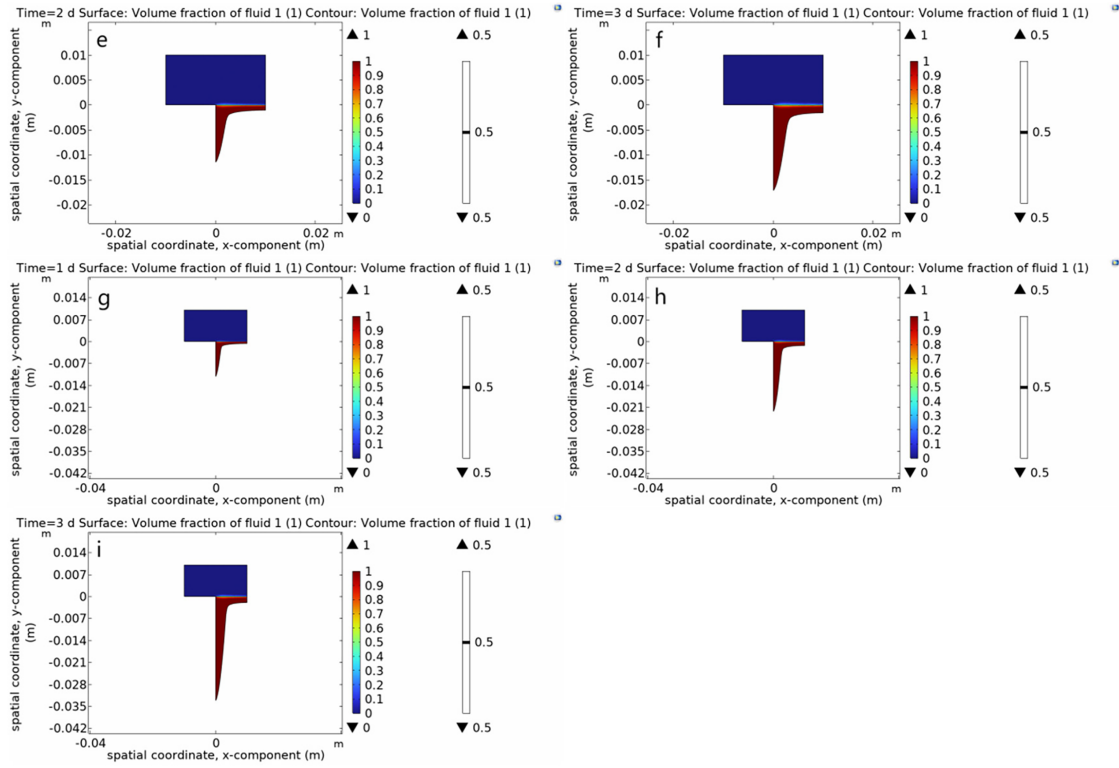


Figure 7: Corrosion product interface (volume fraction of fluid 1 = 0.5 contour) at different temperatures and times. (a–c) 35°C, (d–f) 45°C, (g–i) 55°C. Time correspondence: 24 h (a,d,g), 48 h (b,e,h), 72 h (c,f,i). Porosity = 4%.

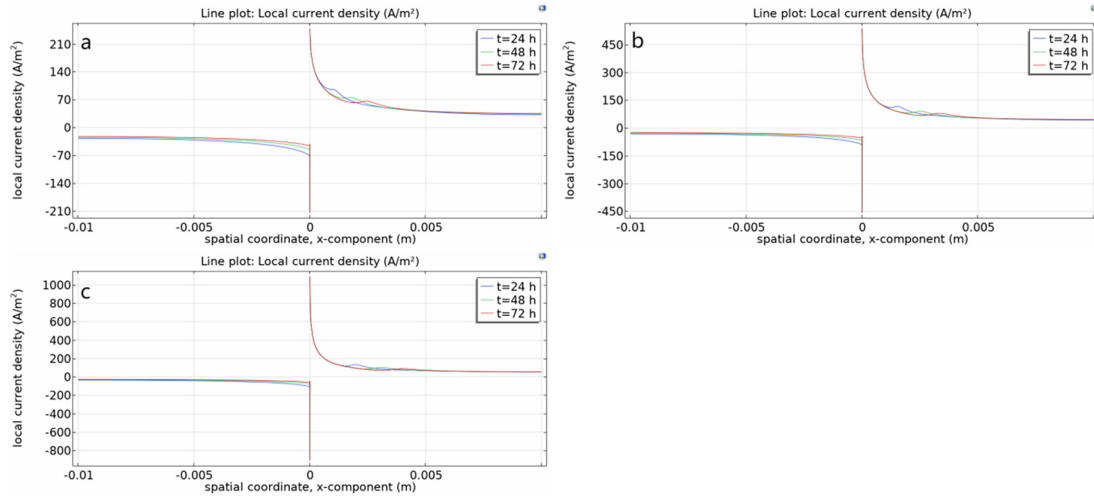


Figure 8: Local current density distribution along the electrode surface at different temperatures: (a) 35°C, (b) 45°C, (c) 55°C. Porosity = 4%.

3.2 Effect of Porosity on the Corrosion of β -Li Phase

3.2.1 Effect of Porosity on Electrolyte Potential

As shown in Fig. 9, when the porosity ε_p of the corrosion product layer (defined as the pore volume fraction, a lower value indicates a denser layer) is 3%, the electrolyte potential distribution ranges from

approximately 2.12 to 2.26 V, when the porosity is 3.5%, the distribution ranges from about 2.10 to 2.26 V. When the porosity is 4.5%, the distribution ranges from about 2.07 to 2.27 V. This is because an increase in the porosity of the corrosion product leads to more ion transport channels and a larger exposed area of the substrate, ultimately accelerating anodic dissolution and shifting the corrosion potential in the negative direction. At a corrosion product porosity of 4.5%, the minimum potential drops to about 2.07 V after 72 h, which is approximately 0.05 V lower compared with that at 3% porosity. This is attributed to the fact that under high porosity conditions, the corrosive medium is more easily penetrating the substrate and results in a continuous decrease in potential.

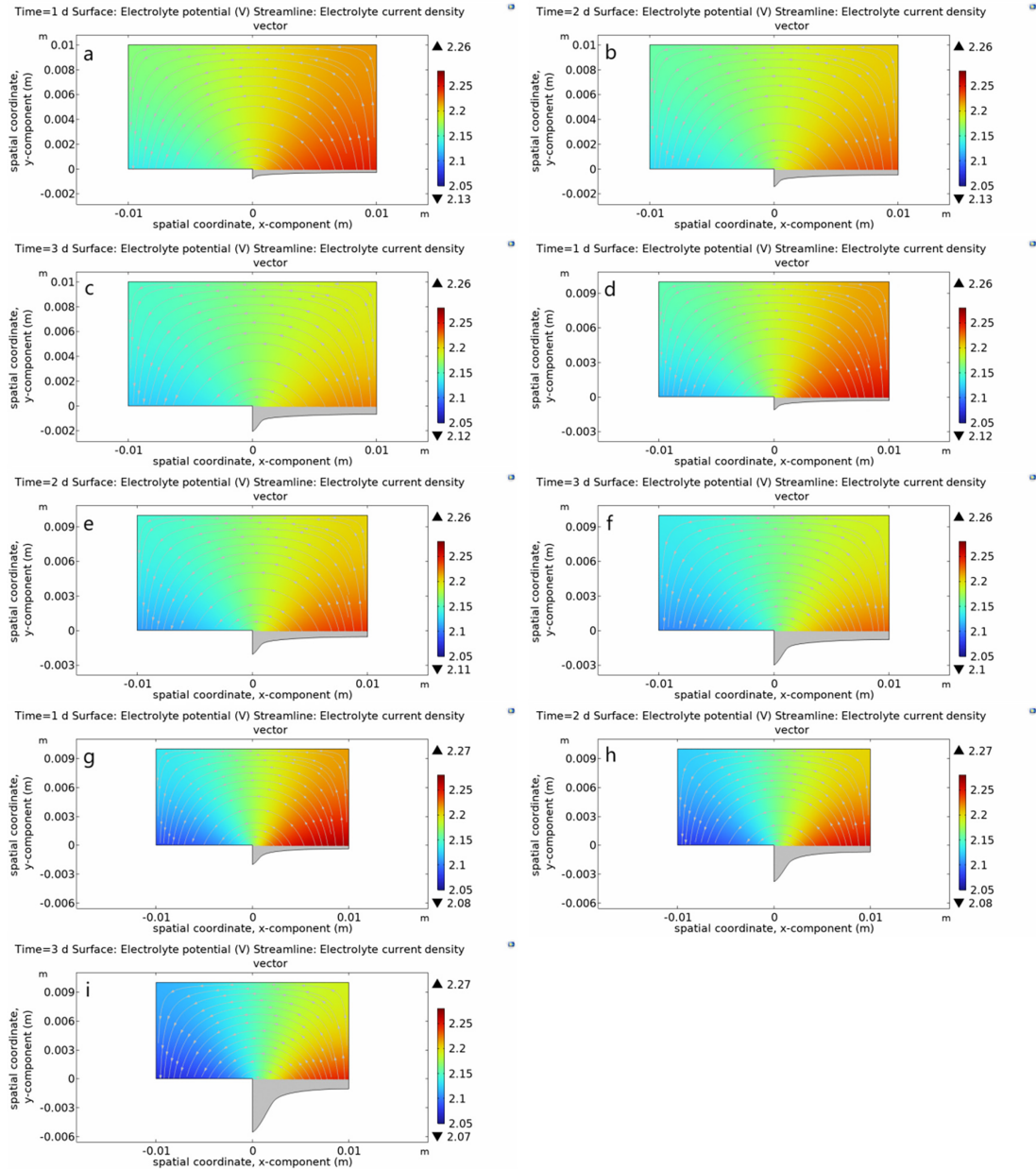


Figure 9: Electrolyte potential distribution for different corrosion product porosities. (a–c) porosity = 3%, (d–f) porosity = 3.5%, (g–i) porosity = 4.5%. Time correspondence: 24 h (a,d,g), 48 h (b,e,h), 72 h (c,f,i).

3.2.2 Effect of Porosity on Electrolyte Current Density

As shown in Fig. 10, when the porosity of the corrosion products is 3%, the maximum electrolyte current density is 60.2 A/m^2 , when the porosity is 3.5%, the maximum electrolyte current density is 88.6 A/m^2 , when the porosity is 4.5%, the maximum electrolyte current density is 162 A/m^2 . It is obvious that as the porosity increases, the current density increases significantly, indicating that higher porosity facilitates ion transport, resulting in a greater electrolyte current density and stronger electrochemical activity. With the extension of corrosion time, the current density of each group of samples shows a slight downward trend. When the porosity is 3%, the current density decreases from 60.2 to 57.6 A/m^2 , a decrease of approximately 4.3%, when the porosity is 3.5%, it decreases from 88.6 to 85.5 A/m^2 , a decrease of approximately 3.5%, when the porosity is 4.5%, it decreases from 162 to 159 A/m^2 , a decrease of approximately 1.9%. The magnitude of the decrease diminishes as the porosity increases, indicating that a high-porosity structure can effectively maintain unobstructed ion channels, leading to an increased corrosion rate.

3.2.3 Effects of Porosity on Electrode Thickness Variation and the Corrosion Product Interface

As shown in Figs. 11 and 12, under different corrosion product porosity conditions, the electrode thickness change exhibits significant differences. When the corrosion product porosity is 3%, the electrode thickness changes at 24, 48, and 72 h are 0.8, 1.0, and 0.5 mm, respectively. When the porosity is 3.5%, the corresponding thickness changes are 1.0, 1.1, and 0.9 mm, respectively, when the porosity is 4.5%, they are 2.0, 1.9, and 2.0 mm, respectively. It can be observed that as the porosity of the corrosion products increases, the corrosion products become looser, the resistance to ion transport decreases, and the current density increases correspondingly, leading to the increasing in electrode thickness change. Further observation of the corrosion product interface reveals that the thickness of corrosion products in the anodic region outside the corrosion pit also varies under different porosity conditions. At 72 h, for porosities of 3%, 3.5%, and 4.5%, the maximum corrosion product thickness is approximately 0.7, 0.8, and 1.0 mm, respectively. Under high porosity conditions, the current distribution becomes more uniform, and extremely localized high current density points no longer appear in the anodic region, avoiding excessive accumulation of corrosion products in local areas and resulting in a more uniform overall product layer. Thus, at low porosity (3%), the dense layer suppresses local current concentration and offers protection, at high porosity (4.5%), the looser structure accelerates corrosion but also avoids extreme localized attack due to more uniform current distribution [36].

3.2.4 Effect of Porosity on Local Current Density

As shown in Fig. 13, when the porosity of the corrosion product layer is 3%, the maximum local current density at the anode is maintained at approximately 60.2 A/m^2 . When the porosity increases to 3.5%, maximum local current density rises to $\sim 88.6 \text{ A/m}^2$. When the porosity further increases to 4.5%, the maximum local current density reaches about 162 A/m^2 . With increasing porosity, the corrosion product layer becomes looser and its structural compactness decreases, significantly deteriorate its physical barrier effect on ion transport, thereby leading to a continuous increase in local current density. Meanwhile, the current density at the cathode also shows a gradual decreasing trend over time. This is mainly due to the gradual deposition of corrosion products on the cathode surface, which covers active sites and increases ion diffusion resistance, resulting in a progressive decline in the reduction reaction rate [37]. Overall, low porosity favors the formation of a dense barrier layer that inhibits local corrosion, whereas high porosity accelerates anodic dissolution and reduces the long-term stability of the electrode.

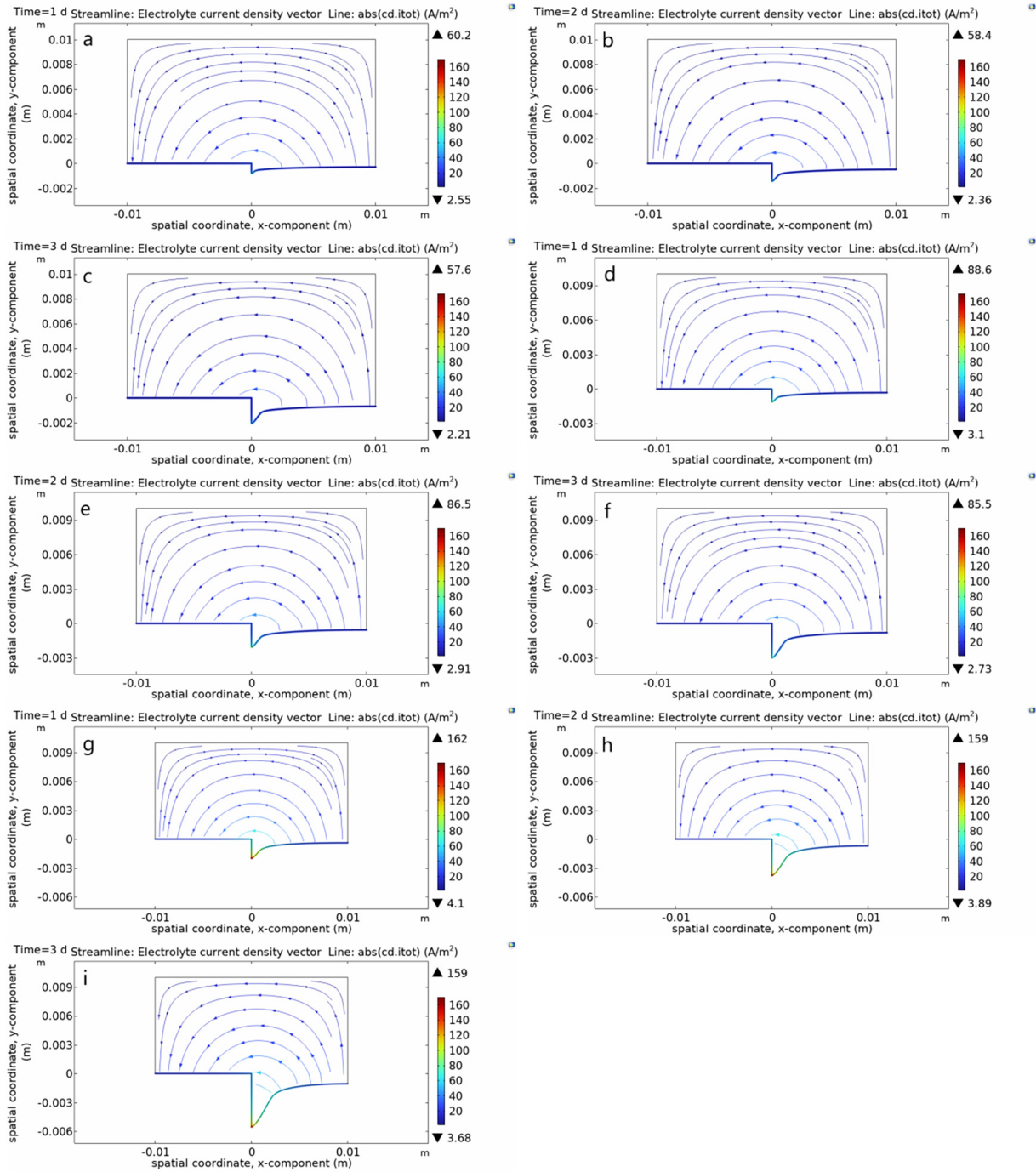


Figure 10: Electrolyte current density distribution for different porosities. (a–c) porosity = 3%, (d–f) porosity = 3.5%, (g–i) porosity = 4.5%. Time correspondence: 24 h (a,d,g), 48 h (b,e,h), 72 h (c,f,i).

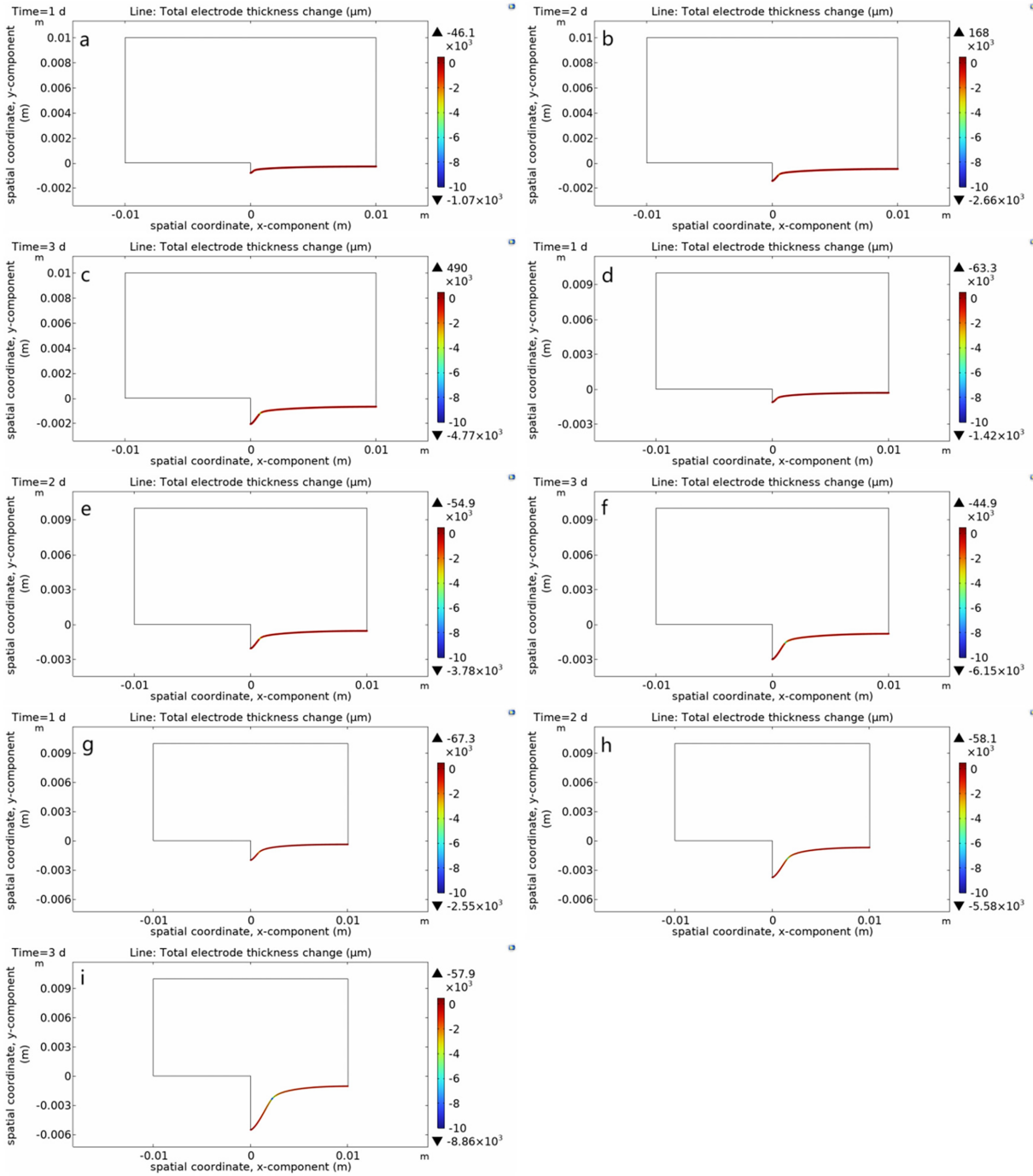


Figure 11: Electrode thickness change for different porosities. (a–c) porosity = 3%, (d–f) porosity = 3.5%, (g–i) porosity = 4.5%. Time correspondence: 24 h (a,d,g), 48 h (b,e,h), 72 h (c,f,i).

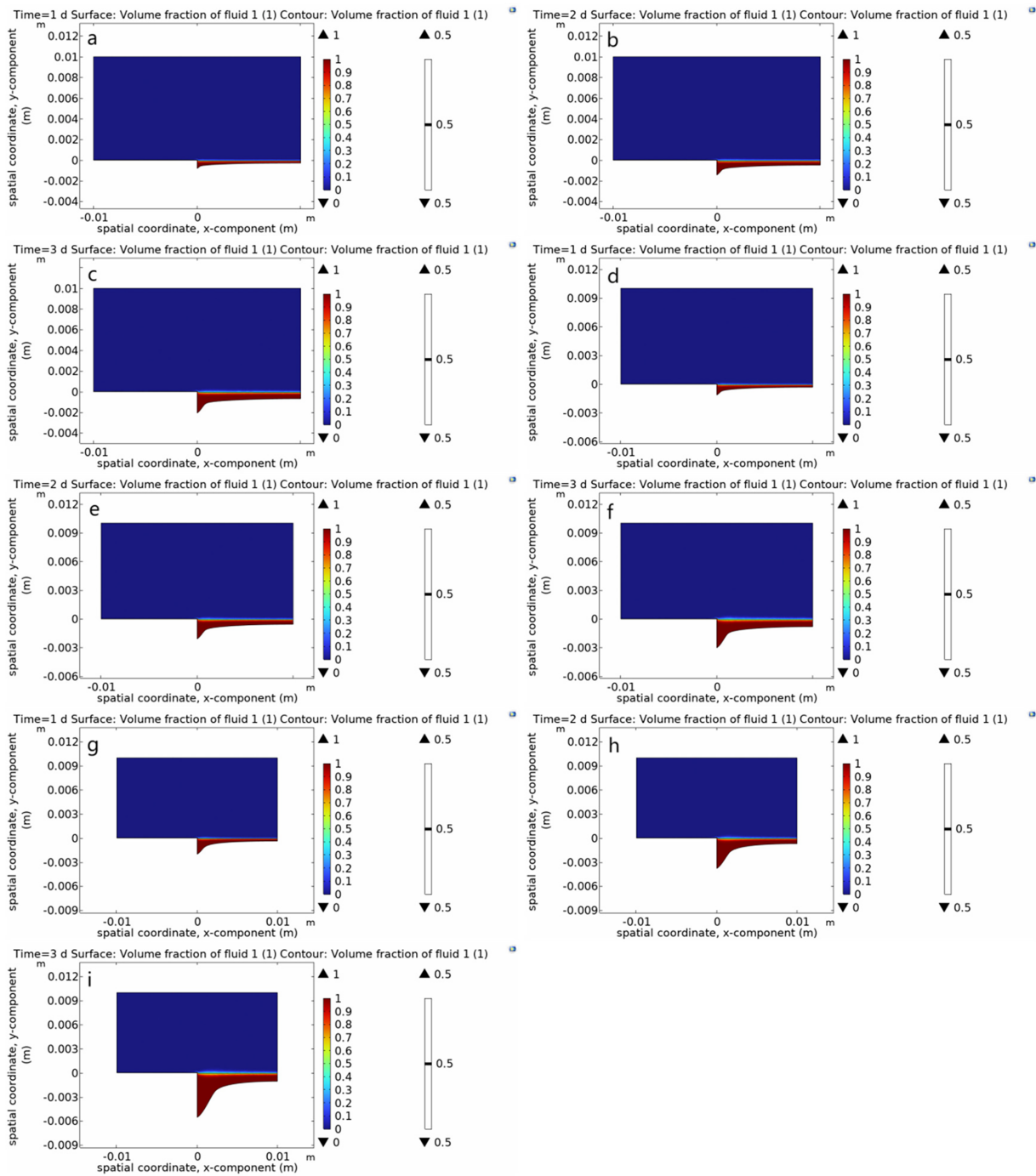


Figure 12: Corrosion product interface for different porosities. (a–c) porosity = 3%, (d–f) porosity = 3.5%, (g–i) porosity = 4.5%. Time correspondence: 24 h (a,d,g), 48 h (b,e,h), 72 h (c,f,i).

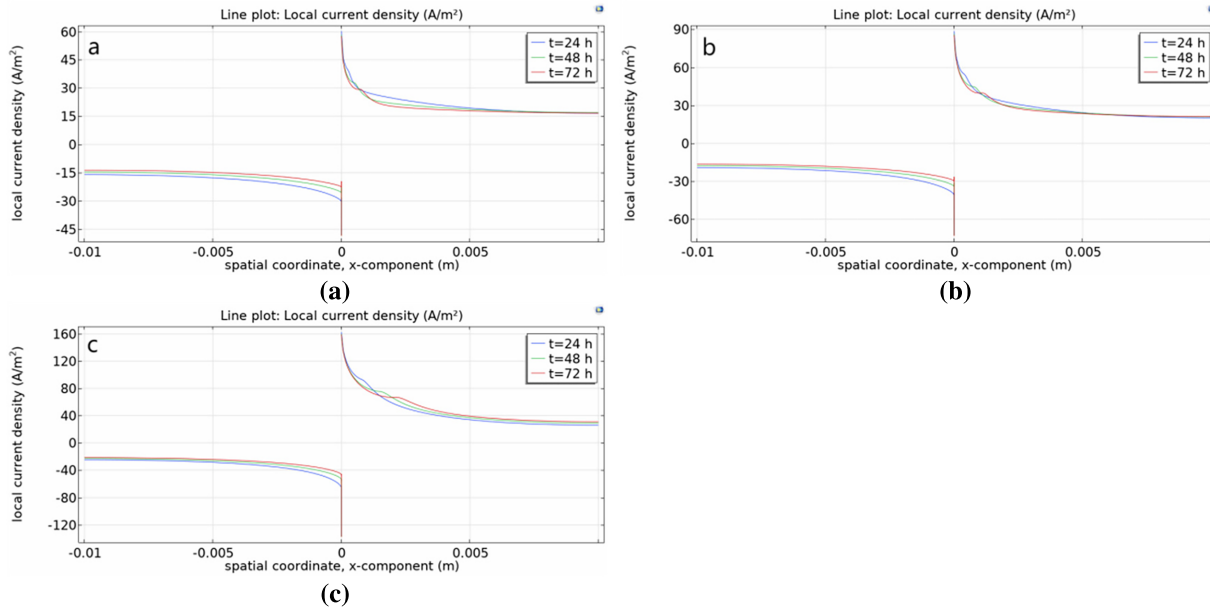


Figure 13: Local current density distribution for different porosities: (a) porosity = 3%, (b) porosity = 3.5%, (c) porosity = 4.5%.

4 Conclusion

In this work, the effects of temperature and corrosion product porosity on the micro-galvanic corrosion behavior between the α -Mg phase and the β -Li phase of the Mg-8Li alloy are systematically investigated using COMSOL Multiphysics numerical simulation software. The conclusions obtained are as follows:

- (1) As the temperature increases from 35°C to 55°C, the exchange current density increases exponentially. After 72 h of immersion at 35°C, 45°C, and 55°C, the maximum electrode thickness changes are 8.2, 17.4, and 34.0 mm, respectively, indicating that for every 10°C rise in temperature, the degree of localized corrosion approximately doubles. Under high-temperature conditions, the non-uniform distribution of current density intensifies, with significant accumulation of corrosion products in current-concentrated regions. The peak local current density increases from 250 A/m² (35°C) to 1100 A/m² (55°C), demonstrating that high temperature significantly accelerates the dissolution of the β -Li phase.
- (2) The porosity of the corrosion product exerts a dual regulatory effect on micro-galvanic corrosion behavior. When the porosity increases from 3% to 4.5%, the electrolyte current density rises from 60.2 to 162 A/m², and the electrode thickness change increases from 0.5 to 2.0 mm, indicating that high porosity significantly enhances ion transport capacity and accelerates anodic dissolution.
- (3) Temperature and corrosion product porosity can significantly alter the dissolution behavior of the β -Li phase and the morphology of the corrosion interface by regulating interfacial ion transport and current density distribution.

Therefore, in practical applications, the micro-galvanic corrosion of the β -Li phase can be mitigated by controlling the ambient temperature or by adjusting the compactness of the corrosion product layer (e.g., via surface modification techniques).

Acknowledgement: Not applicable.

Funding Statement: This research was funded by Postgraduate Research & Practice Innovation Program of Jiangsu Province, grant number SJCX23_2183.

Author Contributions: Yanxin Qiao and Chengtao Li, methodology, Haojie Zhu, software, Yuyang Zhang, validation, Haojie Zhu, Yuyang Zhang and Huiling Zhou, investigation, Haojie Zhu and Yuyang Zhang, data curation, Yuyang Zhang and Haibing Zhang, writing—original draft preparation, Yuyang Zhang and Yanxin Qiao, writing—review and editing, Yuyang Zhang, Yanxin Qiao and Haibing Zhang, supervision, Huiling Zhou. All authors reviewed and approved the final version of the manuscript.

Availability of Data and Materials: Data available on request from the authors.

Ethics Approval: Not applicable.

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

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