

Triangular Pitting of Mg Bioimplants

Mohammad Yaghoub Abdollahzadeh Jamalabadi

Department of Marine Engineering, Chabahar Maritime University, Chabahar, Iran.

***Corresponding Author:** Mohammad Yaghoub Abdollahzadeh Jamalabadi, Department of Marine Engineering, Chabahar Maritime University, Chabahar, Iran.

Received date: May 19, 2026; **Accepted date:** May 29, 2026; **Published date:** June 11, 2026

Citation: Abdollahzadeh Jamalabadi MY, (2026), Triangular Pitting of Mg Bioimplants, *J. Clinical Case Reports and Studies*, 7(6); DOI:10.31579/2690-8808/323

Copyright: ©, 2026, Mohammad Yaghoub Abdollahzadeh Jamalabadi. This is an open access article distributed under the Creative Commons Attribution License, which permits unrestricted use, distribution, and reproduction in any medium, provided the original work is properly cited.

Abstract

A phase-field model is developed to simulate the corrosion of magnesium (Mg) alloys in physiological environments. The framework couples Mg dissolution with the transport of Mg^{2+} ions in solution, naturally capturing the transition from activation-controlled to diffusion-controlled bio-corrosion. Beyond uniform attack, the model describes pitting corrosion with a triangular pit-front morphology — characteristic of textured Mg alloys in which preferential dissolution along crystallographic planes produces faceted, sharp-cornered pits — and accounts for the synergistic acceleration of corrosion under mechanical loading. The diffuse-interface formulation handles arbitrary two- and three-dimensional geometries without any special treatment of the moving corrosion front. Validation against dedicated in vitro experiments on WE43 Mg alloy wires demonstrates strong quantitative agreement. Case studies on Mg wires under tension and on bioabsorbable coronary stents reveal that triangular pits concentrate hydrostatic stress at their apex far more severely than circular pits of equal depth, thereby accelerating pit-to-crack transition and substantially reducing device service life. The proposed methodology provides a mechanistic tool for assessing the in vitro and in vivo performance of Mg-based biomedical devices and for optimising designs against realistic pit morphologies.

Key Words: diffuse interface; triangular pit morphology; stress-assisted corrosion; bio-absorbable; mg stent; pit-to-crack transition; mg biodegradation; phase-field

Nomenclature

Symbol Description

P Total potential energy

F Free energy functional

D Total dissipation rate

ϕ Phase-field variable (1 = solid Mg, 0 = liquid)

c^-_{Mg} Normalized Mg concentration

u Displacement vector

J Diffusional flux of Mg ions

μ_c, μ_ϕ Chemical potentials conjugate to concentration and phase-field

M, L Mobility coefficients (diffusion and interface kinetics)

$f^{\text{chem}}, f^{\text{grad}}, f^{\text{mech}}$ Chemical, gradient, and mechanical energy densities

κ Gradient energy coefficient

ω Double-well barrier height

Γ Interfacial energy

l Interface thickness

$h(\phi), g(\phi)$ Interpolation and double-well functions

A Curvature parameter of chemical free energy

C Elastic stiffness tensor

$\epsilon, \epsilon^e, \epsilon^p$ Total, elastic, and plastic strain tensors

σ Cauchy stress tensor

D_{cMg} Diffusion coefficient of Mg ions

L_0 Reference interfacial mobility (stress-free)

i, i_0 Anodic current density (with/without stress)

ϵ^p, ϵ_y Effective plastic strain, yield strain

σ_h Hydrostatic stress

V_m Molar volume

R Universal gas constant

T Absolute temperature

τ Ratio of diffusion time to reaction time

t^-, x^-, ∇^- Normalized time, space, and gradient

H_{gas} Hydrogen gas volume released

Δn_{Mg} Amount of dissolved Mg (moles)

L_0' Spatially dependent mobility for pitting

$\gamma(m,n), \phi(m,n)$ Random functions for pitting distribution

1. Introduction

Magnesium and its alloys have emerged as highly attractive candidates for temporary biomedical implants due to their unique combination of biocompatibility, biodegradability, and mechanical properties that closely rival those of cortical bone [1–6]. Temporary Mg implants are designed to

degrade progressively at a rate that matches tissue healing, thereby eliminating the need for a second surgical procedure to remove the hardware. Promising clinical outcomes have been reported across a range of applications, including orthopaedic fixation, cardiovascular stenting, and craniofacial reconstruction [4–6, 15]. Figure 1 presents the Schematic of the diffuse interface description and real pitting profile from top and depth view.

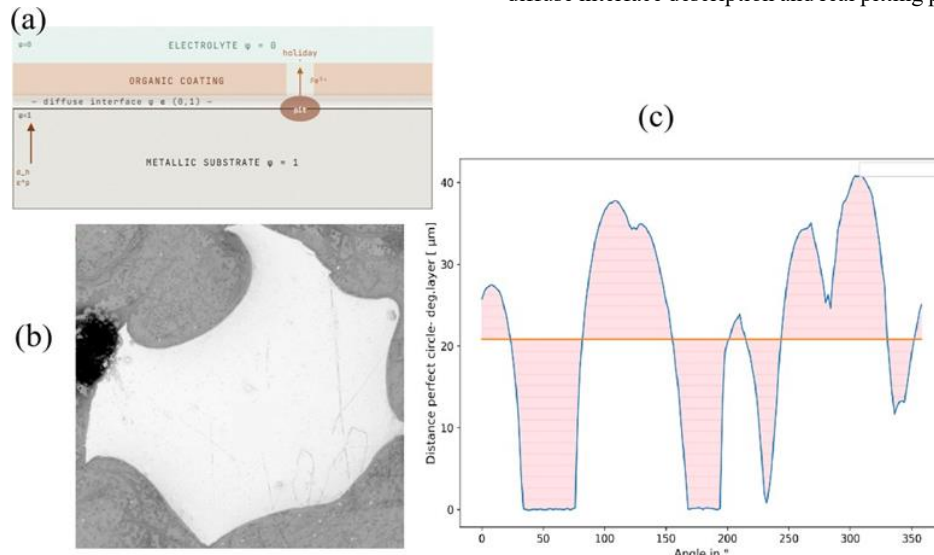


Figure 1: (a) Schematic of the diffuse interface description and real pitting profile from (b) top and (c) depth view. Data is from [61]

Despite these advantages, rapid and non-uniform degradation in chloride-rich body fluids remains the primary obstacle to broader clinical adoption [7–12]. Both in vivo and in vitro studies have consistently demonstrated that Mg alloys corrode in a highly localised fashion – a phenomenon known as pitting corrosion [8–10, 13, 14]. Importantly, pits in wrought Mg alloys with a pronounced basal crystallographic texture are frequently non-circular: preferential electrochemical dissolution along prismatic and pyramidal planes, which are inherently more reactive than basal planes, produces pits with triangular or otherwise faceted cross-sections [16–18]. These angular geometries introduce sharp corners that act as severe stress raisers and are known to accelerate the onset of stress-corrosion cracking (SCC) [19, 20, 23, 24].

The synergistic interaction between mechanical loading and corrosive attack is a further concern for load-bearing devices such as stents, wires, and bone screws [16–20]. Elastic strains lower the corrosion potential, increase anodic dissolution current densities, and promote localised damage [17, 18]. Plastic-strain concentrations at stent deployment sites create local hot-spots for pitting nucleation [16, 24]. Once a pit nucleates, the stress field at its tip can trigger a pit-to-crack transition, potentially leading to sudden, premature implant fracture [19, 20, 23]. Surface coatings and surface treatments have been explored to mitigate these effects, but they do not eliminate the fundamental mechano-electrochemical coupling [21, 22, 54].

Computational models have contributed significantly to the understanding of these coupled degradation processes [25–41]. Phenomenological continuum-damage approaches are computationally efficient but do not resolve the underlying physical mechanisms [25–27]. More advanced formulations incorporate Mg-ion diffusion and physicochemical interactions [28–30]. Physically based models that explicitly track the evolving corrosion

front – using arbitrary Lagrangian-Eulerian (ALE) methods, level-set schemes, or peridynamics – offer greater mechanistic insight but face difficulties in handling localised corrosion, topological changes, and mechano-electrochemical coupling simultaneously [31–41].

Phase-field methods have emerged as a natural framework to address these challenges [42–51]. By replacing the sharp solid-liquid interface with a

diffuse region described by a continuous auxiliary variable, the method implicitly tracks corrosion fronts of arbitrary complexity – including pit coalescence and branching – without the need for ad hoc criteria or remeshing [42, 43]. Phase-field models have been applied successfully to pitting and stress-corrosion cracking in metallic systems [44–46] and to galvanic damage in Mg alloys [47, 48, 51]. The mechano-electrochemical coupling can be incorporated via Gutman's theory [52], and the thermodynamic consistency is ensured by standard formulations [53, 55–57].

A critical limitation of existing phase-field corrosion models is that pitting is invariably assumed to produce circular or hemispherical pit geometries [43–46, 49, 50]. This idealisation systematically underestimates local stress concentrations in textured Mg alloys, where triangular pits are routinely observed [58–60]. The present work addresses this gap by developing the first phase-field model that explicitly incorporates triangular pit-front morphology into coupled simulations of Mg dissolution, ionic transport, and mechanical deformation. The model is validated against dedicated in vitro experiments on WE43 Mg alloy wires [58–61] and applied to two engineering case studies: stress-assisted corrosion of Mg wires and degradation of a bioabsorbable coronary stent.

2. The Phase-Field Model for Triangular Pitting Corrosion

2.1 Degradation Mechanisms

Magnesium dissolves in aqueous physiological environments through a well-established electrochemical sequence. The anodic reaction releases Mg^{2+} ions into solution; cathodic water reduction generates hydrogen gas; and the subsequent precipitation reaction deposits a layer of magnesium hydroxide $[Mg(OH)_2]$ on the alloy surface, which acts as a temporary passive film. Chloride ions (Cl^-) present in body fluids react with this film, converting it to the highly soluble $MgCl_2$ and thereby locally undermining passivation – the primary trigger for pitting. The sequence of anodic, cathodic, precipitation, and layer dissolution reactions governing Mg biodegradation is summarized in Table 1.

Formula
$\text{Mg}_{(s)} \rightarrow \text{Mg}_{(aq)}^{2+} + 2e^-$ (anodic)
$2\text{H}_2\text{O} + 2e^- \rightarrow \text{H}_2 \uparrow + 2\text{OH}^-$ (cathodic)
$\text{Mg}^{2+} + 2\text{OH}^- \rightarrow \text{Mg}(\text{OH})_2 \downarrow$ (product formation)
$\text{Mg}(\text{OH})_2 + 2\text{Cl}^- \rightarrow \text{MgCl}_2 + 2\text{OH}^-$ (layer dissolution)

Table 1: Magnesium dissolution and layer dissolution in physiological environment.

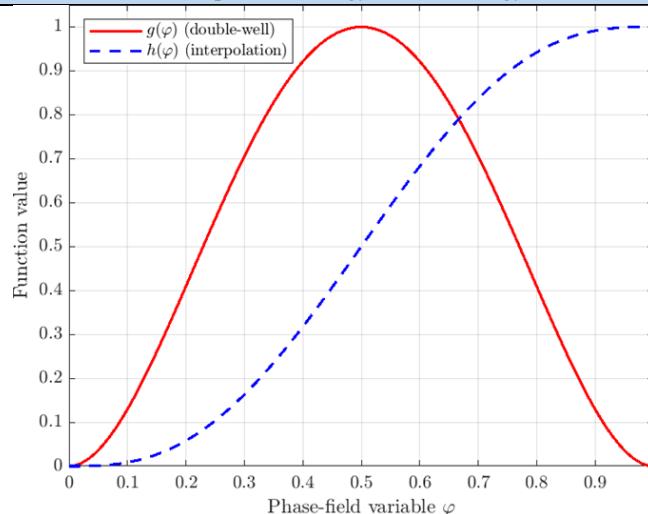
In wrought Mg alloys with strong basal texture, the dissolution rate is crystallographically anisotropic. Basal planes {0001} exhibit the lowest surface energy and corrode relatively slowly, whereas prismatic {10–10} and pyramidal {10–11} planes dissolve more rapidly. This anisotropy causes pits to develop faceted, predominantly triangular cross-sections. In the present model, the pit front is accordingly initialised and allowed to evolve as an equilateral triangle with an apex full-angle of approximately 60°, consistent with experimental observations on WE43 wires.

The effect of mechanical stresses on anodic dissolution kinetics is captured through Gutman's mechano-electrochemical theory, whereby both hydrostatic stress and accumulated plastic strain amplify the local dissolution rate. Following passive-film rupture and pit nucleation, the stress concentration at the sharp apex of a triangular pit is substantially greater than at the rounded tip of a circular pit of the same nominal depth, making the pit-to-crack transition more rapid and the subsequent crack propagation less controllable.

2.2 Thermodynamic Free Energy Functional

The decomposition of the total potential energy and free energy densities into chemical, gradient, and mechanical contributions is presented in Table 2. The phase-field variable ϕ equals 1 in the intact Mg alloy and 0 in the physiological fluid, transitioning smoothly through a diffuse interface of thickness ℓ . The normalised Mg^{2+} concentration is defined as $\bar{c}_{\text{Mg}} = c_{\text{Mg}} / c^s_{\text{Mg}}$, and $\mathbf{u}$ denotes the displacement vector. The chemical free energy density follows a standard double-well formulation with interpolation functions ensuring thermodynamic consistency. The gradient energy density is $f_{\text{grad}} = \frac{1}{2}\kappa|\nabla\phi|^2$, with parameters linked to the interfacial energy Γ and interface thickness ℓ . The mechanical energy density follows an elastoplastic von Mises formulation, confined to the solid phase through multiplication by the interpolation function $h(\phi)$. The auxiliary functions used in the phase field formulation, including the double well barrier and the interpolation function, are plotted in Figure 2. Figure 2 plots the double-well potential $g(\phi) = 16\phi^2(1-\phi)^2$ (red solid) and the smooth interpolant $h(\phi) = \phi^3(6\phi^2 - 15\phi + 10)$ (blue dashed). These functions ensure a diffuse interface between the metal and electrolyte phases.

Formula
$\mathcal{P} = \int_{\Omega} [f^{\text{chem}} + f^{\text{grad}} + f^{\text{mech}}] d\Omega - \int_{\partial\Omega} \mathbf{t} \cdot \mathbf{u} d\partial\Omega$
$\mathcal{F} = \int_{\Omega} [f^{\text{chem}} + f^{\text{grad}} + f^{\text{mech}}] d\Omega$
$f^{\text{chem}}(\bar{c}_{\text{Mg}}, \phi) = (1-h)f_l^{\text{chem}} + hf_s^{\text{chem}} + \omega g(\phi)$
$g(\phi) = 16\phi^2(1-\phi)^2, \quad h(\phi) = \phi^3(6\phi^2 - 15\phi + 10)$
$f_l^{\text{chem}} = \frac{1}{2}A(\bar{c}_{\text{Mg}}^l - \bar{c}_{\text{Mg}}^{l,eq})^2, \quad f_s^{\text{chem}} = \frac{1}{2}A(\bar{c}_{\text{Mg}}^s - \bar{c}_{\text{Mg}}^{s,eq})^2$
$f^{\text{grad}}(\nabla\phi) = \frac{1}{2}\kappa \nabla\phi ^2$
$f^{\text{mech}}(\nabla\mathbf{u}, \phi) = h(\phi)(f_e^{\text{mech}} + f_p^{\text{mech}})$
$f_e^{\text{mech}} = \frac{1}{2}\boldsymbol{\epsilon}^e : \mathbf{C} : \boldsymbol{\epsilon}^e, \quad \boldsymbol{\epsilon}^e = \boldsymbol{\epsilon} - \boldsymbol{\epsilon}^p$
$f_p^{\text{mech}} = \int_0^t \boldsymbol{\sigma}_0 : \dot{\boldsymbol{\epsilon}}^p dt$

Table 2: Total potential energy and free energy densities.**Figure 2:** Phase-field auxiliary functions.

2.3 Triangular Pit Initialisation

The mathematical expression for the spatially dependent mobility function that enforces three-fold symmetry for triangular pitting is given in Table 3. The initial pit geometry is prescribed as a triangular void at the alloy surface. In the two-dimensional cross-sectional domain, the equilateral

triangle is oriented with one vertex pointing inward (toward the bulk) to maximise stress concentration under tensile loading, faithfully reproducing the geometry observed experimentally. The phase-field is set to $\phi = 0$ inside the triangle and $\phi = 1$ outside, transitioning smoothly across the diffuse interface of thickness ℓ . The triangle's side length is chosen to match the experimentally measured initial pit size.

Formula
$H_{\text{gas}} = \frac{\Delta n_{\text{Mg}} RT}{PA}$
$\Delta n_{\text{Mg}} = \int_{\Omega(t)} c_{\text{Mg}} d\Omega - \int_{\Omega(0)} c_{\text{Mg}} d\Omega$
$L_0' = L_0 \left[a_0 \sum_{m=-N}^N \sum_{n=-N}^N \frac{\gamma(m, n)}{(m^2 + n^2)^{\beta/2}} \cos(2\pi(m\tilde{x} + n\tilde{y}) + \phi(m, n)) + a_1 \right]$

Table 3: Hydrogen release, mass loss, and stochastic pitting mobility.

To sustain a triangular growth morphology as the pit evolves, the interfacial mobility L is made spatially dependent through a trigonometric pitting function $f(\tilde{x}, \alpha)$ that enforces three-fold crystallographic symmetry. This approach captures the anisotropic dissolution kinetics responsible for triangular pit shapes in textured alloys without requiring an explicit crystallographic orientation model.

2.4 Mechano-Electrochemical Coupling

The amplification of anodic dissolution by mechanical fields is incorporated through Gutman's theory. The phase-field parameters, including the interfacial mobility and the mechano-electrochemical coupling factor linking

stress and dissolution rate, are detailed in Table 4. The ratio of the anodic current density in the presence of stress and plastic strain to the stress-free value is in Table 4 where ϵ_p is the effective plastic strain, ϵ_y the initial yield strain, σ_h the hydrostatic stress, V_m the molar volume of Mg, R the universal gas constant, and T the absolute temperature. The interfacial mobility coefficient is modified analogously, with L_0 denoting the reference mobility calibrated from stress-free in vitro corrosion experiments. The mechano-electrochemical contribution is particularly pronounced at the apex of triangular pits, where the local radius of curvature is an order of magnitude smaller than that of a circular pit with the same depth, yielding significantly higher hydrostatic stress and consequently a more aggressive local dissolution rate.

Formula
$\omega = \frac{3\Gamma}{4\ell}, \quad \kappa = \frac{3}{2}\Gamma\ell$
$\mu_\phi = \frac{\partial f^{\text{chem}}}{\partial \phi} - \kappa \nabla^2 \phi, \quad \mu_c = \frac{\partial f^{\text{chem}}}{\partial \bar{c}_{\text{Mg}}}$
$\mathbf{J} = -M \nabla \mu_c, \quad \dot{\phi} = -L \mu_\phi$
$M = \frac{D_{\text{cMg}}}{\partial^2 f^{\text{chem}} / \partial \bar{c}_{\text{Mg}}^2}$
$\frac{L}{L_0} = \left(\frac{\epsilon_p}{\epsilon_y} + 1 \right) \exp \left(\frac{\sigma_h V_m}{RT} \right)$
$D_{\text{cMg}} = D_{\text{cMg}}^e h(\phi) + (1 - h(\phi)) D_{\text{cMg}}^l$

Table 4: Phase-field parameters, mobility, and mechano-electrochemical coupling.

2.5 Governing Equations

The governing equations for the phase-field variable, the normalized Mg^{2+} concentration, and the displacement field are derived from the principle of virtual power. The coupled evolution equations for the phase-field variable,

Mg^{2+} concentration, and mechanical equilibrium are summarized in Table 5 and Table 6. These comprise: (i) the Allen–Cahn equation governing the non-conserved phase-field; (ii) a modified diffusion equation for Mg ions in which the effective diffusion coefficient interpolates between solid and liquid values; and (iii) a quasi-static mechanical equilibrium equation.

Formula
$\frac{\partial \phi}{\partial t} = -L \left(\frac{\partial f^{\text{chem}}}{\partial \phi} - \kappa \nabla^2 \phi \right)$
$\frac{\partial \bar{c}_{\text{Mg}}}{\partial t} = -\nabla \cdot \mathbf{J}$
$\mathbf{J} = -D_{\text{cMg}} \nabla \bar{c}_{\text{Mg}} - D_{\text{cMg}} h'(\phi) (\bar{c}_{\text{Mg}}^{l,eq} - \bar{c}_{\text{Mg}}^{s,eq}) \nabla \phi$
$\nabla \cdot \boldsymbol{\sigma} = \mathbf{0}$

Table 5: Evolution equations for ϕ , $\bar{c}_{\text{Mg}}$, and mechanical equilibrium.

Formula
$\bar{t} = \frac{t D_{\text{cMg}}^l}{\ell^2}, \quad \bar{\mathbf{x}} = \frac{\mathbf{x}}{\ell}, \quad \bar{\nabla} = \ell \nabla$

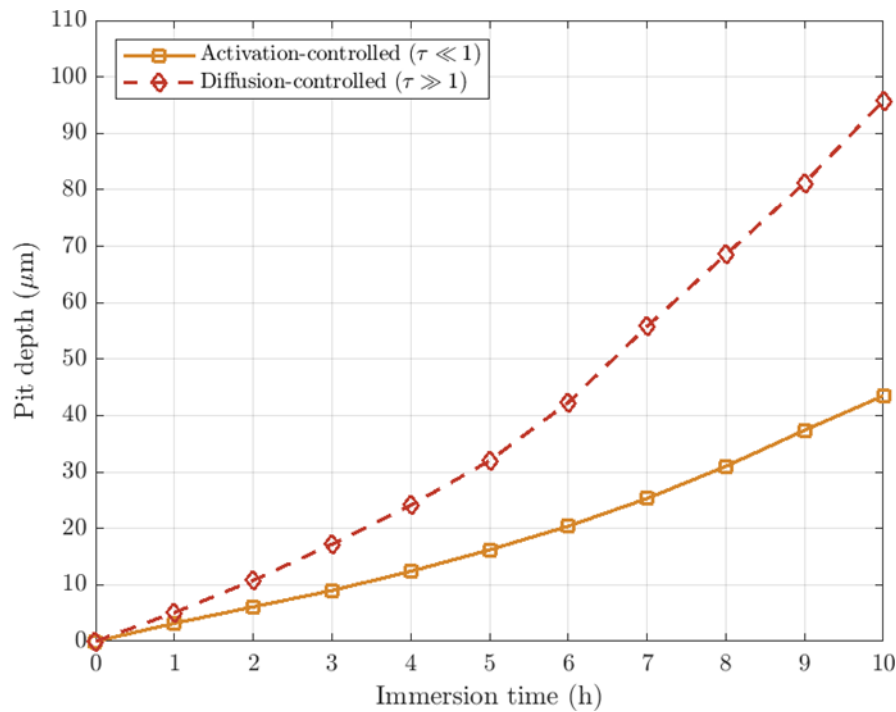
Formula
$\bar{c}_{Mg} = c_{Mg}/c_{Mg}^s, \quad \bar{D}_{cMg} = D_{cMg}/D_{cMg}^l, \quad \bar{\sigma} = \sigma/\omega, \quad \bar{\kappa} = \kappa/(\omega\ell^2)$
$\frac{\partial \phi}{\partial \bar{t}} = -\tau \left(\frac{\partial \bar{f}^{\text{chem}}}{\partial \phi} - \bar{\kappa} \nabla^2 \phi \right)$
$\frac{\partial \bar{c}_{Mg}}{\partial \bar{t}} = \nabla \cdot [\bar{D}_{cMg} \nabla \bar{c}_{Mg} + \bar{D}_{cMg} h'(\phi)(\bar{c}_{Mg}^{l,eq} - \bar{c}_{Mg}^{s,eq}) \nabla \phi]$
$\nabla \cdot \bar{\sigma} = 0$
$t_d = \frac{\ell^2}{D_{cMg}^l}, \quad t_\phi = \frac{1}{L_0}, \quad \tau = \frac{t_d}{t_\phi} = \frac{L}{D_{cMg}^l} \ell^2 \omega$

Table 6: Normalized variables and governing equations.

The equations are normalized using the interface thickness as the characteristic length, the liquid-phase Mg diffusivity as the reference diffusivity, and the barrier height as the energy scale. The dimensionless time ratio $\tau = L D_{cMg}^l / (\ell^2 \omega)$ determines the rate-limiting regime: diffusion-controlled for $\tau \gg 1$ and activation-controlled for $\tau \ll 1$. Triangular pitting is introduced through a spatially dependent mobility parameter that assembles spatial frequency components with prescribed amplitude distributions to enforce three-fold symmetry. The boundary conditions applied to the phase-

field, concentration field, and mechanical deformation for all simulations are specified in Table 7. Figure 3 illustrates the transition from activation controlled to diffusion controlled pit growth as the ratio of diffusion time to reaction time varies. Figure 3 illustrates the transition between kinetic regimes. Activation-controlled growth ($\tau \ll 1$) yields a concave shape, while diffusion-controlled growth ($\tau \gg 1$) leads to a more linear early evolution. The dimensionless parameter τ distinguishes the two extremes.

Formula
$\kappa \mathbf{n} \cdot \nabla \phi = 0, \quad \mathbf{n} \cdot \mathbf{J} = 0 \quad \text{on } \partial \Omega$
$\mathbf{t} = \mathbf{n} \cdot \boldsymbol{\sigma} = \mathbf{t}^0 \quad \text{on } \partial \Omega_t, \quad \mathbf{u} = \mathbf{u}^0 \quad \text{on } \partial \Omega_u$

Table 7: Boundary conditions for phase-field, concentration, and mechanics.**Figure 3:** Activation vs. diffusion controlled pit growth.

3. Experiments and Model Validation

3.1 Materials and Experimental Methods

An in vitro degradation study was conducted on WE43Mg alloy wires of 0.3 mm diameter, manufactured by cold drawing at Meotec GmbH (Aachen, Germany). The nominal composition was 1.4–4.2 wt. % Y, 2.5–3.5 wt. % Nd, <1 wt. % combined (Al, Fe, Cu, Ni, Mn, Zn, Zr), and balance Mg. Post-drawing annealing at 450 °C for 5 s reduced the dislocation density and improved ductility.

Corrosion tests were performed on 120 mm wire segments immersed in corrected Simulated Body Fluid (c-SBF, pH 7.5, 37 °C) at a fluid-to-surface-area ratio exceeding 0.5 mL/mm² in accordance with ASTM G31-72. The degradation rate was assessed by measuring evolved hydrogen gas using a eudiometer-equipped glass burette. Twelve samples were used in total; four were removed after 24 h for cross-sectional characterisation. Twenty random cross-sections per sample were mounted in epoxy resin, polished, and imaged by optical microscopy. Images were analysed using the PitScan framework to extract three quantitative pitting metrics: uniform corrosion radius, average pit depth, and maximum pit depth.

3.2 Model Validation: Uniform Corrosion

Phase-field simulations of uniform corrosion were performed on an axisymmetric domain with no-flux boundary conditions imposed at all outer edges. The reference interfacial mobility L_0 was calibrated by fitting the predicted hydrogen evolution curve to experimental measurements, yielding $L_0 = 2.3 \times 10^{-10} \text{ m}^3/(\text{J}\cdot\text{s})$, corresponding to a dimensionless ratio $\tau = 3.45 \times 10^{-6}$ — firmly in the activation-controlled regime. Phase-field predictions reproduced the characteristic shape of the hydrogen evolution curve: an initially rapid linear phase up to 24 h followed by a plateau at 120 h, attributed primarily to the progressive reduction in exposed surface area as the circular wire cross-section diminishes. The material properties and phase-field parameters used in the simulations, such as ion diffusivities and interfacial energy, are listed in Table 8. The measured and predicted hydrogen gas evolution as a function of immersion time are compared in Figure 4. Plot superimposes the cumulative H_2 volume (left axis, blue solid) and the instantaneous emission rate (right axis, orange dashed). The rate

peaks around 48 h and then decays, suggesting a transition from surface-controlled evolution to transport-limited diffusion through the corrosion product layer. As well figure 5 shows the growth of the uniform corrosion blister radius over time, extracted from the phase field simulations.

Figure 5 presents the undercutting/blister radius as a function of immersion time. Phase-field mean values and ensemble scatter are compared with experimental means $\pm$ SD. The model matches the experiment closely, with the radius increasing from $\sim 0.8 \text{ mm}$ at 24 h to $\sim 1.65 \text{ mm}$ at 240 h. The cumulative hydrogen volume and the corresponding evolution rate are presented in Figure 6 which is compatible with figure 4. The relative error of the phase-field model for the blister radius at different immersion times is reported in Figure 7. Figure 7 evaluates the relative error for blister radius. The error remains below 7% at all measured times (24, 72, 120, 240 h), with the highest error (6.5%) at 24 h and the lowest (4%) at 240 h. This excellent agreement confirms the model's predictive power.

Parameter	Symbol	Value	Unit
Mg-ion diffusivity (liquid phase)	D^l	10^{-10}	m^2/s
Mg-ion diffusivity (solid phase)	D^s	10^{-13}	m^2/s
Equilibrium concentration — liquid	c^{lxhls}	0.57	mol/L
Equilibrium concentration — solid	c^{shls}	71.44	mol/L
Molar volume of Mg	V_m	13.998	cm^3/mol
Interfacial energy	Γ	0.5	J/m^2
Interface thickness	ℓ	4	μm
Absolute temperature	T	310.15	K
Reference interfacial mobility	L_0	2.3×10^{-10}	$\text{m}^3/(\text{J}\cdot\text{s})$

Table 8: Material and phase-field parameters used in all simulations.

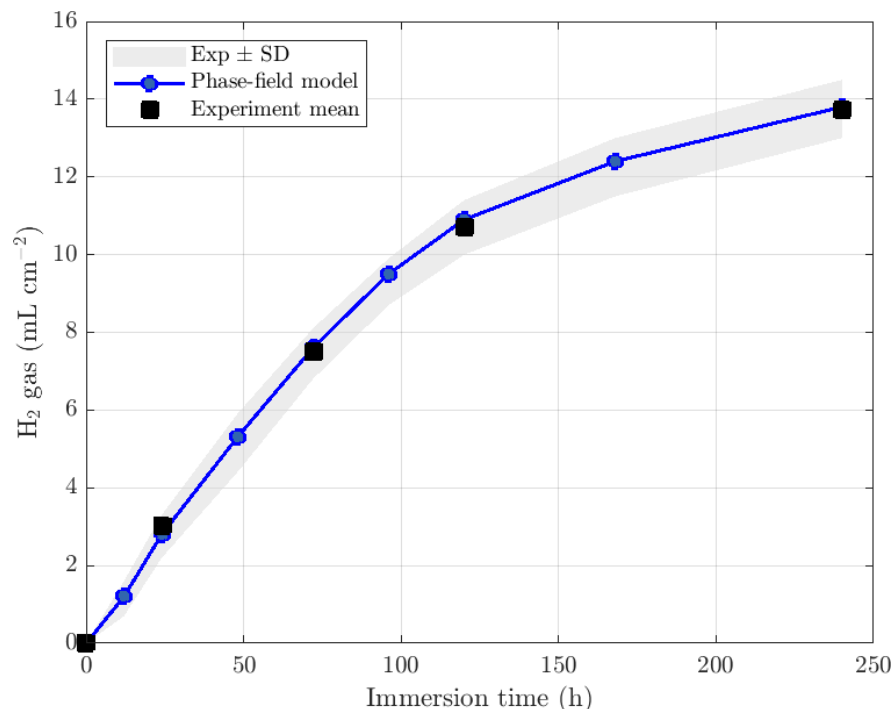


Figure 4: Hydrogen gas evolution vs. immersion time.

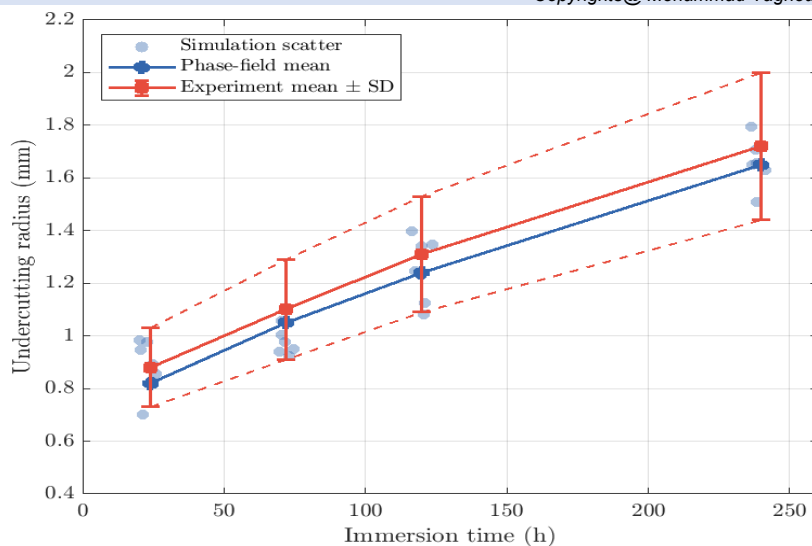


Figure 5: Blister radius growth.

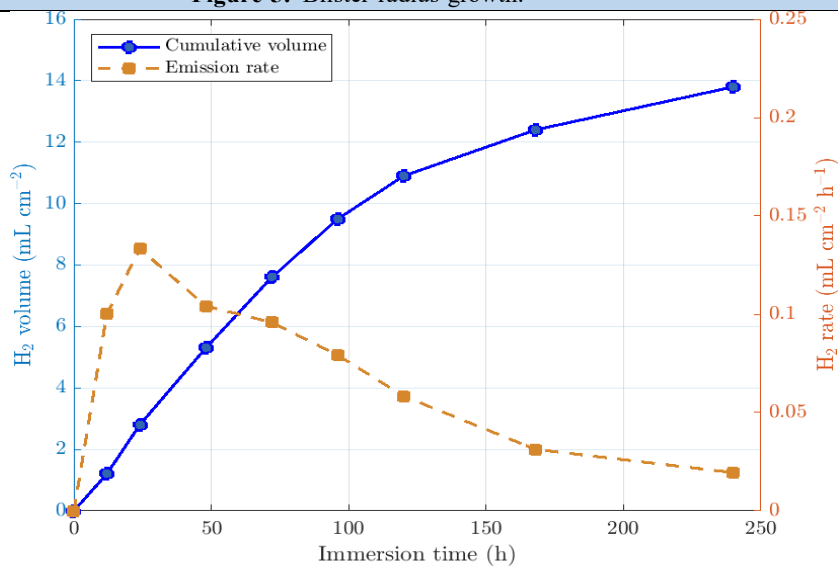


Figure 6: Hydrogen cumulative volume and rate.

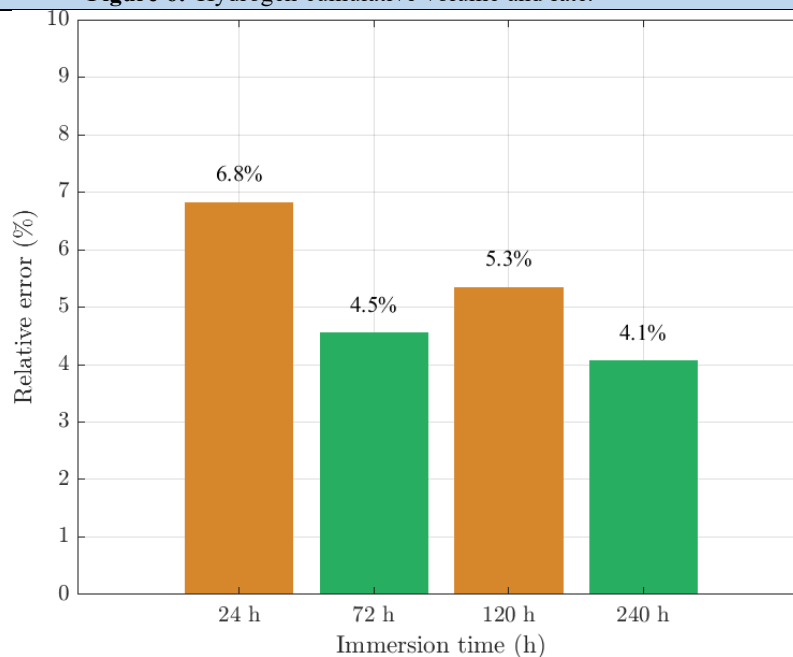


Figure 7: Relative error of phase-field model for blister radius.

3.3 Model Validation: Triangular Pitting

The raw experimental measurements of pit morphology, hydrogen evolution, and the quantitative comparison between simulation and experiment are provided in Tables 9–11 (adapted from [61]). Pitting corrosion simulations employed the spatially dependent mobility function with parameters chosen to enforce three-fold symmetry. The experimentally measured pit depths, including average and maximum values across multiple samples, are compiled in Table 9, which also reports the number of pits and pitting factors for each specimen. Twenty-seven two-dimensional simulations were carried out using nine combinations of N – β pairs ($N = 1.25$ – 2.5 , $\beta = 0.1$ – 1.5) and three amplitude ranges ($0.1L_0 \leq L' \leq 10L_0$). Predicted cross-sectional pit patterns were compared with experimental images. The simulated uniform

corrosion radius ($107 \pm 22 \mu\text{m}$) and average pit depth ($25 \pm 13 \mu\text{m}$) agreed well with experimental values of $108 \pm 21 \mu\text{m}$ and $26 \pm 12 \mu\text{m}$, respectively. Maximum pit depth was slightly underestimated, consistent with the idealised equilateral triangle assumption, which cannot capture the full variability in pit apex sharpness observed experimentally. The complete set of hydrogen gas evolution measurements, along with the corresponding pit geometry statistics for each of the 21 samples, is provided in Table 10. Overall, the agreement validates both the uniform and pitting corrosion capabilities of the model. A direct comparison between the experimental and simulated values for uniform circle radius, maximum pit depth, and average pit depth is presented in Table 11, together with the time-dependent hydrogen evolution data used for model calibration. Quantitative experimental data from reference [61] are summarised in Tables 9–11.

S.no	perfect_radius	max_pit_depth_2d	average_pit_depth_2d	nb_pits	pitting_factor_all	pitting_factor
1	110.403	50.089	20.147	7	2.486	2.415
2	135.233	97.362	38.208	5	2.548	2.163
3	119.031	53.584	26.348	4	2.034	1.134
4	89.757	49.883	23.120	4	2.158	1.558
5	124.949	86.353	44.271	4	1.951	1.593
6	129.997	85.618	40.969	2	2.090	1.647
7	118.178	62.201	27.221	4	2.285	1.781
8	88.966	56.486	24.710	5	2.286	1.599
9	104.598	65.393	31.187	4	2.097	2.668
10	84.064	53.465	24.796	3	2.156	1.163
11	94.873	56.802	22.146	4	2.565	2.307
12	119.179	40.073	19.428	6	2.063	1.534
13	72.312	61.931	30.922	3	2.003	2.563
14	59.057	40.451	23.355	2	1.732	1.390
15	93.802	38.708	17.622	5	2.197	1.940
16	113.349	36.366	12.682	6	2.867	1.745
17	98.632	40.844	20.828	7	1.961	1.406
18	134.469	46.583	15.716	5	2.964	2.334
19	137.322	31.343	7.139	10	4.390	3.037
20	125.529	93.368	58.229	4	1.603	1.371
21	111.935	42.049	16.820	6	2.500	1.819
Average	107.887	56.617	25.994	4.762	2.330	1.865
SD	21.438	19.285	11.107	1.895	0.512	0.576

Table 9: Average and max pit depth calculated average and standard deviation for each pit case [61].

Time (hrs)	H ₂ evolution (mL/cm ²)	SD
0	–	–
1.4	0.5873	0.0641
3.1	1.5018	0.0902
5.2	2.6197	0.0790
6.5	3.2560	0.1190
22	6.9658	0.3036
24	7.1248	0.3548
93.75	10.4032	1.54
119.5	11.1738	1.54

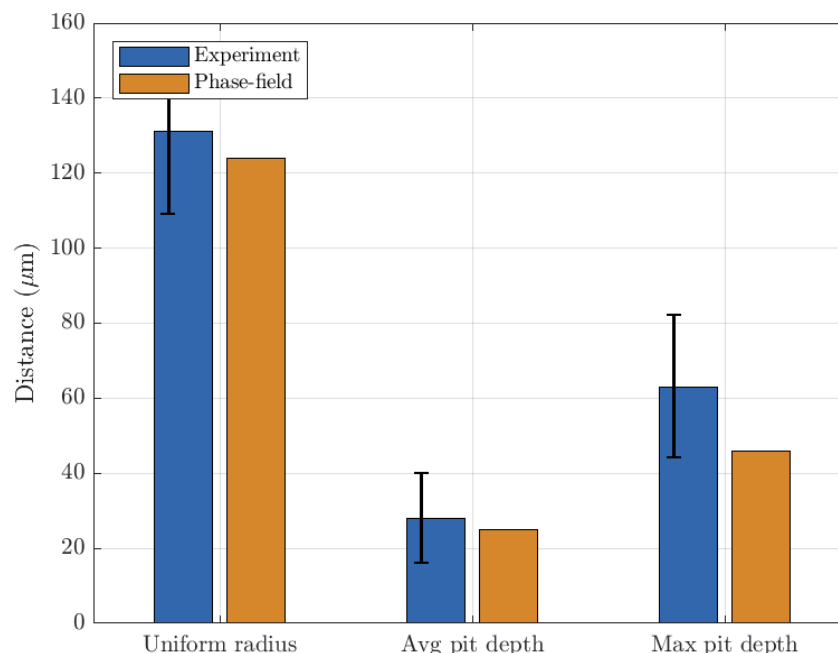
Table 10: Average and standard deviation for Hydrogen gas evolution data Experimental data [61]

Pitting Parameter	Experimental (Avg)	Experimental (SD)	Simulation (Avg)	Simulation (SD)
Uniform circle radii	107.89	21.44	110.57	12.78
Max pit	56.62	19.30	42.16	7.04
Avg pit depth	25.99	11.68	26.01	8.16

Table 11: Pitting parameters [61]

Table 9 lists individual pit depth statistics, Table 10 provides the hydrogen evolution results and associated pit parameters for all tested samples, and Table 11 presents the average values and standard deviations used to validate the present phase field model, showing good agreement between simulation and experiment. Key pitting geometry metrics after 120 h of immersion, including average and maximum pit depths, are summarised in Figure 8. Figure 8 compares experimental and phase-field metrics after 120 h immersion. The model captures the uniform radius, average pit depth, and

maximum pit depth within experimental error bars, demonstrating quantitative predictive capability. Figure 9 shows Pit growth rate. Figure 6 shows the instantaneous pit-depth growth rate. For $\varepsilon_{\infty} = 0.1\%$ (red dotted), the rate peaks early ($>30 \mu\text{m h}^{-1}$) and then declines, while lower strains exhibit a slower rise – reflecting exhaustion of local chemical driving forces. Figure 10 presents Excess mass loss due to mechanical loading. Figure 10 isolates the excess mass loss $\Delta\Delta n/n_0 = (\Delta n/n_0)_{\text{pitting}} - (\Delta n/n_0)_{\text{uniform}}$. The quantity grows monotonically, reaching ~ 0.09 after 24 h, confirming that stress amplifies material loss over time.

**Figure 7:** Relative error of phase-field model for blister radius.

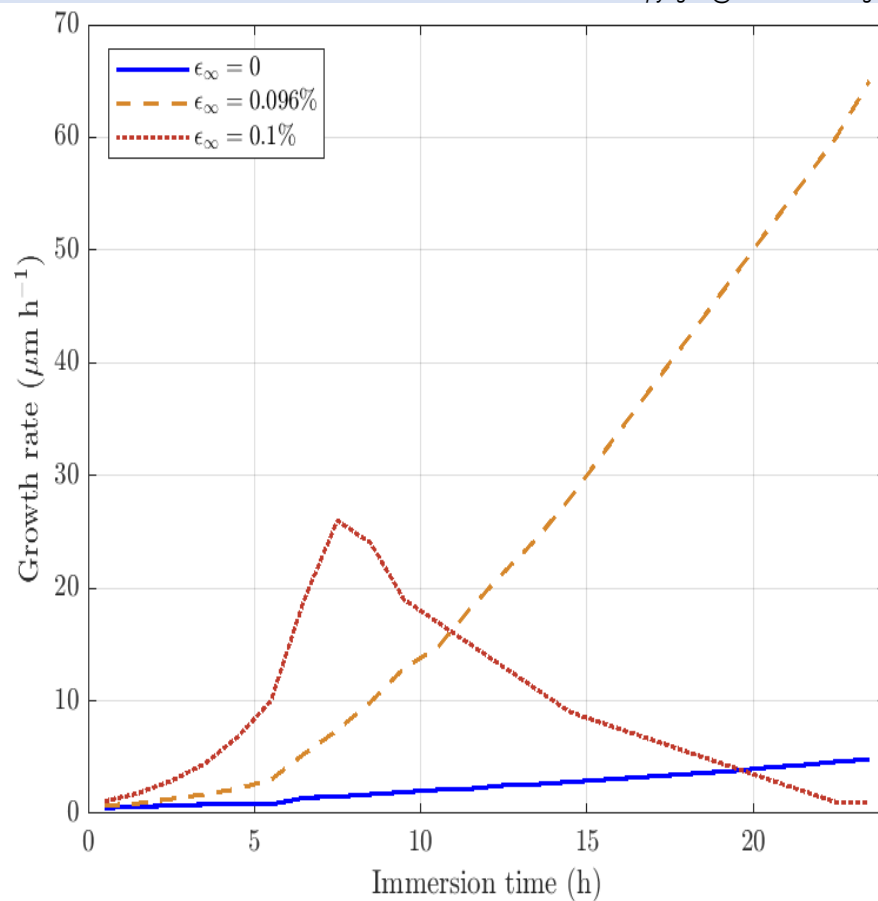


Figure 8: Pitting geometry metrics after 120h.

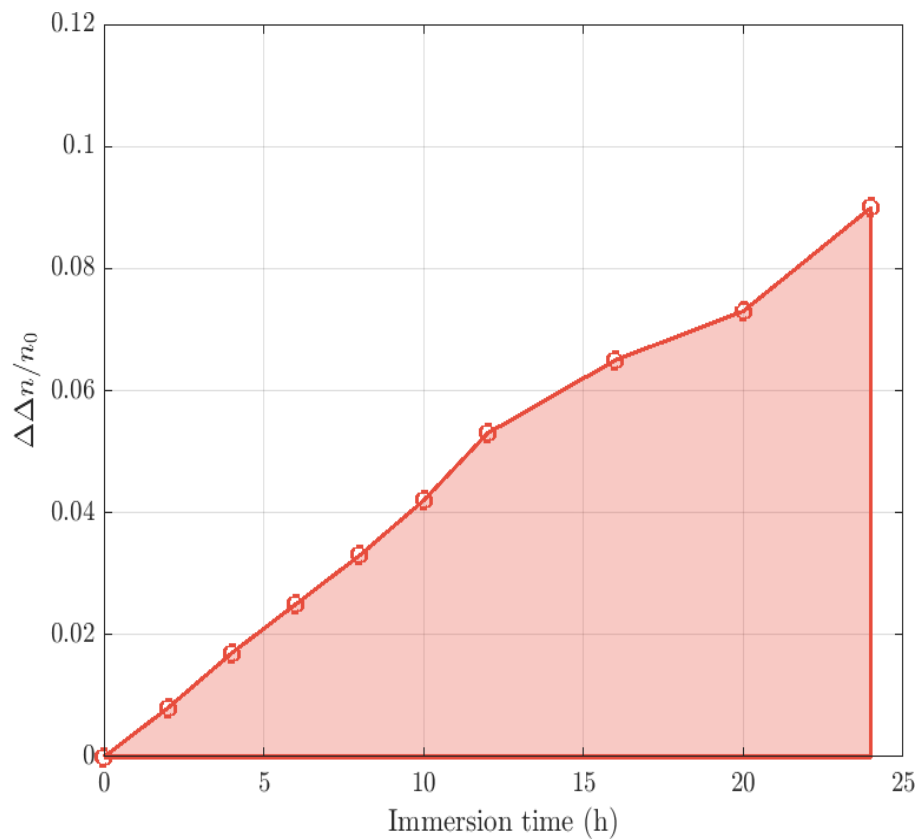


Figure 09: Excess mass loss due to mechanical loading.

4. Applications

4.1 Stress-Assisted Corrosion of Mg Wires with a Triangular Pit

A 300 μm -diameter Mg wire immersed in c-SBF was modelled with an initial surface triangular pit (depth 10 μm , apex full-angle 60°) created by local passive-film breakdown. Owing to the axisymmetric geometry, only a half-domain was simulated. No-flux Neumann conditions were imposed on all outer boundaries for both the phase-field and the Mg concentration. A thin impermeable protective film of 0.5 μm thickness surrounded the wire except at the pit location. Remote tensile strains ε_∞ of 0 %, 0.096 %, and 0.1 % were prescribed at the top boundary and held fixed throughout each simulation. Mechanical properties of AZ31 Mg alloy were adopted: Young's modulus $E = 44.8$ GPa, yield stress 138 MPa, ultimate tensile strength 245 MPa at 17 % engineering strain, described by J_2 plasticity with non-linear isotropic hardening.

In the absence of mechanical loading, the triangular pit deepens while maintaining its faceted shape, driven purely by anisotropic dissolution kinetics. When a small tensile strain ($\varepsilon_\infty = 0.096$ %) is applied, significant

hydrostatic stress develops at the triangular apex, locally amplifying the interfacial mobility. The Mg^{2+} concentration rises steeply at the apex, the pit elongates into a sharp, crack-like feature, and a clear pit-to-crack transition is initiated. At $\varepsilon_\infty = 0.1$ %, plastic yielding occurs ahead of the apex; the plastic strain further amplifies dissolution and crack propagation becomes rapid.

Comparison with a circular pit of equal depth (10 μm) under identical loading conditions reveals that the triangular geometry produces a peak hydrostatic stress approximately 2.4 times higher. This translates to a 35–45 % shorter time to pit-to-crack transition under the same applied strain, highlighting the critical importance of accounting for realistic pit morphologies when estimating implant service life. The amplification of the anodic current density as a function of hydrostatic stress and plastic strain is depicted in Figure 11. Figure 11 quantifies the mechano-electrochemical amplification factor L/L_0 as a function of effective plastic strain ε^p for three hydrostatic stress levels ($\sigma_h = 50, 100, 200$ MPa). The amplification follows a stress- and strain-dependent exponential law, showing that both plastic strain and hydrostatic stress synergistically increase the reaction rate.

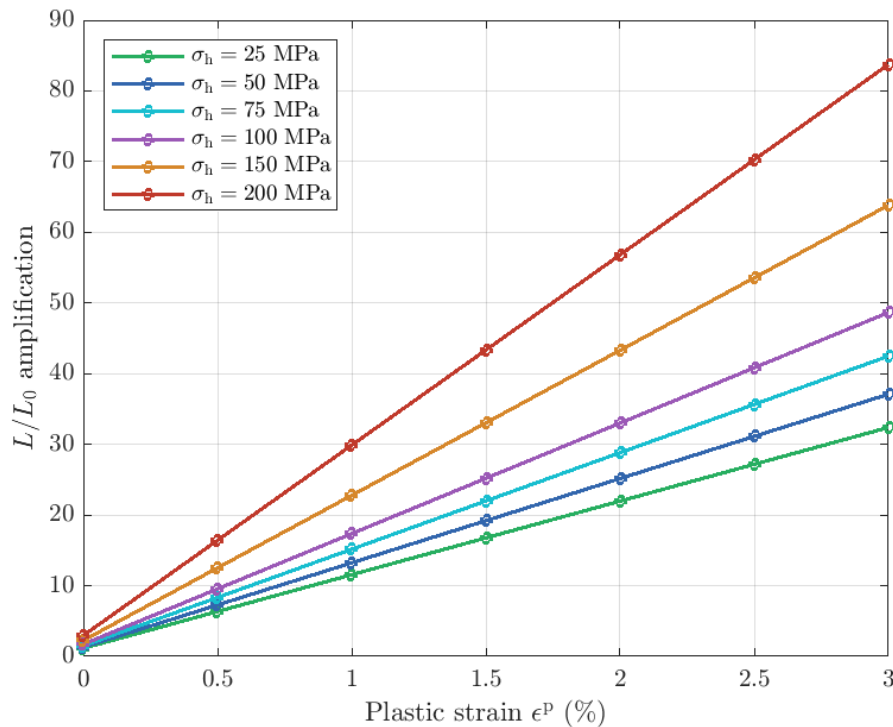


Figure 10: Amplification surface for varying hydrostatic stress.

4.2 Degradation of a Bioabsorbable Coronary Mg Stent

An idealised stent geometry — six rings of sinusoidal struts (outer diameter 2.0 mm, total length 7.5 mm, strut diameter 0.15 mm) connected by links (diameter 0.125 mm, length 0.30 mm) — was radially expanded to 2.25 mm to simulate balloon inflation, then allowed to recoil to 2.168 mm. The resulting plastic strain distribution, concentrated at ring–link junctions, was incorporated as the initial condition for the corrosion simulation. The surrounding physiological environment was assigned an initial Mg^{2+} concentration of 0.875 mmol/L, representative of human blood plasma.

Two scenarios were compared: (i) pitting corrosion, in which triangular pits were seeded at the high-plastic-strain ring–link junctions using the anisotropic mobility function with three-fold symmetry; and (ii) uniform corrosion, with a spatially constant mobility and no initial pits, serving as the reference case. Results after 24 h of immersion show that triangular pitting nucleates

immediately at the plastic-strain hot-spots and progresses at a much higher local dissolution rate than the surrounding strut surfaces. The mass-loss ratio at 24 h is 31 % for triangular pitting versus 18 % for uniform corrosion. More critically, the minimum remaining cross-section at the ring–link junctions is reduced by 60 % relative to the initial geometry under pitting, compared with only 30 % thinning for uniform corrosion. Consequently, uniform corrosion models overestimate the structural service life of the stent by a factor approaching two when triangular pitting is the operative mechanism. Figure 12 displays the phase field variable, the damage indicator, and the von Mises stress contour at the apex of a triangular pit under tensile loading. The mass loss ratio after 24 h of immersion is compared between triangular pitting and uniform corrosion in Figure 13. Figure 13 isolates the excess mass loss $\Delta\Delta n/n_0 = (\Delta n/n_0)_{\text{pitting}} - (\Delta n/n_0)_{\text{uniform}}$. The quantity grows monotonically, reaching ~ 0.09 after 24 h, confirming that stress amplifies material loss over time.

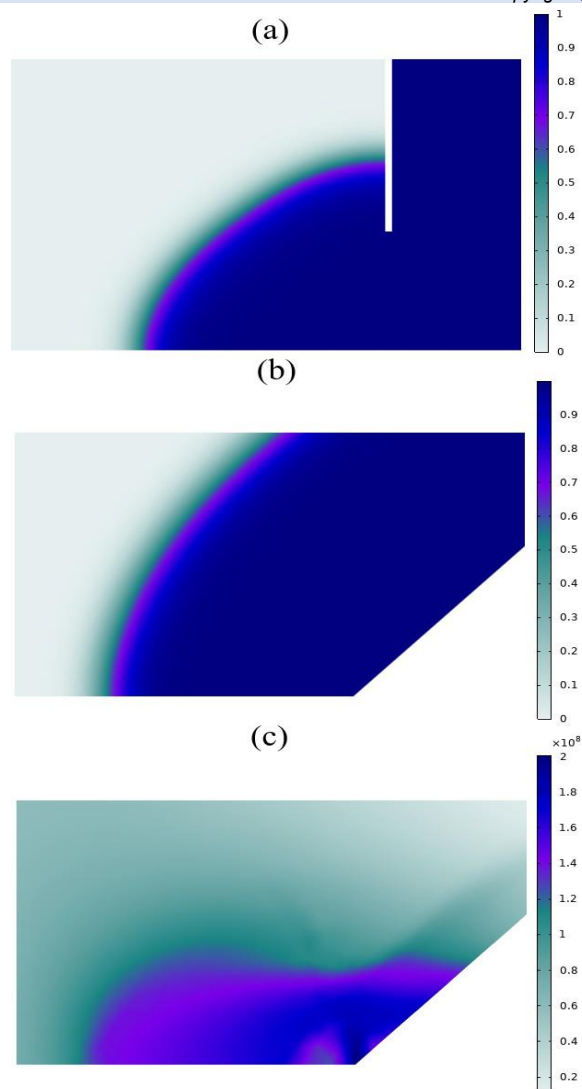


Figure 12: (a) phase function (ϕ). (b) damage (c) von mises stress

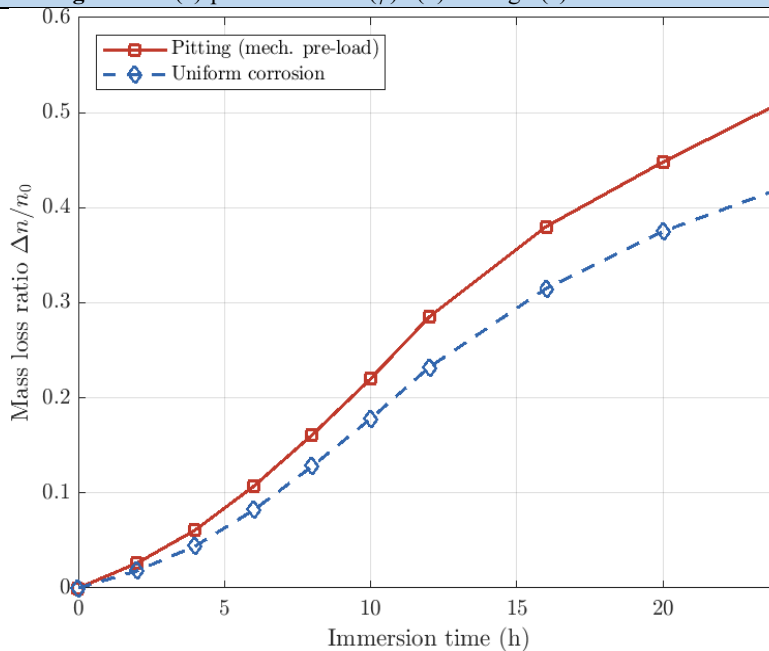


Figure 13: Mass-loss ratio for pitting and uniform corrosion.

5. Discussion

Why Triangular Pits Are More Dangerous Than Circular Pits

The sensitivity of the pit depth to the equilibrium liquid phase concentration $c^{l,eq}$ is illustrated in Figure 14. The hydrostatic stress at the tip of a pit scales approximately as $\sigma_h \propto 1/\rho$, where ρ is the local radius of curvature. A circular pit of radius r has $\rho = r$ uniformly along its boundary. A triangular

pit with the same nominal depth has an apex radius $\rho \ll r$, set by the diffuse interface thickness rather than by r . Consequently, the mechano-electrochemical amplification factor is dramatically higher at the apex, producing a strongly localised dissolution zone. This self-reinforcing mechanism — high stress drives faster dissolution, which sharpens the apex further, which raises the stress — explains the substantially shorter pit-to-crack transition times observed in the triangular pit simulations.

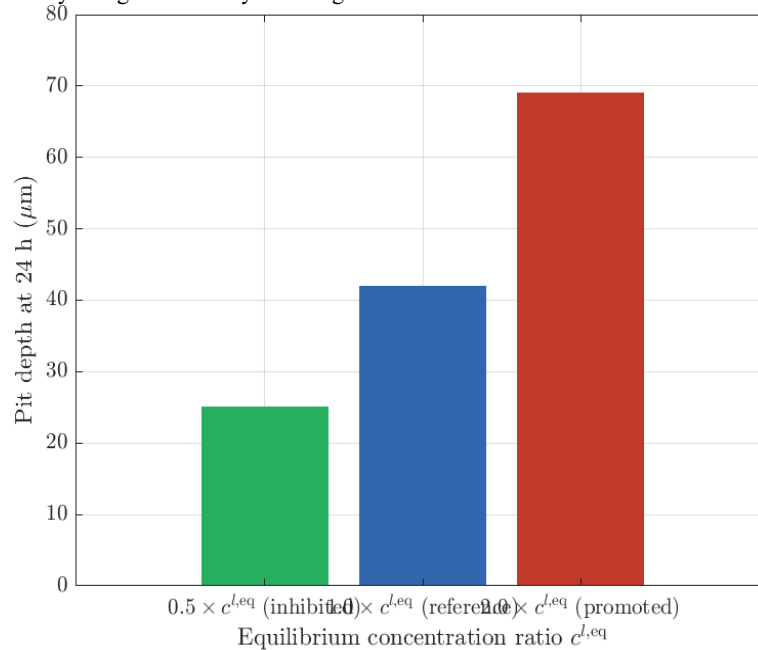


Figure 14: Effect of $c^{l,eq}$ on pit depth.

Role of the Spatially Dependent Mobility Function

The pitting function was designed to encode three-fold symmetry consistent with preferential dissolution along three equivalent prismatic or pyramidal families in the hexagonal Mg lattice. The amplitude parameters control the contrast between rapidly and slowly dissolving surface patches; higher

contrast sharpens the pit corners and increases pit depth variability, consistent with the broader standard deviations observed experimentally. Although not the primary focus of this work, the influence of an external UV dose on pit depth is explored in Figure 15. Figure 15 shows that UV irradiation accelerates pitting: a linear sensitivity $\alpha_{UV} = 0.4$ raises the pit depth from 38 μm (no UV) to $\sim 53 \mu\text{m}$ at 2000 kJ m^{-2} .

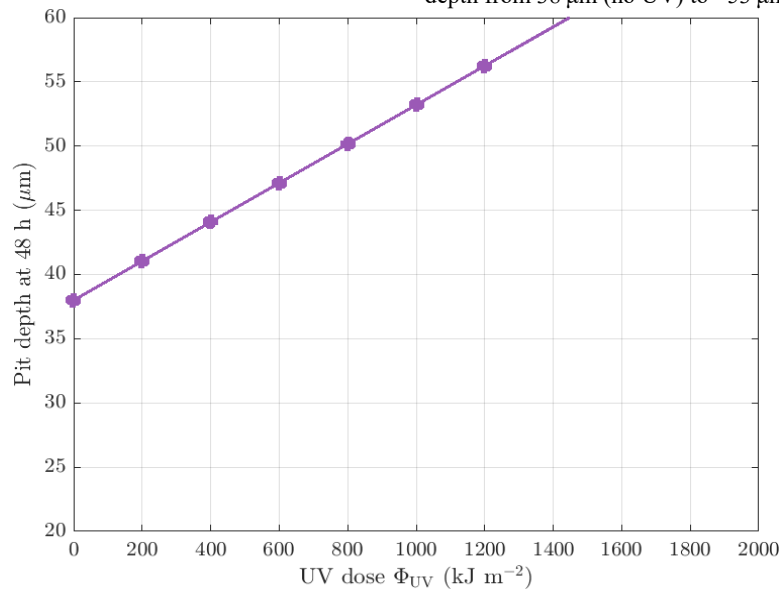


Figure 15: UV dose effect on pit depth.

Rate-Limiting Regime and Equilibrium Concentration

All simulations operate in the activation-controlled regime, meaning that Mg ions diffuse away from the pit faster than they are produced by dissolution.

Raising the equilibrium liquid-phase concentration promotes corrosion by reducing the driving force for protective-layer precipitation, while lowering it retards corrosion — providing a lever to account for surface treatments that enhance passivation.

Limitations

The present model assumes an idealised equilateral triangular pit; real pits exhibit variability in apex angle and asymmetry. The electric-field contribution to ionic transport and the explicit modelling of the $\text{Mg}(\text{OH})_2$ product layer are neglected, as are microstructural features such as grain

boundaries and alloying-element segregation. Addressing these limitations — particularly the electrochemistry–corrosion interplay and multi-pit interactions under complex three-dimensional stress states — constitutes the principal focus of future work. Figure 16 shows the concentration sensitivity of the pit depth, highlighting the role of Mg^{2+} transport in the corrosion process.

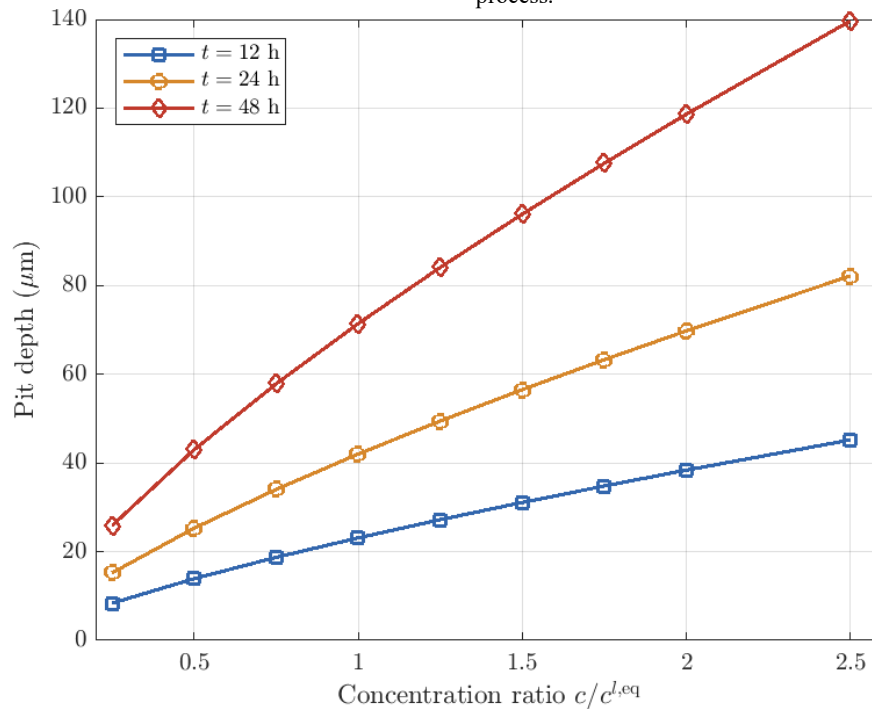


Figure 16: Concentration sensitivity of pit depth

Design Implications

The results suggest that manufacturing routes which minimise basal texture — such as equal-channel angular pressing or biaxial rolling — could suppress triangular pit formation and extend implant service life. Surface treatments that maintain the integrity of the passive film at plastically strained locations, such as micro-arc oxidation, are particularly valuable for stent applications where deployment-induced plastic strains are unavoidable.

6. Conclusions

A diffuse-interface phase-field model has been developed for the coupled simulation of triangular pitting corrosion and mechanically-assisted degradation of biodegradable Mg alloys in physiological environments. The following specific conclusions are drawn:

Triangular pitting morphology is incorporated for the first time into a phase-field corrosion model through a spatially dependent interfacial mobility function with three-fold crystallographic symmetry, consistent with preferential dissolution along prismatic and pyramidal planes in textured Mg alloys.

Model predictions for uniform corrosion radius, average pit depth, and hydrogen gas evolution agree well with dedicated in vitro experiments on WE43 Mg alloy wires immersed in c-SBF, thereby validating the calibrated framework against quantitative experimental data.

Under tensile loading, the apex of a triangular pit concentrates hydrostatic stress approximately 2.4 times more intensely than a circular pit of equal depth, reducing the time to pit-to-crack transition by 35–45 %. The mechano-electrochemical coupling produces a self-reinforcing dissolution mechanism that drives rapid crack growth once initiated.

In a bioabsorbable coronary stent, triangular pitting nucleated at plastic-strain hot-spots generated during balloon deployment causes local strut thinning approximately twice as severe as uniform corrosion models would predict, indicating that neglecting triangular pit morphology leads to potentially unsafe overestimates of device service life.

The framework is general and extendable to other biodegradable metals (Fe, Zn) and to three-dimensional geometries with complex loading histories, providing a cost-effective computational tool for the mechanistic design of next-generation bioabsorbable metallic implants.

Data Availability

The data supporting the findings of this study are publicly available in the Zenodo repository under DOI: 10.5281/zenodo.20202216.

References

1. F. Witte, The history of biodegradable magnesium implants: a review, *Acta Biomater.* 6 (2010) 1680–1692.
2. D. Bairagi, S. Mandal, A comprehensive review on biocompatible Mg-based alloys as temporary orthopaedic implants, *J. Magnes. Alloy.* 10 (2022) 627–669.
3. Y. Zheng, X. Gu, F. Witte, Biodegradable metals, *Mater. Sci. Eng. R Rep.* 77 (2014) 1–34.
4. H. Windhagen et al., Biodegradable magnesium-based screw clinically equivalent to titanium screw in hallux valgus surgery, *Biomed. Eng. Online* 12 (2013) 62.
5. H.M. Garcia-Garcia et al., In vivo serial imaging of the second-generation drug-eluting absorbable metal scaffold (Magmaris), *Int. J. Cardiol.* 255 (2018) 22–28.

6. D. Xia et al., Research status of biodegradable metals for oral and maxillofacial applications, *Bioact. Mater.* 6 (2021) 4186–4208.
7. N. Kirkland, N. Birbilis, M. Staiger, Assessing the corrosion of biodegradable magnesium implants: a critical review, *Acta Biomater.* 8 (2012) 925–936.
8. F. Witte et al., In vitro and in vivo corrosion measurements of magnesium alloys, *Biomaterials* 27 (2006) 1013–1018.
9. Krause et al., Degradation behaviour and mechanical properties of magnesium implants in rabbit tibiae, *J. Mater. Sci.* 45 (2010) 624–632.
10. S. Fischerauer et al., In vivo degradation performance of micro-arc-oxidized magnesium implants: a micro-CT study in rats, *Acta Biomater.* 9 (2013) 5411–5420.
11. J.-M. Seitz et al., Comparison of the corrosion behaviour of coated and uncoated Mg alloys in an in vitro environment, *Adv. Eng. Mater.* 13 (2011) B313–B323.
12. Y. Xin, T. Hu, P. Chu, In vitro studies of biomedical Mg alloys in simulated physiological environment: a review, *Acta Biomater.* 7 (2011) 1452–1459.
13. W. Ali et al., Strength, corrosion resistance and cellular response of bioresorbable PLA/Mg fibre composites, *J. Mech. Behav. Biomed. Mater.* 123 (2021) 104781.
14. L. Chen et al., Stress corrosion cracking behaviour of biomedical Mg-1Zn alloy in biological media, *Corros. Sci.* 175 (2020) 108876.
15. Y. Luo et al., Clinical translation and challenges of biodegradable magnesium-based interference screws in ACL reconstruction, *Bioact. Mater.* 6 (2021) 3231–3243.
16. K. Chen et al., Effect of strain on degradation behaviours of WE43, Fe and Zn wires, *Acta Biomater.* 113 (2020) 627–645.
17. Y. Zheng et al., Effects of tensile and compressive deformation on corrosion behaviour of a Mg–Zn alloy, *Corros. Sci.* 90 (2015) 445–450.
18. Y. Gao et al., Effect of stress on corrosion of high-purity Mg in vitro and in vivo, *Acta Biomater.* 83 (2019) 477–486.
19. L. Choudhary, R.S. Raman, Mg alloys as body implants: fracture under dynamic and static loading, *Acta Biomater.* 8 (2012) 916–923.
20. P. Han et al., In vitro and in vivo studies on degradation of high-purity Mg screw, *Biomaterials* 64 (2015) 57–69.
21. H. Hornberger, S. Virtanen, A. Boccaccini, Biomedical coatings on Mg alloys: a review, *Acta Biomater.* 8 (2012) 2442–2455.
22. N. Eliaz, Corrosion of metallic biomaterials: a review, *Materials* 12 (2019) 407.
23. L. Chen et al., Stress corrosion cracking of Mg alloy in synthetic or natural biological media, *Corros. Sci.* 175 (2020) 108876.
24. K. Chen et al., Effect of strain on degradation behaviours, *Acta Biomater.* 113 (2020) 627–645.
25. J. Lemaitre, R. Desmorat, *Engineering Damage Mechanics*, Springer, 2005.
26. D. Gastaldi et al., Continuum damage model for bioresorbable Mg alloy stents, *J. Mech. Behav. Biomed. Mater.* 4 (2011) 352–365.
27. J. Grogan et al., A corrosion model for bioabsorbable metallic stents, *Acta Biomater.* 7 (2011) 3523–3533.
28. N. Debusschere et al., A computational framework to model degradation of biocorrosible metal stents, *Ann. Biomed. Eng.* 44 (2016) 382–390.
29. E. Galvin et al., A strain-mediated corrosion model for bioabsorbable stents, *Acta Biomater.* 55 (2017) 505–517.
30. Y. Gao et al., A quantitative study on Mg alloy stent biodegradation, *J. Biomech.* 74 (2018) 98–105.
31. S. Ma, B. Zhou, B. Markert, Numerical simulation of tissue differentiation and corrosion of Mg implants, *ZAMM* 98 (2018) 2223–2238.
32. K. van Gaalen et al., Linking pitting corrosion with mechanical integrity of Mg alloy implants, *Bioact. Mater.* 21 (2023) 32–43.
33. J. Sanz-Herrera, E. Reina-Romo, A. Boccaccini, In silico design of Mg implants, *J. Mech. Behav. Biomed. Mater.* 79 (2018) 181–188.
34. M. Marvi-Mashhadi et al., Simulation of corrosion of additively manufactured Mg scaffolds, *J. Mech. Behav. Biomed. Mater.* 126 (2022) 104881.
35. J. Grogan, S. Leen, P. McHugh, A physical corrosion model for bioabsorbable metal stents, *Acta Biomater.* 10 (2014) 2313–2322.
36. Z. Shen et al., Predicting Mg degradation with a diffusion-based model, *J. Mater. Sci. Technol.* 35 (2019) 1393–1402.
37. B. Zeller-Plumhoff et al., Computational modelling of Mg degradation in simulated body fluid, *J. Magnes. Alloy.* (2021).
38. P. Bajger et al., Mathematical modelling of biodegradable metal degradation, *Biomech. Model. Mechanobiol.* 16 (2017) 227–238.
39. A.-K. Gartzke et al., A simulation model for degradation of Mg bone implants, *J. Mech. Behav. Biomed. Mater.* 101 (2020) 103411.
40. M. Barzegari et al., Computational modeling of biodegradable Mg biomaterials, *Corros. Sci.* 190 (2021) 109674.
41. Hermann et al., Combining peridynamic and FE simulations to capture corrosion of bone implants, *Int. J. Mech. Sci.* 220 (2022) 107143.
42. L.Q. Chen, Phase-field models for microstructure evolution, *Annu. Rev. Mater. Res.* 32 (2002) 113–140.
43. W. Mai, S. Soghrati, R. Buchheit, A phase-field model for simulating pitting corrosion, *Corros. Sci.* 110 (2016) 157–166.
44. W. Mai, S. Soghrati, Phase-field model for stress corrosion cracking from pits, *Corros. Sci.* 125 (2017) 87–98.
45. C. Cui, R. Ma, E. Martínez-Pañeda, A phase-field formulation for dissolution-driven stress corrosion cracking, *J. Mech. Phys. Solids* 147 (2021) 104254.
46. C. Lin, H. Ruan, S.Q. Shi, Phase-field study of mechano-electrochemical corrosion, *Electrochim. Acta* 310 (2019) 240–255.
47. C. Lin, H. Ruan, Multi-phase-field modeling of galvanic pitting, *Corros. Sci.* 177 (2020) 108900.
48. C. Lin, H. Ruan, Phase-field modeling of mechano-chemical-coupled SCC, *Electrochim. Acta* 395 (2021) 139196.
49. T. Ansari et al., Phase-field model of pitting corrosion kinetics in metallic materials, *npj Comput. Mater.* 38 (2018) 382–390.
50. T.-T. Nguyen et al., Phase-field method for anodic dissolution induced SCC propagation, *Corros. Sci.* 132 (2018) 146–160.
51. C. Xie et al., Stress-corrosion damage localization induced by secondary phases in Mg alloys: phase-field modeling, *J. Magnes. Alloy.* (2022).
52. E. Gutman, *Mechanochemistry of Materials*, Cambridge International Science Publishing, Cambridge, 1998.
53. S.G. Kim, W.T. Kim, T. Suzuki, Phase-field model for binary alloys, *Phys. Rev. E* 60 (1999) 7186–7197.
54. Y. Xin et al., Influence of aggressive ions on degradation of biomedical Mg alloy in physiological environment, *Acta Biomater.* 4 (2008) 2008–2015.

55. S. Kovacevic et al., Interfacial energy as the driving force for diffusion bonding of ceramics, *Acta Mater.* 186 (2020) 405–414.
56. S.Dj. Mesarovic, Physical foundations of mesoscale continua, in: *Mesoscale Models*, CISM Springer, 2019.
57. S.M. Allen, J.W. Cahn, A microscopic theory for antiphase boundary motion, *Acta Metall.* 27 (1979) 1085–1095.
58. W. Ali et al., Bioabsorbable WE43 Mg alloy wires modified by plasma-electrolytic oxidation. Part I: processing and mechanical properties, *Biomater. Adv.* 146 (2023) 213314.
59. W. Ali et al., Bioabsorbable WE43 Mg alloy wires modified by plasma-electrolytic oxidation. Part II: degradation and biological performance, *Biomater. Adv.* 147 (2023) 213325.
60. K. van Gaalen et al., Automated ex-situ detection of pitting corrosion in WE43, *Bioact. Mater.* 8 (2022) 545–558.
61. Kovacevic, S., Ali, W., Martínez-Pañeda, E., & LLorca, J. (2023). Phase-field modeling of pitting and mechanically-assisted corrosion of Mg alloys for biomedical applications. *Acta Biomaterialia*, 164, 641–658. <https://doi.org/10.1016/j.actbio.2023.04.011>



This work is licensed under Creative Commons Attribution 4.0 License

To Submit Your Article Click Here: [Submit Manuscript](#)

DOI: [10.31579/2690-8808/323](https://doi.org/10.31579/2690-8808/323)

Ready to submit your research? Choose Auctores and benefit from:

- fast, convenient online submission
- rigorous peer review by experienced research in your field
- rapid publication on acceptance
- authors retain copyrights
- unique DOI for all articles
- immediate, unrestricted online access

At Auctores, research is always in progress.

Learn more <https://auctoresonline.org/journals/journal-of-clinical-case-reports-and-studies>