

MODELLING OF POWDER MELTING AND SLAG INFILTRATION FOR INGOT CASTING

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During bottom-poured ingot casting, a mineral powder spread over the liquid metal must melt at a controlled rate to protect the steel surface and ensure the proper formation of the ingot skin, while part of the resulting liquid slag infiltrates the gap between the solidifying skin and the mould to provide lubrication. From deposition and melting to infiltration, the powder undergoes a complex thermophysical evolution involving heat transfer, phase transformations and volume shrinkage driven by the large density difference between the cold porous powder and the dense molten slag. Despite its industrial importance, this evolution has never been described in a unified and physically consistent framework. We present a one-dimensional thermochemical front-tracking model that accounts for the melting of a casting powder layer up to its infiltration into the gap between ingot and mould. The melting front is tracked explicitly on a deforming mesh using a Landau–ALE formulation that naturally accommodates the contraction of the domain resulting from the density difference between the powder and the slag. All material properties are derived self-consistently from CALPHAD equilibrium calculations of the industrial powder. The model captures a contraction of approximately 74% of the porous layer upon melting, and its steady-state response agrees closely with an independent analytical solution. Description of infiltration into the mould gap, enforced through a velocity boundary condition tied to the filling kinematics, is ensured through an overall mass conservation in the coupled system.

KEYWORDS: INGOT CASTING – CASTING POWDER – INFILTRATION – CALPHAD – FRONT-TRACKING

INTRODUCTION

In bottom-poured ingot casting of steels and specialty alloys, a mineral casting powder is deposited on the free surface of the rising liquid metal. As the powder heats and melts, it forms a layered structure consisting of a powder layer on top, a mushy slag below, and a liquid slag pool in contact with the metal. The resulting slag layer performs three main functions: thermal insulation, chemical protection against re-oxidation, and lubrication between the solidifying skin and the mould [1]. The rate at which the powder melts and feeds the liquid slag pool directly governs the quality of the ingot skin: a powder that melts too slowly or too rapidly produces ripple marks and associated internal metallurgical defects [1-4]. As the metal front rises during bottom-poured filling, the liquid slag flows under gravity into the narrow gap that develops between the solidifying skin and the mould wall despite the high viscosity of mould slags restricts this flow [1]. This infiltrated slag solidifies against the cold mould wall, forming a thin film that protects the developing skin and provides lubrication during final stripping [1].

Modelling this system raises three main challenges. First, the powder is a multicomponent, multiphase medium whose thermophysical properties, enthalpy, density, and heat capacity, vary strongly and non-linearly over the melting range and must be obtained from CALPHAD thermodynamic equilibrium calculations [5] rather than prescribed empirically. Second, the local phase fractions predicted by CALPHAD as melting proceeds directly govern the powder's apparent density, which increases from $590 \text{ kg}\cdot\text{m}^{-3}$ [6] to nearly $2950 \text{ kg}\cdot\text{m}^{-3}$ [5] as gas pores progressively close, producing a significant volumetric contraction necessitating the development of a front-tracking model on a deformable mesh [7, 8]. Third, the mass lost through infiltration must be consistently accounted for in the melting model to satisfy the overall mass balance. To the authors' knowledge, such a coupling has not yet been quantitatively formulated for bottom-poured ingot casting [1].

The present contribution develops such a model in two stages. In the first stage, a reference thermophysical model of the melting oxide matrix is formulated from coupled energy and mass conservation equations in

Landau coordinates on a deforming mesh. All properties are derived from CALPHAD calculations [5]. In addition, the effective thermal conductivity follows the Russell porous-medium model [9]. Applied to the industrial powder Scorialit VN202-73 [6], the model reproduces the full contraction of the powder layer upon melting while strictly conserving mass. In the second stage, infiltration is introduced as a velocity boundary condition at the bottom of the column, determined from the filling conditions.

Sections 2 and 3 detail the mathematical model and the associated results, respectively; Section 4 presents the conclusions and perspectives, including the planned extension to carbon combustion chemistry, which is known to inhibit powder melting in a counterintuitive manner [1, 10, 11].

2. MATHEMATICAL MODEL

2.1 Physical system and ALE front-tracking formulation

As shown in Fig. 1, the system is approximated by a one-dimensional column above the rising metal front of a bottom-poured ingot, structured from bottom to top into liquid metal lm , liquid and mushy slag grouped as a single dense domain ds , casting powder cp , and gas g . Solids and liquids within ds are treated at local thermodynamic equilibrium. Only the ds/cp interface is tracked, coinciding with the solidus isotherm $T^{ds/cp} = T_s$ [7].

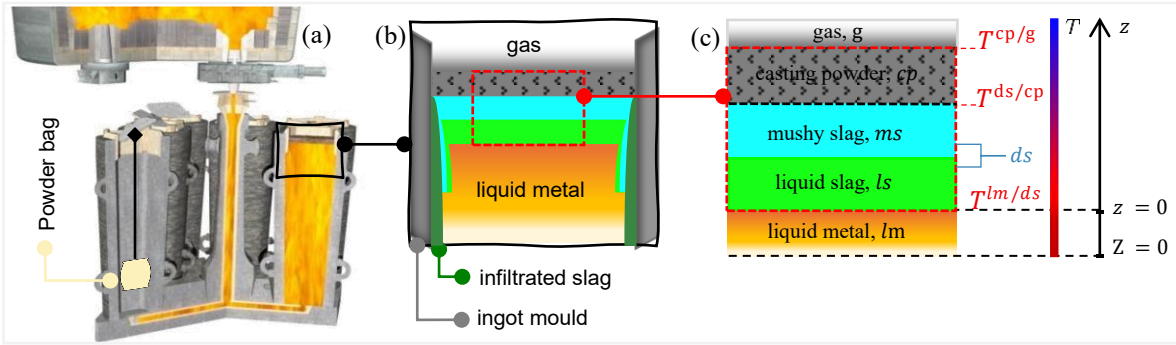


Fig.1 - Schematic representation of the physical system with (a) a global view of the bottom-poured ingot filling operation (inspired from [1]), (b) the distribution of materials during the filling operation, with the infiltrated slag at the periphery of the ingot/mould boundary and (c) the one-dimensional distribution of the casting powder domains during its melting above the liquid metal.

The metal front position $Z(t)$ rises at a constant velocity v_F during the filling stage, then stays fixed at the final filling height when feeding is achieved. The filling time t_F is defined as the time at which this transition occurs:

$$Z(t) = \begin{cases} v_F t, & 0 < t \leq t_F, \\ v_F t_F, & t > t_F \end{cases} \quad (1)$$

As the physical interfaces move as filling and melting proceed, each domain $d \in \{ds, cp\}$ is mapped onto a fixed normalized coordinate $\xi \in [0, 1]$ using a Landau transformation [7]. This transformation converts the moving boundaries into fixed computational limits (Fig. 2). We denote by $z_-^d(t)$ and $z_+^d(t)$ the positions of the lower and upper boundaries of domain d . These positions refer, respectively, to $z^{lm/ds}$ and $z^{ds/cp}$ for $d = ds$, and $z^{ds/cp}$ and $z^{cp/g}$ for $d = cp$:

$$\xi(z, t) = \frac{z - z_-^d(t)}{L^d(t)}, \quad L^d(t) = z_+^d(t) - z_-^d(t) \quad (2)$$

The mesh velocity v_ξ^d , distinct from the material velocity v , is interpolated linearly between the velocities of the lower and upper bounding interfaces, respectively named v_-^d and v_+^d , so that the grid deforms continuously while remaining fixed in the ξ coordinate:

$$v_\xi^d = (1 - \xi)v_-^d + \xi v_+^d \quad (3)$$

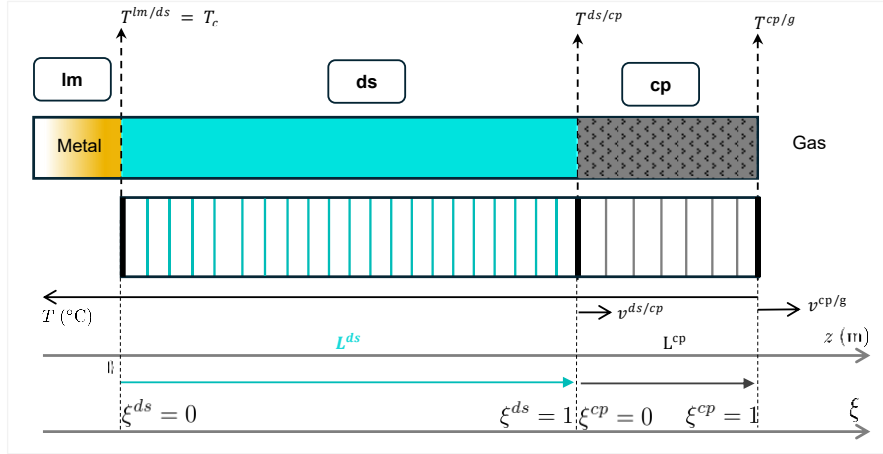


Fig.2 - Landau-ALE mapping of the ds and cp domains onto the normalized coordinate ξ .

In each domain $d \in \{ds, cp\}$ of length L^d , the energy and mass balances are formulated for the intrinsic volume-averaged properties of that domain. Any such domain-averaged property $\langle X^d \rangle$ is obtained from the phase-intrinsic properties $\langle X \rangle^\varphi$ (see §2.2) through

$$\langle X^d \rangle = \sum_{\varphi} g^{\varphi} \langle X \rangle^{\varphi} \quad (4)$$

where g^{φ} is the volume fraction of phase φ within domain d . Applied to enthalpy, this gives the volume-averaged enthalpy $\langle \rho h \rangle^d$, used in the energy balance together with the volume-averaged conductivity $\langle \lambda \rangle^d$; applied to density, it gives $\langle \rho \rangle^d$, used in the mass balance. These balances are first written in the physical coordinate z , then transformed into the Landau coordinate ξ and discretised by a finite-difference scheme [8, 12]. The energy balance in volumetric enthalpy form reads:

$$\frac{\partial \langle \rho h \rangle^d}{\partial t} \Big|_{\xi} - \frac{v_{\xi}^d}{L^d} \frac{\partial \langle \rho h \rangle^d}{\partial \xi} + \frac{1}{L^d} \frac{\partial}{\partial \xi} (\langle \rho h \rangle^d \langle v^d \rangle) - \frac{\langle \lambda^d \rangle}{(L^d)^2} \frac{\partial^2 T^d}{\partial \xi^2} - \frac{1}{(L^d)^2} \frac{\partial \langle \lambda^d \rangle}{\partial \xi} \frac{\partial T^d}{\partial \xi} = 0 \quad (5)$$

Since $\langle \rho h \rangle^d(T)$ is a non-linear function of temperature through the phase fractions $g^{\varphi}(T)$, its expression is linearised at each time iteration before resolution of Eq. (5), with respect to the previously estimated temperature [13].

Similarly, the mass conservation equation is defined in the moving frame as:

$$\frac{\partial \langle \rho^d \rangle}{\partial t} \Big|_{\xi} - \frac{v_{\xi}^d}{L^d} \frac{\partial \langle \rho^d \rangle}{\partial \xi} + \frac{1}{L^d} \frac{\partial}{\partial \xi} (\langle \rho^d \rangle \langle v^d \rangle) = 0 \quad (6)$$

At the internal ds/cp interface, continuity of heat flux and mass conservation are enforced, respectively:

$$\langle \lambda^{ds} \rangle \frac{\partial T}{\partial z} \Big|_{ds/cp} - \langle \lambda^{cp} \rangle \frac{\partial T}{\partial z} \Big|_{ds/cp} - \left((\langle v^{ds} \rangle - v^{ds/cp}) \langle \rho h \rangle^{ds} + (\langle v^{cp} \rangle - v^{ds/cp}) \langle \rho h \rangle^{cp} \right) = 0 \quad (7)$$

$$(\langle \rho^{ds} \rangle - \langle \rho^{cp} \rangle) v^{ds/cp} = \langle v^{ds} \rangle \langle \rho^{ds} \rangle - \langle v^{cp} \rangle \langle \rho^{cp} \rangle \quad (8)$$

At the bottom interface, $z = 0$, the liquid metal at pouring temperature $T^{lm/d}$ imposes a convective heat input, and at the top interface, $z = L$, heat is lost by convection to the ambient gas at temperature $T^{d/g}$:

$$-\langle \lambda^d \rangle \frac{\partial T}{\partial z} \Big|_{z=0} = -h^{lm/d} [T(0, t) - T^{lm/d}] \quad -\langle \lambda^d \rangle \frac{\partial T}{\partial z} \Big|_{z=L} = h^{d/g} [T(L, t) - T^{d/g}] \quad (9)$$

where $h^{lm/d}$ and $h^{d/g}$ are the heat transfer coefficients at the lm/d and d/g interfaces, respectively (Tab.1). At the initial time ($t = 0$), a single cp domain in contact with lm is at rest at the initial and boundary conditions listed in Tab.1:

Tab.1 - Initial and boundary conditions

T_0	L_0^{cp}	g_0^g	$v^{lm/cp}$	$h^{lm/cp}$	$h^{cp/g}$	$T^{lm/cp}$	$T^{cp/g}$
100 °C	0.1 m	0.8	0 m·s ⁻¹	100 W·m ⁻² ·K ⁻¹	100 W·m ⁻² ·K ⁻¹	1500 °C	300 °C

2.2 CALPHAD-based properties and effective thermal conductivity

Tab.2 gives the mineral base composition of the industrial powder Scorialit VN202-73 [6]. Silica (SiO₂) and lime (CaO) are the main constituents; alumina (Al₂O₃), soda (Na₂O) and fluorine (F) are fluxes that lower the melting temperature and the viscosity. The other oxides are minor.

Tab.2 - Mineral base composition of Scorialit VN202-73 (wt%) [6].

SiO ₂	CaO	Al ₂ O ₃	Fe ₂ O ₃	Na ₂ O	MgO	MnO	TiO ₂	F
44.63	33.43	8.05	1.88	5.26	0.73	0.03	0.29	5.70

The thermodynamic equilibrium of the mineral base is precomputed with Thermo-Calc and the TCOX10 database [5], at fixed composition and atmospheric pressure, stepped in temperature. Each step yields the phase volume fractions $g^p(T)$, mass densities $\rho^p(T)$ and specific enthalpies $h^p(T)$, tabulated and combined into the volume-averaged mixture properties for each domain d , $\langle(\rho h)^d\rangle$. Since the gas mass is negligible compared with the condensed phases, the properties averaged over domain cp rely on CALPHAD only for its condensed phases; the sole gas property required is its thermal conductivity, obtained from the correlation of Bird et al. [14]. However, these fractions are discontinuous at each phase transition, the sharpest being at the solidus $T_S \approx 953$ °C. Consequently, the phase fractions are all regularized over a half-width of 3°C with a smoothed Heaviside function [15], thus also limiting the enthalpy derivatives used in the iterative linearization previously mentioned.

Fig.3 shows the resulting volume fractions provided by Thermo-Calc [5]. Melting proceeds through several stages, dominated successively by wollastonite (brown line) and anorthite (dark green line), with the ionic liquid (blue line) becoming dominant only above T_S and complete at the liquidus $T_L \approx 1250$ °C. The present multi-stage path contrasts with the constant or empirically fitted properties generally assumed in the literature [10,11] and consequently provides an enhanced description of the material properties evolution. Fig.4 shows the corresponding enthalpy and density: the enthalpy rises monotonically with a slope change at T_S , while the density decreases from ≈ 2950 kg·m⁻³ at room temperature to ≈ 2500 kg·m⁻³ at 1500 °C.

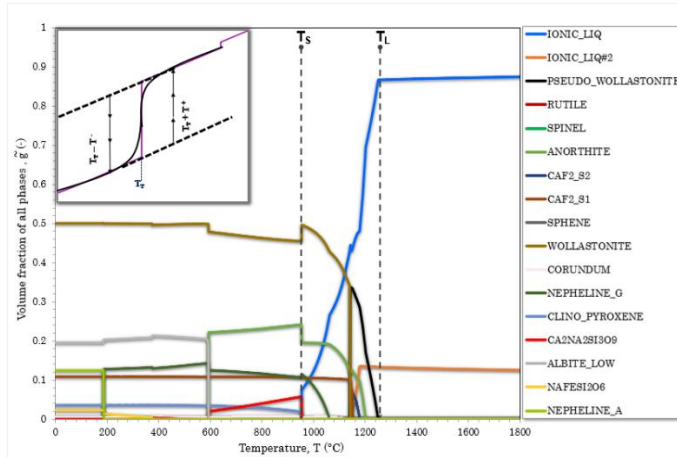


Fig.3 - Equilibrium volume fractions of the stable phases in the casting powder Scorialit VN202-73 (Tab.2) [5]

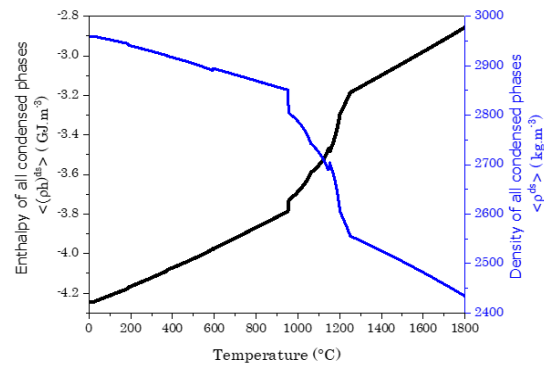


Fig.4 - Averaged volumetric enthalpy and density of the mineral base of Scorialit VN202-73 (Tab.2) [5].

Initially, the entire column is the porous cp domain, with gas confined throughout ($g^g = 0.80$) [6]. Once the solidus isotherm T_s is reached at the interface, the dense domain ds forms. The porosity sharply decreases at T_s until full densification is achieved ($g^g \rightarrow 0$), so cp domain rapidly shrinks while ds domain grows at its expense, as illustrated in Fig.5.

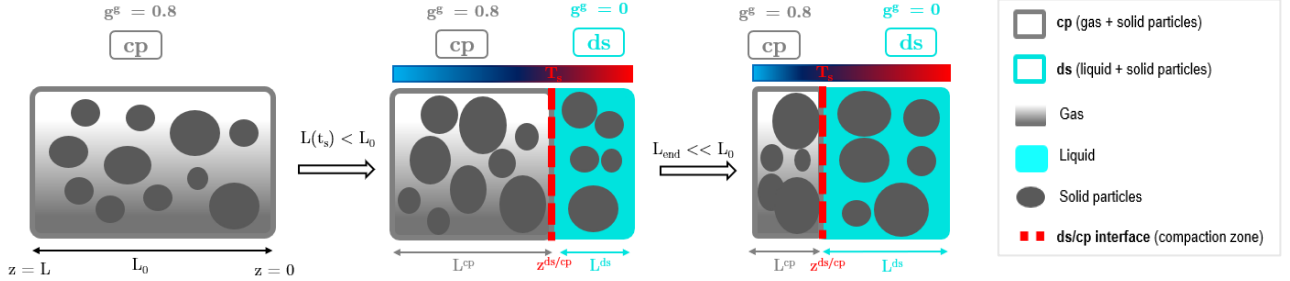


Fig.5 - Schematic illustration of the growth of the dense domain ds at the expense of the porous domain cp during melting of the casting powder ($g^g = 0.80 \rightarrow 0$).

The effective thermal conductivity of the cp domain accounts for this porosity through the Russell model [9]:

$$\langle \lambda^{cp} \rangle = \left[\frac{1 - (g^g)^{1/3}}{\langle \lambda^s \rangle} + \frac{(g^g)^{1/3}}{(1 - (g^g)^{2/3})\langle \lambda^s \rangle + (g^g)^{2/3}\langle \lambda^g \rangle} \right]^{-1} \quad (10)$$

The dense-phase conductivity $\langle \lambda^s \rangle$ is itself a phase-average, analogous to the domain averages defined in (§2.2): $\langle \lambda^s \rangle = \sum_{\varphi} g^{\varphi} \langle \lambda \rangle^{\varphi}$, where g^{φ} is the volume fraction of phase φ within the dense material s . In the absence of phase-resolved conductivity data, the same value is assigned to every phase φ , so that the average reduces to that common value, with a constant dense conductivity $\langle \lambda^s \rangle = 1.5 \text{ W} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$ [10, 16] and a gas conductivity following the power-law correlation of Bird et al. [14], as given by Eq. (11), where $\lambda_0 = 0.026 \text{ W} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$ is the air conductivity at $T_0 = 300 \text{ K}$ and $n = 0.8$ (T in K). The gas volume fraction g^g decreases on melting as the pores fill, following Eq. (12):

$$\langle \lambda \rangle^g(T) = \lambda_0 \left(\frac{T}{T_0} \right)^n \quad (11)$$

$$g^g = 1 - \frac{\langle \rho^{cp} \rangle}{\langle \rho^s \rangle} \quad (12)$$

In addition, the apparent density of the cp domain, $\langle \rho^{cp} \rangle$, is obtained from the dense-material density $\langle \rho^s \rangle$ (from CALPHAD) weighted by the condensed fraction $(1 - g^g)$: $\langle \rho^{cp} \rangle = (1 - g^g) \cdot \langle \rho^s \rangle$. This makes $\langle \lambda^{cp} \rangle$ rise sharply as porosity decreases at T_s . Fig.6 confirms this at the domain level: g^g drops discontinuously from 0.80 to 0 at T_s , density jumps from ≈ 590 to $\approx 2950 \text{ kg} \cdot \text{m}^{-3}$, and conductivity jumps from ≈ 0.3 to $\approx 1.5 \text{ W} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$, a five-fold increase concentrated entirely at the ds/cp interface. Beyond T_s , both quantities are governed solely by the dense ds phase: density decreases gradually with thermal expansion while conductivity remains essentially constant.

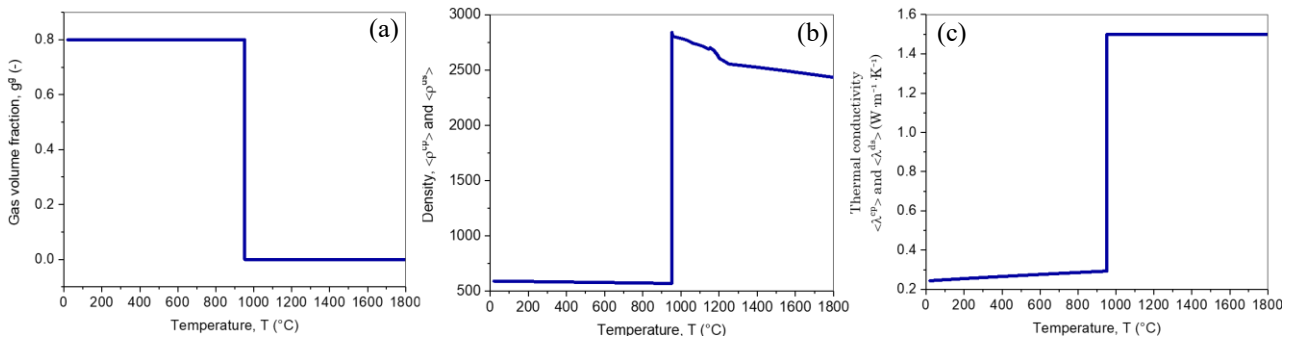


Fig.6 - Properties of the cp and ds domains: (a) porosity, (b) density and (c) thermal conductivity.

3. RESULTS AND VALIDATION

3.1 Bi-domain model with gas porosity

At $t = 0$, the column is a single porous cp domain of length L_0 . Once T_s is reached, ds forms at the interface and grows at the expense of cp , which shrinks as its material densifies across the interface; because the dense phase occupies a fraction of the volume of the porous material it replaces, the column contracts sharply ($L_{\text{end}} \ll L_0$, driven entirely by the density change at the ds/cp interface).

The bi-domain [$ds|cp$] is activated once T_s is reached, and the coupled energy, front-tracking and mass balances are iterated to convergence at each time step ($\Delta t = 1$ s, 2001 nodal points per domain). The porous cp layer contracts from $L_0 = 0.1$ m to $L_{\text{end}} = 0.0259$ m ($\approx 74\%$). This is slightly below the $\approx 80\%$ expected from $g_0^g = 0.8$ alone; the remaining $\approx 6\%$ gap is due to the thermal expansion of the dense phase as it heats above room temperature (Fig.4), which partly offsets the densification. Total system mass is conserved throughout, gas mass being neglected.

Fig.7 shows temperature, density and material velocity profiles at $t = 25$, 1000 and 7000 s. At $t = 25$ s, just prior to reaching the solidus temperature, a temperature plateau near 1000 °C persists, inherited from the initial conditions. The density is mainly that of the powder, except at the boundaries of the domain where dilation takes place due to heating (positive velocity values). At $t = 1000$ s, upon further heating, ds has formed and a temperature gradient has developed across ds . An abrupt change of density marks the interface position, and velocity has decreased to negative values in cp due to the transformation of the powder into the dense domain, while the thickening ds layer raises the thermal resistance to the incoming heat flux. At $t = 7000$ s, temperature is linear across ds , density is stable on both sides of the now-fixed interface, and velocity has dropped to zero throughout the domains: the matter is no longer moving, and the temperature field is set by the boundary conditions, $T(z = 0) \approx 1270.33$ °C at the lm/ds boundary and $T(z = L) \approx 529.65$ °C at the ds/g boundary.

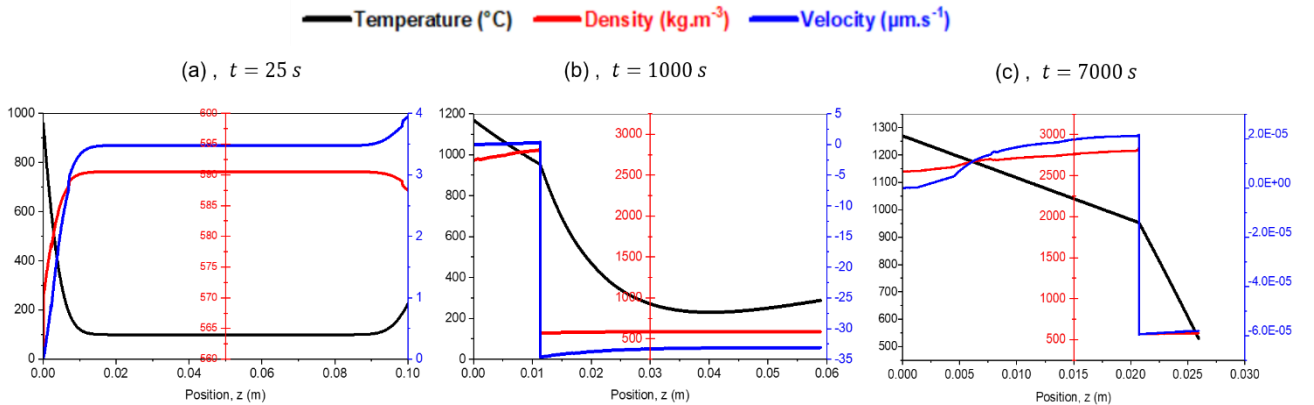


Fig.7 - Time evolution of the bi-domain model with gas porosity: temperature, density and material velocity profiles at (a) $t = 25$ s, (b) 1000 s and (c) 7000 s.

The kinematics of the ds/cp interface reveal three successive regimes (Fig.8a): the velocity is zero before the solidus is reached ($t < 26$ s); it peaks at $15.86 \mu\text{m} \cdot \text{s}^{-1}$ once melting begins (at $t \approx 71$ s), as the thin, newly formed ds layer offers little conductive resistance, $L^{\text{ds}} / \langle \lambda^{\text{ds}} \rangle$, to the heat flux reaching the front. This resistance then grows as ds thickens — even though $\langle \lambda^{\text{ds}} \rangle$ itself remains high and nearly constant — progressively throttling the flux and driving the front velocity close to zero by $t \approx 3000$ -4000 s. Correspondingly (Fig.8b), L^{cp} decreases and L^{ds} increases until both plateau, so that the total column length stabilizes with the thermal field.

As shown in Fig.9, the inlet flux (lm/ds) decreases from its initial peak as ds thickens, while the outlet flux (cp/g) increases from near zero; both converge to the same steady-state value, $q \approx 22965.2 \text{ W} \cdot \text{m}^{-2}$.

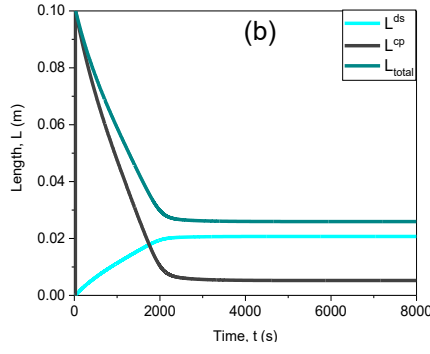
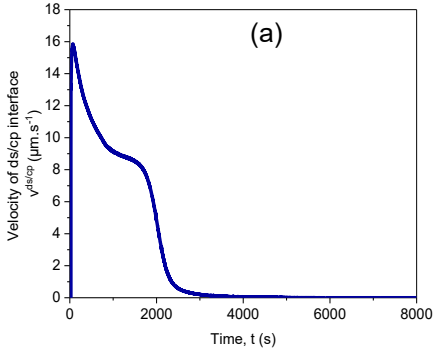


Fig.8 - Kinematics of the ds/cp interface: (a) velocity, and (b) evolution of the ds. cp and total column lengths with time

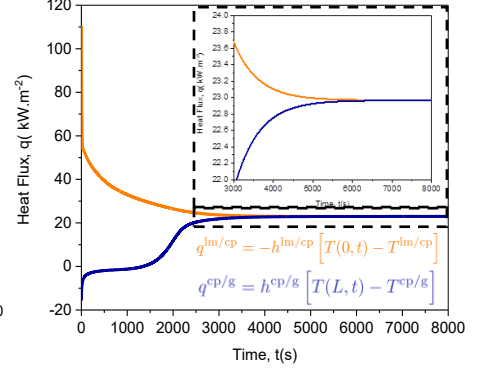


Fig.9 - Inlet and outlet heat flux converging to the analytical steady-state value

3.2 Analytical validation

At steady state, the conductive flux is uniform across the thickness and the series-resistance analytical solution built from the length-weighted harmonic-mean conductivity reproduces the simulation exactly: with $T^{lm/cp} = 1500 \text{ }^{\circ}\text{C}$, $T^{cp/g} = 300 \text{ }^{\circ}\text{C}$ and $h^{lm/cp} = h^{cp/g} = 100 \text{ W} \cdot \text{m}^{-2} \cdot \text{K}^{-1}$ (see Tab.1), the analytical model provides an estimate of the steady-state flux $q = 22948 \text{ W} \cdot \text{m}^{-2}$, $T(0) = 1270.52 \text{ }^{\circ}\text{C}$ and $T(L) = 529.48 \text{ }^{\circ}\text{C}$. These values are in agreement with the simulated steady-state flux of $22965.2 \text{ W} \cdot \text{m}^{-2}$ (Fig.9). The scheme remains conservative throughout the run, with the relative error in mass conservation closing to the order of 10^{-4} and in energy conservation to the order of 10^{-3} at $t = 7000 \text{ s}$.

A more general analytical solution, without assuming a single effective conductivity, provides a stricter test of the model. At steady state the domain splits into a dense zone ds ($[0, z^{ds/cp}]$, $T > T_s$, compacted) and a porous zone cp ($[z^{ds/cp}, L_{end}]$, $T < T_s$, powder). In steady regime, $\partial T / \partial t = 0$ and the flux is constant in the absence of a volumetric source. Assuming that thermal conductivity remains uniform within each domain, temperature evolves linearly in each domain:

$$T^{ds}(z) = T(0) + \frac{T_s - T(0)}{z^{ds/cp}} z, \quad T^{cp}(z) = T_s + \frac{T(L_{end}) - T_s}{L_{end} - z^{ds/cp}} (z - z^{ds/cp}) \quad (13)$$

with boundary conditions $T(0) = T^{lm/ds} - q/h^{lm/ds}$ and $T(L_{end}) = T^{cp/g} + q/h^{cp/g}$, and flux continuity at the interface:

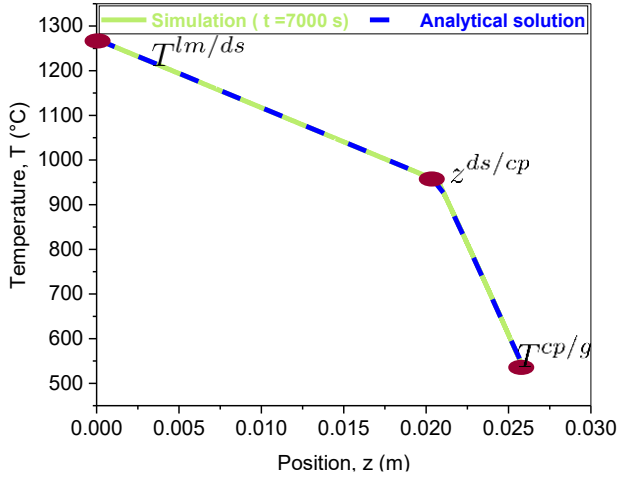
$$q = \langle \lambda^{ds} \rangle \frac{T(0) - T_s}{z^{ds/cp}} = \langle \lambda^{cp} \rangle \frac{T_s - T(L_{end})}{L_{end} - z^{ds/cp}} \quad (14)$$

where $\langle \lambda^{ds} \rangle = \langle \lambda^s \rangle = 1.5 \text{ W} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$ and $\langle \lambda^{cp} \rangle$ follows the Russell model Eq. (9) evaluated at the mean zone temperature. Mass conservation between the initial and final states, closed by the mean density of domain ds over its (linear) temperature profile, gives the total length:

$$L_{end} = z^{ds/cp} + L_0 - \frac{z^{ds/cp} \cdot \langle \rho^{ds} \rangle}{\langle \rho_0^{cp} \rangle} \quad (15)$$

Because $\langle \lambda^{cp} \rangle$ depends on the mean temperature of the cp zone — itself a function of $z^{ds/cp}$ through the zone's extent — and because the mean density of ds also depends on $z^{ds/cp}$, this quantity cannot be calculated directly. Eqs. (13)–(15) are therefore solved step by step: $z^{ds/cp}$ is updated at each step until the mass residual becomes zero.

Table 3 compares the simulation and the analytical solution at $t = 7000 \text{ s}$: the two agree to within $0.2 \text{ }^{\circ}\text{C}$ for temperature and 0.04 mm for the interface position.



Tab.3 - Comparison of the simulation and analytical steady state at $t = 7000$ s.

Quantity	Simulation	Analytical	Deviation
$T(0)$	1270.33 °C	1270.52 °C	0.19 °C (0.01%)
$z^{ds/cp}$	20.726 mm	20.755 mm	0.029 mm (0.14%)
L_{end}	25.949 mm	25.986 mm	0.037 mm (0.14%)
ΔL	74.051 mm	74.014 mm	0.037 mm (0.05%)
$T(L_{end})$	529.65 °C	529.48 °C	0.17 °C (0.03%)

3.3 Infiltration into the mould gap

Beyond feeding the liquid slag pool, the melted powder lubricates the ingot skin against the mould wall: liquid slag in contact with the cold mould solidifies into a thin film that separates the contracting steel shell from the rigid mould during withdrawal [1]. An insufficient film sharply raises solid-state friction and promotes shell tearing [2]; an excessive one causes uneven heat extraction and non-uniform shell growth [1]. The infiltration mechanism modelled here (Fig.10) is the very process that builds this film: liquid slag accumulating in ds is progressively drawn into the mould gap and solidified there, lowering the ds domain's boundary position z^{ds} .

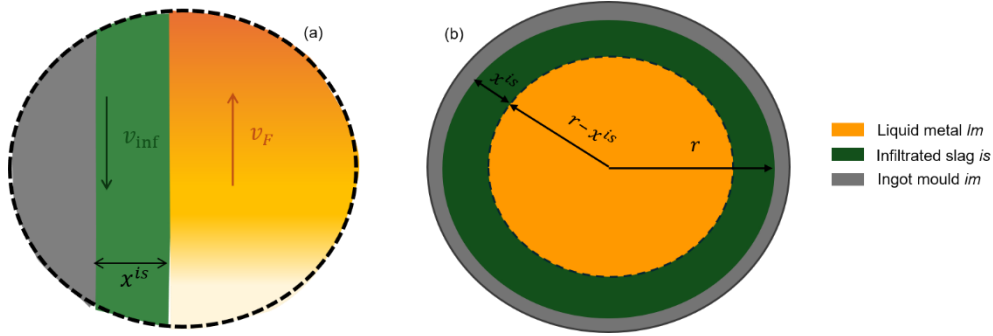


Fig.10 - Schematics of the infiltration mechanism: (a) one-dimensional view showing the infiltration front receding at velocity v_{inf} as the metal front rises at velocity v_F , building up the infiltrated slag layer of thickness x^{is} ; (b) cross-sectional view of the ingot showing melting feeding the ds domain while infiltration removes mass from ds , hence creating the is domain between mould and ingot, with thickness x^{is} .

The infiltration velocity is defined from the solidified film thickness x^{is} at the mould wall:

$$v_{inf} = \left. \frac{dz^{ds}}{dt} \right|_{ds \rightarrow is} \quad (16)$$

The infiltration velocity is defined from the solidified film thickness x^{is} at the mould wall. For an ingot of radius r cast at speed v_F , the resulting drop in the ds column height follows from the annular section $S(t) = \pi(2r \cdot x^{is} - (x^{is})^2)$ drained per casting increment, and reads:

$$\left. \frac{dz^{ds}}{dt} \right|_{ds \rightarrow is} = v_F \frac{S(t)}{\pi r^2} = (2\varphi - \varphi^2) v_F, \quad \varphi = \frac{x^{is}}{r} \quad (17)$$

With the ingot casting radius $r = 335$ and the infiltrated slag thickness $x^{is} = 1.25$ mm [6], Eq. (16) gives an infiltration velocity $v_{inf} = -1.5 \times 10^{-5} \text{ m} \cdot \text{s}^{-1}$, held constant over the filling window $t_F = H_{ingot}/v_F = 2.4/0.002 = 1200$ s for an ingot height $H_{ingot} = 2.4$ m, consistent with Eq. (1).

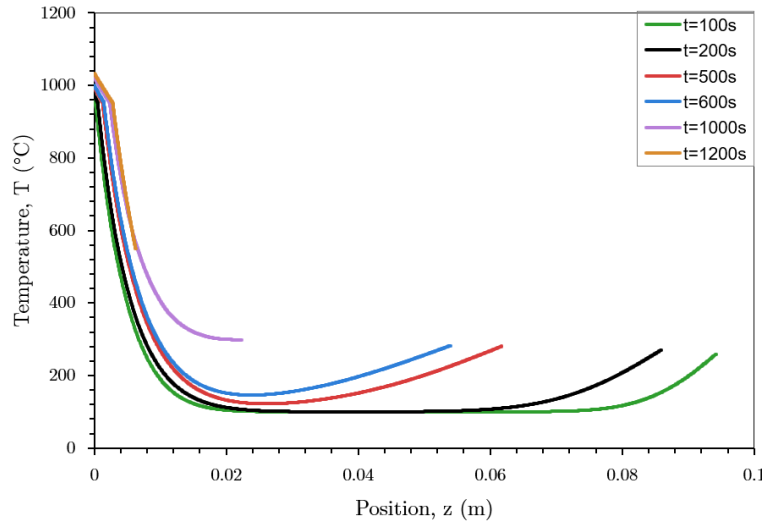


Fig.11 - Temperature profile across the infiltrated domain ds at successive times, from bi-domain activation ($t = 100$ s) to the end of filling ($t = 1200$ s).

Fig.11 shows the time evolution of the temperature profile over the filling window. For each profile, a steep temperature drop from $\approx 950 - 1000$ °C at the ds/cp interface, on the casting powder side, to a minimum of ≈ 100 °C is observed within a few millimeters. It then rises slightly toward the far boundary due to the imposed gas temperature. The ds domain itself shrinks markedly with time, from $z \approx 0.09$ m at $t = 100$ s to $z \approx 0.01$ at $t = 1200$ s: ds grows by melting at the cp side but loses mass by infiltration at the mould wall, and this steady shrinkage shows that infiltration removes mass faster than melting supplies it. The ds/cp -side temperature also rises slightly over time (≈ 980 °C at $t = 1200$ s, against ≈ 950 °C at $t = 100$ s), as the lm/ds boundary approaches a quasi-steady thermal state. This confirms that the infiltrated film stays thin rather than accumulating: the mechanism continuously consumes the ds material as it forms, which is exactly the behaviour needed for it to act as a lubricating skin rather than a growing obstruction. With $M_0 = 59.05$ kg $\cdot$ m $^{-2}$ the initial areal mass, the cumulative infiltrated mass reaches 49.14 kg $\cdot$ m $^{-2}$ (83.2%) at $t = t_F$.

4. CONCLUSIONS AND PERSPECTIVES

- A one-dimensional bi-domain ALE model has been developed for the melting of a fully porous casting powder ($g_0^g = 0.8$) coupled to the infiltration of liquid slag into the mould gap.
- Thermophysical properties are derived from CALPHAD equilibrium calculations with no empirical fitting, the effective conductivity still following the Russell porous-medium model.
- The model reproduces the strong contraction of the powder layer upon melting ($\approx 74\%$).
- Steady-state temperature and flux profiles are validated analytically, with the numerical scheme conserving both mass and energy.
- The melting-interface kinematics show three well-defined regimes governed by the growing thermal resistance of ds .
- Infiltration continuously consumes the ds material as it forms, removing the majority of the initial areal mass by the end of filling and confirming its role as a lubricating skin rather than a growing obstruction.

The following perspectives are identified for future work:

- Sintering: extending the cp domain to solid-state sintering ahead of melting, currently not resolved, to capture densification and property changes before T_S is reached.
- Carbon and gas-phase chemistry in cp : coupling combustion ($C + O_2 \rightarrow CO_2$) and the endothermic Boudouard reaction ($C + CO_2 \rightarrow 2 CO$) to multicomponent gas-phase transport [17]. Increasing the carbon

content is known to slow melting rather than accelerate it, despite the exothermic combustion; this carbon paradox is well documented industrially and experimentally [1, 10, 18], and has been attributed to two mechanisms: carbon coating the oxide grains and inhibiting sintering, and the endothermic Boudouard reaction consuming part of the heat released by combustion.

- Decarbonation: carbonates present in real powders decompose between 600 and 900 °C [19], adding an endothermic sink and a CO₂ source that feeds the Boudouard reaction; this pathway will be included alongside free-carbon combustion.
- Experimental validation: identify kinetic and thermophysical parameters from TGA, DSC and high-temperature XRD, and validate against temperature and off-gas data from furnace and ingot tests.
- Reduced model: a simplified version is intended to serve as a thermal boundary condition at the ingot top surface, for integration into the industrial casting codes.

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