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Towards the Additive Manufacturing Of Zr-based Metallic Glasses Using Liquid Phase Sintering: reactivity and phase transformation kinetics at the crystalline/amorphous interface

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Abstract

Bulk metallic glasses (BMG) present great potential for high-tech applications considering the excellent mechanical and corrosion properties. The on-going development of powder additive manufacturing techniques offers the opportunity to produce rather large parts composed of BMG. Relative to the sinter based additive manufacturing techniques, the sintering of a Zr-based BMG (AMZ4, $Zr_{59,3}Cu_{28,8}Al_{10,4}Nb_{1,5}$) using Zn-based additives was identified as promising: Zn alloys have indeed a melting point ($T_{m,Zn} = 419,5$ °C) lower than the crystallization temperature of the BMG ($T_x \simeq 470$ °C). Here, the study focuses on the understanding of the reactivity between AMZ4 and Zn as a function of temperature and time. A powder blend and a model interface AMZ4/Zn were studied combining *post-mortem* characterizations (SEM, EDX, XRD, micro-hardness. . .) with *in situ* characterization (DSC). Experimental construction of time-temperature-transformation diagram shows that there is a temperature/time window where sintering is permitted without crystallization. Beyond, an intermetallic phase is formed at the interface between the BMG and Zn, this phase was identified in terms of shape, composition and crystallography. A growth mechanism was proposed combining thermodynamic and kinetic aspects.

Keywords: metallic glasses, powder metallurgy, sintering, diffusion, thermal analysis, scanning electron microscopy

1. Introduction

2 Bulk metallic glasses (BMG) present attractive mechanical properties and corrosion resistance due to the
3 absence of structural defects, such as dislocations and grain boundaries, usually found in crystalline materials.
4 Since their recent discovery [1], more and more studies focus on their characterization, their synthesis or their

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5 shaping due to their potential applications in many advanced fields (medical, aeronautics, . . .). Nevertheless,
6 there is a significant limitation on dimensions of BMG materials considering the manufacturing methods
7 that can reach the critical cooling rate required to retain their amorphous structure.

8 In that framework, additive manufacturing (AM) techniques applied to BMG attracted much attention in
9 the recent years. It would enable to overcome this dimensional limitation and to build complex shapes with
10 minimal feedstock, *e.g.* powder, without inducing the crystallization of BMG. The production of microscopic
11 metallic glass powder by gas atomization has been repeatedly reported in the literature with emphasis on
12 morphology, size distribution, oxygen content and thermal stability. [2–4]. Due to their high glass-forming
13 ability, numerous Zr-based metallic glass powders are even commercially available such as the industrial-
14 grade alloy AMZ4 ($\text{Zr}_{59.3}\text{Cu}_{28.8}\text{Al}_{10.4}\text{Nb}_{1.5}$) with a particle size distribution suitable for AM techniques.
15 To this day, it is also the most used and studied Zr-based BMG relative to AM [5]. So far, most studies
16 focused on laser-based AM since the cooling rate is theoretically high enough to bypass crystallization and,
17 therefore, to process BMG [6–10]. However, depending on the glass-forming ability of the initial powder
18 and on the processing parameters, partial crystallization of BMG happens in the lower layers known as heat
19 affected zones (HAZ) [10]. On the contrary, very few studies were devoted to the sinter-based AM (SBAM)
20 techniques (direct ink writing, binder jetting, fused deposition modeling, metal injection moulding, etc.)
21 [5]. Even if densification of metallic glass powders was reported using mostly spark plasma sintering (SPS)
22 [11–14], it is still challenging to sinter BMG particles without mechanical pressure in order to retain the 3D
23 printed shape.

24 Aiming at SBAM techniques, the liquid phase sintering of a metallic glass powder using metal-based
25 additives was identified as promising. For example, Zn alloys could be used as biocompatible additives due
26 to their melting temperature ($T_{m,Zn} = 419,5 \text{ }^\circ\text{C}$) lower than the crystallization temperature of many BMG
27 such as some Zr-based BMG ($T_x \simeq 470 \text{ }^\circ\text{C}$ for the industrial grade Zr-based metallic glass powder called
28 AMZ4). Zn alloys would permit the densification of metallic glass powder compact via liquid phase sintering.
29 For now, many challenges remain in understanding the interactions between the BMG and crystalline metals,
30 whether solid or liquid.

31 In the literature, such interactions have been reported by few diffusion bonding studies [15–21]. In-
32 terestingly, Zhang et al. studied the joining process between a Zr-based BMG ($\text{Zr}_{62}\text{Cu}_{33}\text{Al}_4\text{Ti}$) and a Zn
33 alloy (3 wt% Al) around 415°C , *i.e.* above both the glass transition temperature of the BMG ($T_g=410^\circ\text{C}$)
34 and the liquidus temperature of the Zn alloy ($T_L=397^\circ\text{C}$) [15, 16]. Using an ultrasound-assisted brazing
35 method, joints with satisfactory mechanical properties (shear strength) were obtained and the growth of
36 a reactive layer up to $17 \mu\text{m}$ was discussed relative to cavitation effects and Al and Zn diffusion into the
37 Zr-based BMG. Even if most studies accomplished diffusion bonding in the supercooled liquid region of
38 the BMG (*i.e.* between T_g and T_x) [15–18], Saadati *et. al.* elaborated a micron-sized dissimilar joint,
39 between a Cu-based BMG ($\text{Cu}_{50}\text{Zr}_{43}\text{Al}_7$) and pure aluminium, $10 \text{ }^\circ\text{C}$ below the T_g of the BMG in order

40 to preserve amorphous structure and initial geometry of the BMG [20, 21]. Saadati *et. al.* emphasized the
41 role of oxygen in the formation of the diffusion layer by proposing a bonding mechanism which relies on
42 the dissolution and precipitation of surface oxide layers induced by elemental inter-diffusion at the interface.
43 In the current study, the bonding mechanism will be discussed regarding both the oxide layer and possible
44 diffusion paths in BMG. Some other studies commented the reactivity between BMG and crystalline met-
45 als by analyzing the as-formed interphase between BMG and crystalline metals in BMG/metal composites
46 [22–31]. For example, aiming at mechanical properties, interpenetration phase composites were elaborated
47 by casting a Zr based glass forming liquid into 3D stainless steel [22] or SiC ceramic foam [23]. The increase
48 of the tensile plasticity [22] or fracture strength [23] was attributed, in both cases, to the formation of a nm-
49 to μm -thick interphase. Nevertheless, these studies only focused on the analysis of the as-formed interphase
50 without considering its growth mechanism. More commonly, BMG/metal composites were manufactured by
51 consolidation of mixed powder using SPS [25–29]. Dudina *et. al.* showed that, in the case of a composite
52 composed of a Al matrix and Fe-based BMG particles reinforcements, the growth of an interfacial layer
53 highly depends on the temperature : no micrometer-sized interphase was detected for composites sintered
54 at 500°C for 3 min whereas 10-15 μm thick interphase was characterized for composites sintered at 570°C
55 for 3 min. Unfortunately, the other reported studies investigated the sintering of BMG/metals composite
56 without varying the duration and the temperature of the heat treatment. These studies chose to focus on
57 other parameters such as the volume fraction [26, 29] or the powder size of BMG [28]. In that framework, it
58 was considered extremely appropriate to carry out a study on the reactivity between a BMG and crystalline
59 metal considering the effect of several key parameters such as time and temperature of the heat treatment.

60 The present work focuses on the characterization of the reactivity between a Zr-based metallic glass
61 powder and Zn additive metal around the T_g of the BMG and around the T_m of Zn. The goal of this study
62 is twofold : (i) investigate the possibility of liquid sintering or brazing a Zr-based BMG using Zn metal and
63 (ii) describe the reactivity between these two materials. For the first time, the system Zr-based BMG /
64 Zn is investigated through two forms: powders blends and interdiffusion couples. The reactivity between
65 Zr-based BMG and Zn will be described in details and compared to the crystallization kinetics of the sole
66 BMG.

67 2. Materials and methods

68 2.1. Materials

69 The Zr-based BMG used in this study is a commercial gas-atomized powder with a nominal composition
70 (at%) of $\text{Zr}_{59.3}\text{Cu}_{28.8}\text{Al}_{10.4}\text{Nb}_{1.5}$ (former trade name AMZ4, Heraeus GmbH, Germany). The glass transition
71 temperature (T_g) and the crystallization onset temperature (T_x) of this BMG are 415.8 °C and 474.1 °C
72 respectively at a heating rate of 30 K/min using a DSC Sensys Evo (SETARAM®) as described hereafter

73 (see 2.2). The corresponding heat flow curve of AMZ4 powder is reported in Supplementary Materials (see
74 Figure S1) alongside with the heat flow curve of the Zn powder used (see Figure S2). The AMZ4 powder
75 (particle size distribution : $d_{v10} = 11.4 \mu\text{m}$, $d_{v50} = 20.2 \mu\text{m}$, $d_{v90} = 36.7 \mu\text{m}$) was mixed with the Zn
76 powder (purity > 99,5%, $T_{m,Zn,measured} = 417.6 \text{ }^\circ\text{C}$, $d_{v10} = 5.5 \mu\text{m}$, $d_{v50} = 12.2 \mu\text{m}$, $d_{v90} = 28.6 \mu\text{m}$) using
77 a 3D shaker mixer (Turbula®) for 30 min to obtain a powder blend composed of 70 vol% of AMZ4 and
78 30 vol% of Zn (AMZ4/Zn, 70/30v). For AMZ4 and Zn powders, the oxygen contents were measured using
79 an EMGA 620 W analyser from Horiba Jobin Yvon company and were equal to 782 ppm and 3555 ppm by
80 weight ($\mu\text{g/g}$), respectively.

81 2.2. Differential scanning calorimetry

82 In order to investigate *in situ* the isothermal kinetics either of the crystallization of AMZ4 or of the
83 reaction between AMZ4 and Zn, Differential Scanning Calorimetry (DSC) measurements were performed
84 using a DSC Sensys Evo (SETARAM®) equipped with 3D sensor under Ar flow (10 mL/min, Air Liquide
85 Alphagaz 2) for temperatures ranging from 325 °C to 450 °C. For each measurement, around 80 mg of powder
86 were poured into an opened alumina crucible, kept under flowing Ar for 30 min at 25 °C, then heated from
87 ambient temperature at 30 K/min until the isothermal temperature is reached. The latter was maintained
88 long enough for the different events to occur. For these experiments, the AMZ4 powder and the blend of
89 powders AMZ4/Zn (70/30v) were used. The data were analyzed using SETARAM software (CALISTO)
90 and Python (libraries : lmfit, numpy and pandas). For each heat flow curve, the baseline was fitted using a
91 polynomial function of order 2 before being subtracted to the signal. The following treatments are detailed
92 later in the results section (see 3.1).

93 2.3. Heat treatments in salt bath to simulate isothermal sintering

94 As the control of the temperature is critical, heat treatments of mixed powders were carried out using
95 salt bath. AMZ4/Zn (70/30v) powder blends were encapsulated in quartz tube (inner diameter 3 mm, outer
96 diameter 5 mm) under primary vacuum ($<5.10^{-2}$ mbar after 3 argon purges). The quartz tubes were then
97 immersed in salt bath for different times (1 - 30 min) at different temperatures (350 – 425 °C) before being
98 quenched into a water bath. Finally, the powders were recovered from the tubes to be analysed by X-
99 Ray diffraction (XRD, Bruker AXS D8 Advance Diffractometer, Cu $K\alpha$ radiation, 19-90°) and by scanning
100 electron microscopy (SEM, Tescan Vega3) equipped with an energy-dispersive X-ray spectroscopy (EDX)
101 detector.

102 2.4. Study of Plane Interfaces using Diffusion bonding experiments

103 In order to study the interface between AMZ4 and Zn, diffusion bonding experiments were carried out.
104 AMZ4 and Zn powders were firstly independently sintered using SPS under vacuum with a uniaxial pressure

105 of 64 MPa and with a heating rate of 100 K/min: at 420 °C for 2 min for AMZ4 and 370 °C for 20 min
 106 for Zn. The relative density of sintered AMZ4 and Zn pieces were measured equal to $98.1 \pm 0.5 \%$ to 98.4
 107 $\pm 0.3 \%$ respectively using Archimedes' principle. The surfaces were then polished down to #4000 SiC
 108 paper and ultrasonically cleaned in an acetone bath for 3 min. Then, the two plates of 5 mm square and
 109 3 mm thick were brought into contact and wrapped carefully with graphite sheets to form the AMZ4/Zn
 110 solid state interdiffusion couple. Then, the interdiffusion couple was placed in the graphite die and heat
 111 treated at 375 °C and 400 °C for 5, 20 and 60 min in SPS (64 MPa, vacuum, heating rate 100 K/min). The
 112 samples were then embedded in epoxy resin, mirror polished (diamond suspension, 1 μm) and analysed by
 113 SEM-EDX. Following the SEM analysis, micro-hardness measurements (10 gf, 10 s) were also performed on
 114 the embedded and polished samples along the studied interfaces using Micromet 5104 (Buehler®). Each
 115 reported Vickers hardness value is the average of at least five consecutive measurements.

116 For a better understanding, the different complementary experimental approaches are summarized on
 117 the Figure 1.

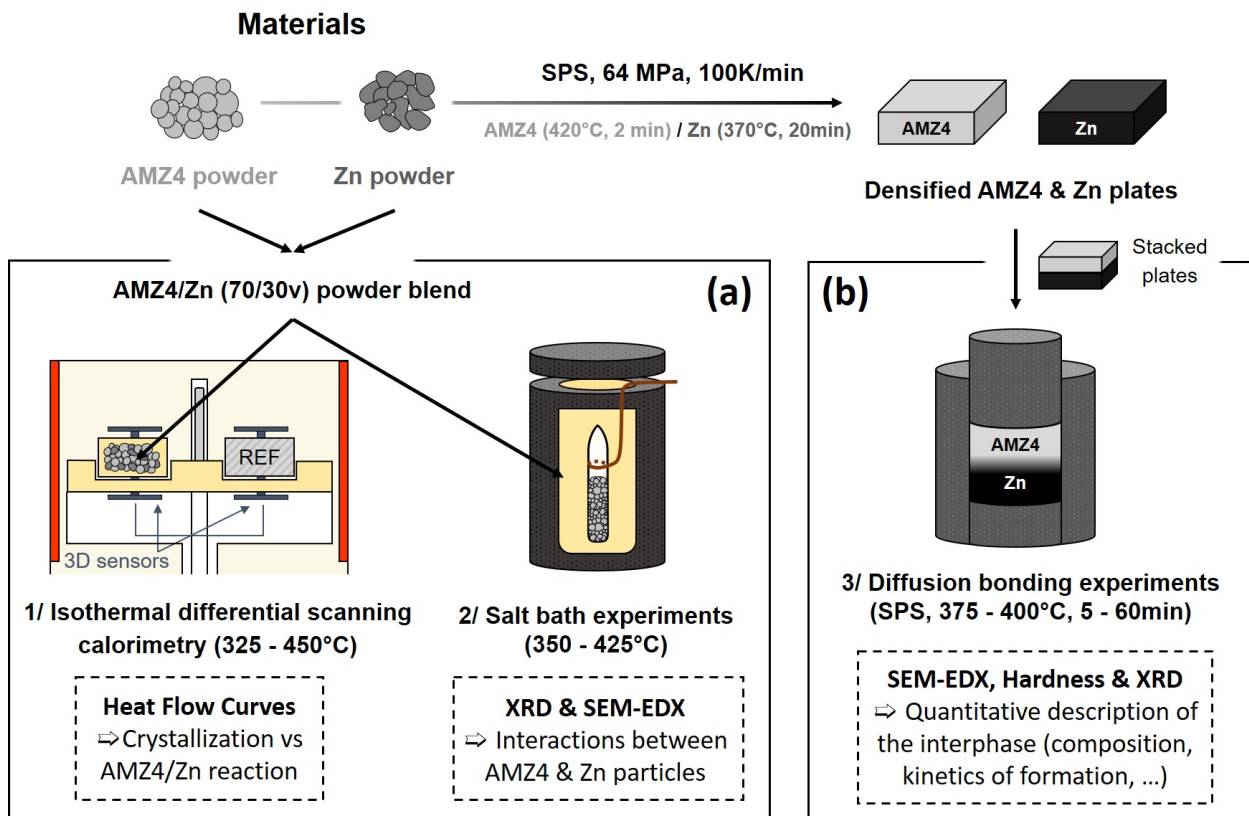


Figure 1: Representation of the experimental procedure for the study of the reactivity between AMZ4 and Zn powders : (a) experiments carried out on AMZ4/Zn (70/30v) powder blends and (b) diffusion bonding experiments of AMZ4/Zn interdiffusion couples

118 3. Results

119 3.1. Isothermal crystallization kinetics of AMZ4 and reaction kinetics of AMZ4/Zn system using DSC

120 Previous researchs investigated the isothermal crystallization kinetics of the industrial grade AMZ4 above
121 its T_g using DSC measurements [8, 32]. Here, the isothermal crystallization kinetics of AMZ4 and reaction
122 kinetics of AMZ4/Zn system are also studied below the T_g of AMZ4. Each heat flow curve (see Figure 2
123 a,c) displays a distinct exothermic peak whose temporality and shape change according to the temperature.
124 In the case of AMZ4 powder, these peaks reflect the intrinsic crystallization of AMZ4 whereas, in the case
125 of the AMZ4/Zn system, they highlight a reaction between AMZ4 and Zn. The use of the 3D sensor enables
126 the detection of low intensity peaks for the AMZ4 powder after 200 min of holding time at 400°C, *i.e.*
127 below the T_g of AMZ4, and for the AMZ4/Zn system after around 300 min of holding time at 325°C. For
128 each sample, the first and second derivatives of the heat flow with regard to the time were calculated to
129 identify the beginning (t_0) and the end (t_{endset}) of the exothermic peaks. After subtraction of the baseline,
130 the progress of the reaction, X , as a function of time for a given isothermal temperature can be calculated
131 using the heat flow associated with the event (dH/dt) via the equation 1 [32]. The Figure 2 (b,d) shows the
132 computed evolutions of the reaction as a function of time ($t=0$ at the beginning of the holding step).

$$X(t)_T = \frac{\int_{t_0}^t (\frac{dH}{dt}) dt}{\int_{t_0}^{t_{endset}} (\frac{dH}{dt}) dt} \quad (1)$$

133 Firstly, concerning the kinetics of the intrinsic crystallization of AMZ4 and of the reaction between
134 AMZ4 and Zn, the exothermic peaks were detected later and were more extended in time (logarithmic
135 scale) when the temperature decreased. For example, the intrinsic crystallization of AMZ4 occurred in less
136 than 10 min at 450°C while it took more than 500 min at 400°C. Comparing the two systems, it is clear
137 that the kinetics of the intrinsic crystallization and of AMZ4/Zn reaction are highly different: at 400 °C, the
138 crystallization of AMZ4 began after 3 hours, whereas an intense exothermic peak was detected earlier than
139 10 min for the AMZ4/Zn powder blend. Moreover, the shape of the exothermic peaks are highly different
140 which suggests that the intrinsic crystallization and the reaction between AMZ4 and Zn took place through
141 very different mechanisms. The crystallization mechanism of BMG is complex and likely occurs in several
142 steps starting with the growth of crystalline spherulites and ending with different crystalline phases [33–35].
143 The crystallization mechanism of AMZ4 will not be discussed hereafter as a dedicated study would be needed
144 involving atomic-scale investigations (short-range order, nucleation of the nano crystallites, growth, etc) as
145 shown by Gautier *et. al.* for a Ti-based BMG [33]. The mechanism linked to the AMZ4/Zn reaction will
146 be discussed with direct observation later on.

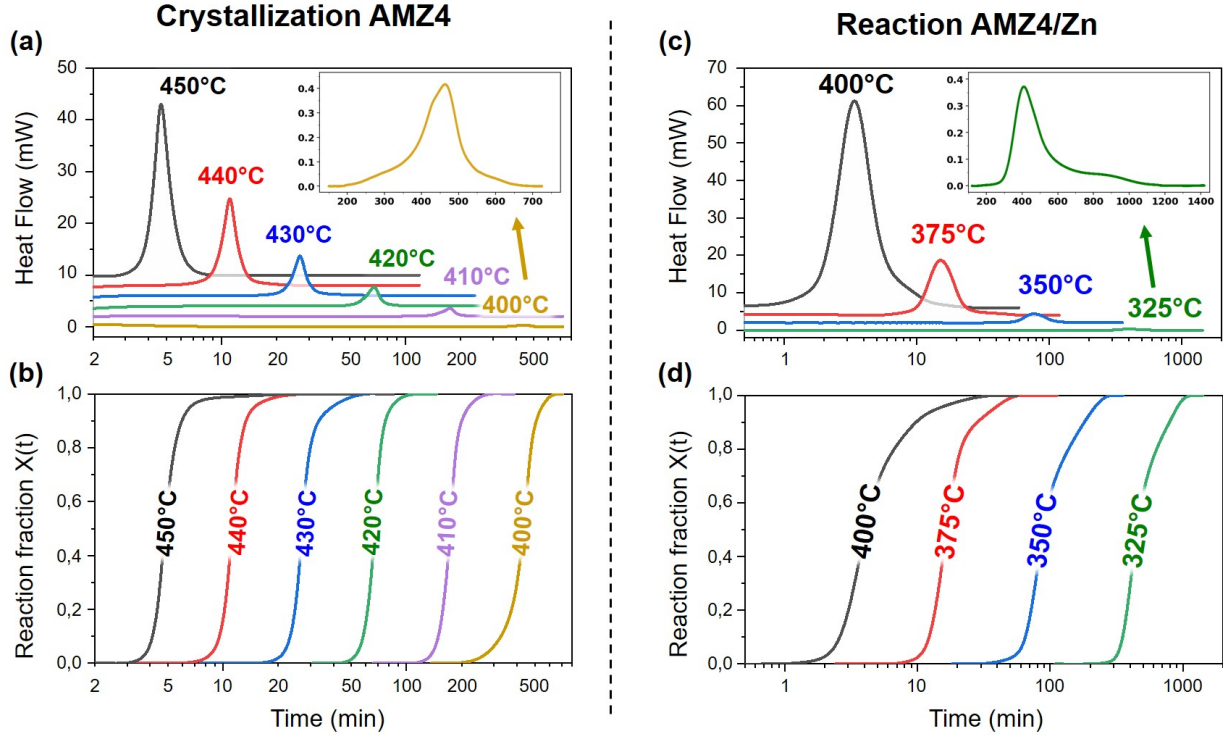


Figure 2: DSC results for AMZ4 (a,b) and AMZ4/Zn (70/30v) (c,d) samples: (a,c) Heat flow curves (325-450°C) measured by DSC, (b,d) reaction fraction as a function of time (log scale)

147 Based on these results, time-temperature-transformation (TTT) diagrams were constructed and are
 148 presented in Figure 3 (a). Similar TTT diagrams have already been determined upon heating and cooling
 149 for AMZ4 [8, 32]. The apparent activation energy of crystallization or of reaction (E_a) was also computed
 150 via an Arrhenius-like equation (Equation 2 [32]) as a function of the annealing time required to reach a
 151 certain reaction progress (t_X):

$$t_X = A * \exp\left(\frac{E_a}{RT}\right) \quad (2)$$

152 with A a constant pre-exponential factor, T the temperature (in K) and R the universal gas constant.
 153 These plots show that the AMZ4/Zn reaction will always take place long before the intrinsic crystallization
 154 of AMZ4 (see Figure 3 (a)). This can be commented by analyzing the difference in activation energy of the
 155 reactions in the two systems (see Figure 3 (b)) : the activation energy is approx. 150 kJ/mol lower for the
 156 AMZ4/Zn reaction. If the values with the least uncertainty, *i.e.* between 30 and 60% of reaction progress,
 157 are averaged, the activation energy of crystallization of industrial grade AMZ4 powder and of the reaction
 158 between AMZ4 and Zn equals 365.5 ± 0.4 and 212.6 ± 1.3 kJ/mol respectively. The slight variations of the
 159 activation energy as a function of the reaction progress suggest that there is an evolution of the step that

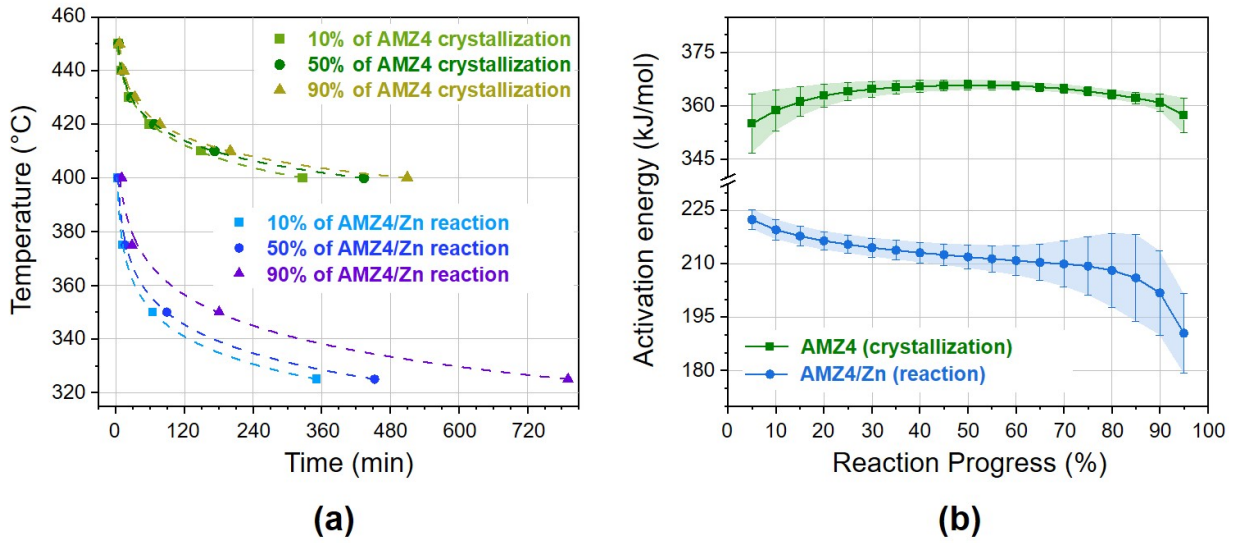


Figure 3: (a) Isothermal time-temperature-transformation (TTT) diagrams and (b) Activation energy as a function of reaction progress for both AMZ4 and AMZ4/Zn systems

160 controls the reaction over time. For example, different reactions can take place and the global reaction could
 161 start with an interfacial chemical reaction and continue with bulk diffusion. Once again, this will be more
 162 discussed later on (see 4.2). Concerning the liquid phase sintering of the system AMZ4/Zn, the kinetics
 163 of the reaction between Zn and AMZ4 powders seems too high below the melting point temperature of Zn
 164 ($T_{m,Zn} = 419,5 \text{ } ^\circ\text{C}$) to enable the liquid phase sintering of AMZ4. Even if it would be possible to avoid
 165 the crystallization of AMZ4 with a 20 min isothermal treatment at $425 \text{ } ^\circ\text{C}$, Zn particles would react with
 166 AMZ4 particles before melting. However, based on the TTT diagrams, heat treatments can be designed to
 167 control this reactivity, which could allow the consolidation of the AMZ4/Zn powder compact by solid state
 168 diffusion bonding.

169 3.2. Study of the AMZ4/Zn reaction using *ex situ* analysis

170 Post mortem analysis by XRD and SEM-EDX of powder blends immersed in salt bath for different times
 171 (1-30mn) at different temperature ($350\text{-}425^\circ\text{C}$) enable to further investigate the reaction between AMZ4 and
 172 Zn particles. Experimentally, it was observed that, after all heat treatments, the powder compacts were
 173 still friable and were crushed using a spatula. This experimental observation indicated that the powder
 174 compacts did not densify enough to have a sufficient mechanical cohesion. Further characterizations were
 175 performed at a microscopic scale.

176 The Figure 4 shows XRD diffractograms for powder blends immersed in $400 \text{ } ^\circ\text{C}$ salt bath for 5, 10 and
 177 20 min and also a fully crystallized AMZ4 sample (reference). Initially, the powder blend is composed of Zn
 178 particles with defined diffraction peaks and amorphous AMZ4 particles which display a broad diffraction

179 peak between $2\Theta=30^\circ$ and $2\Theta=50^\circ$ characteristic of the amorphous phase. During the heat treatment at
180 400°C , new crystalline peaks form at the expense of Zn peaks until no distinct Zn peaks can be detected
181 after 20 min of heat treatment at 400°C . These new peaks are only the result of the reaction between AMZ4
182 and Zn. Indeed, the new peaks do not correspond to the peaks that can be detected in a crystallized AMZ4
183 sample and the typical AMZ4 amorphous bump is still detected. Moreover, it can be considered that, for
184 AMZ4/Zn (70/30v) powder blend, Zn is the limiting reactant and only a part of the amorphous volume of
185 AMZ4 particles reacted with Zn particles. The newly formed crystalline phase(s) could not be identified
186 using the common reference databases so that they will be referred as Reactive Affected Zone (RAZ) until
187 its identification later in this article.

188 In order to estimate the respective evolution of the different phases, the integrated intensities of the
189 peaks were used to compute the area fraction of each phase (AMZ4, Zn and RAZ phases) :

$$\text{Area fraction of phase } i = \frac{\sum \text{Integrated intensity of phase } i}{\sum \text{Integrated intensity of all peaks}} \quad (3)$$

190 The integrated intensity of the peaks has been determined by fitting each peak with pseudo-Voigt func-
191 tion using Lmfit Python Library [36]. As a function of time and temperature, 2D color maps with linear
192 interpolation between experimental points were computed (see Figure 4 (b)). In comparison to the Figure
193 2, the kinetics determined by XRD is in accordance with the DSC curves and the TTT diagram, even if the
194 experimental conditions are slightly different. For example, there is no heating ramp as the quartz tubes
195 are directly immersed in the salt bath at the set temperature so that a slight incubation time must be taken
196 into account. At 425°C , the formation of RAZ phase(s) is very fast : peaks relative to Zn phase are no
197 longer detected after 5 min.

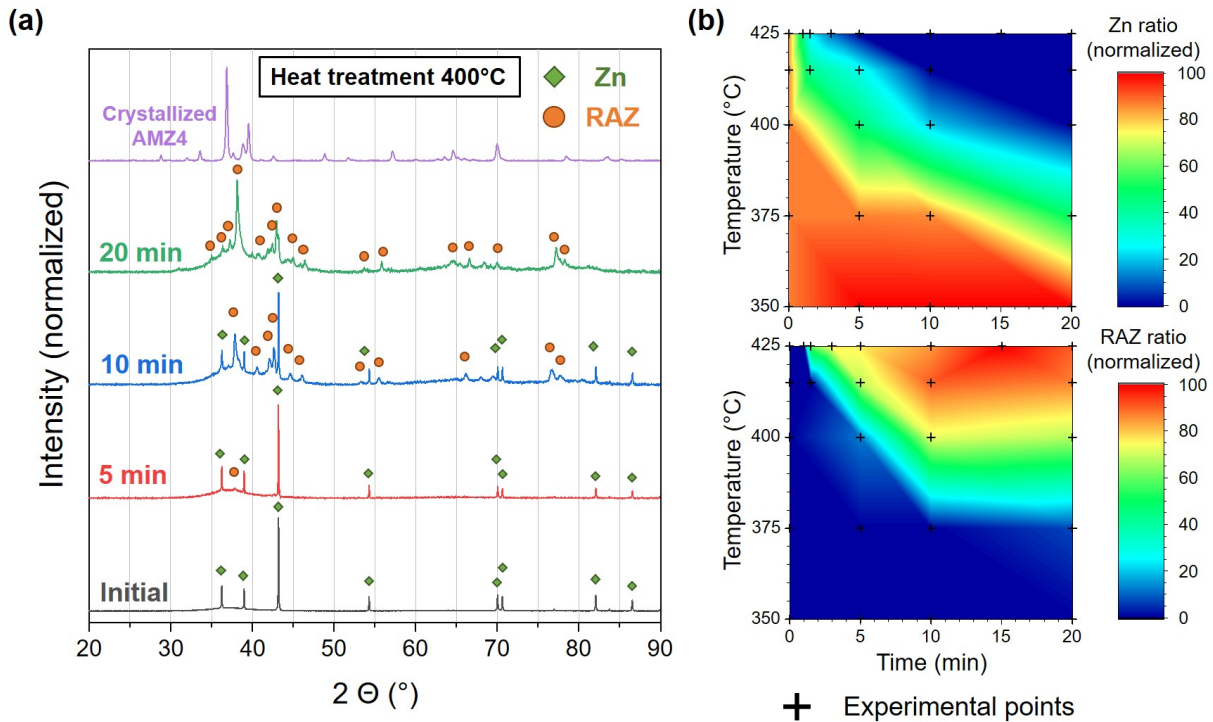


Figure 4: XRD diffractograms of salt bath experiments showing (a) diffractograms for heat treatment at 400°C and (b) 2D colour maps of the evolution of the fraction of Zn and RAZ phase(s) computed from integrated intensity of peaks as a function of time (0-20min) and temperature (350-425°C)

198 SEM-EDX analysis (see Figure 5) revealed that Zn particles reacted with AMZ4 particles by forming
 199 cones at the surface of AMZ4 particles rather than forming bridges between. This may explain the friability
 200 of the samples recovered from the tubes. Depending on the respective size of AMZ4 and Zn particles, the
 201 number and size of the cones varied from one particle to another during the heat treatment: for example,
 202 the smallest Zn particles reacted first with AMZ4 particles. SEM-EDX analysis of all heat treated samples
 203 suggests that the cones were likely formed by the solid state spreading of Zn on the surface of AMZ4 particles
 204 followed by interdiffusion and reaction between AMZ4 and Zn parts. Indeed, elemental distributions of the
 205 main elements showed that interdiffusion of the elements happened along the cones to the AMZ4 particles.
 206 Interestingly, the basis of the cones seems to be less concentrated in Cu than the top of the cone whereas
 207 the distribution of other elements is gradual from the basis to the top of the cone. Therefore, these different
 208 zones may reflect the presence of new crystalline phase(s) analyzed by XRD. The particular interdiffusion
 209 phenomena occurring during the reaction between Zn and AMZ4 particles must be studied in more detail
 210 using a model interface.

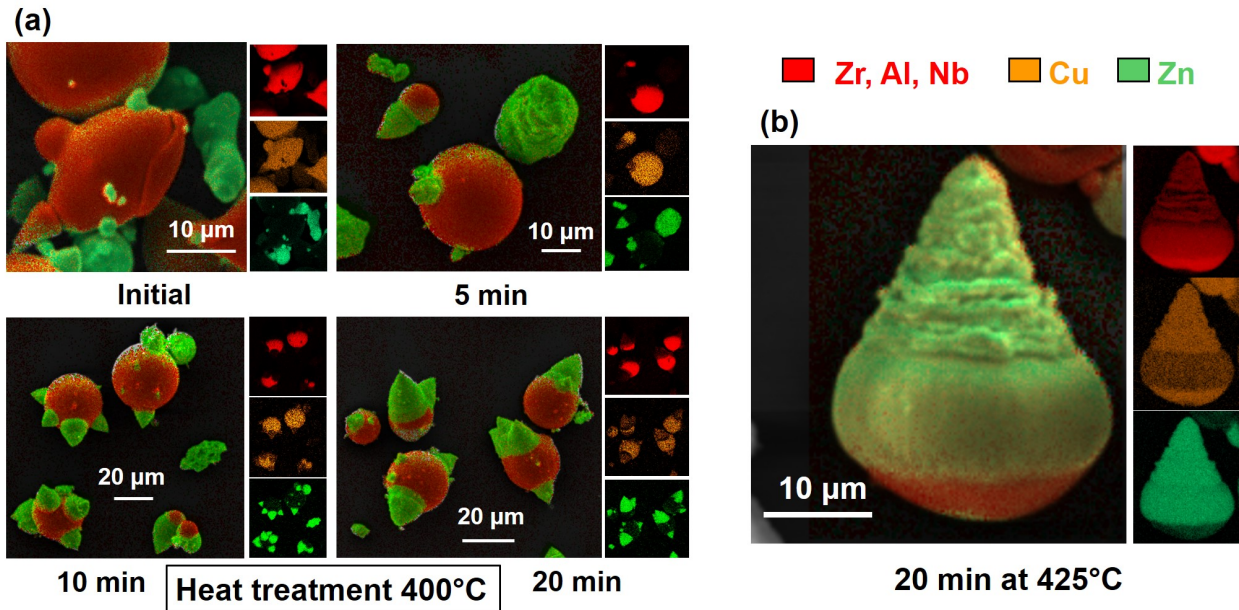


Figure 5: SEM-EDX analysis of salt bath experiments for AMZ4/Zn (70/30v) powder blends (a) for 0, 5, 10 and 20 min at 400°C and (b) 20 min at 425°C

211 3.3. Microstructural evolution of AMZ4/Zn interface

212 In order to facilitate the investigation of the reactivity between these two species, AMZ4/Zn model
 213 planar interfaces were synthesized and studied. In this way, the influence of the powder size distribution was
 214 not considered and quantitative EDX analysis were correctly performed. The Figure 6 presents the 110 μm
 215 thick interphase, the Zn/RAZ interface and the AMZ4/RAZ interface of the AMZ4/Zn interdiffusion couple
 216 heat treated at 400 °C for 60 min. One can notice that there are no unbounded regions, that the Zn/RAZ
 217 interface is much flatter than the AMZ4/RAZ interface and that micropores are present in the interphase
 218 close to the AMZ4/RAZ interface. Moreover, the chemical contrast suggests that there are two distinct
 219 zones composing the interphase close to the AMZ4/RAZ interface. The Figure 7 shows, for the 400 °C
 220 treated interdiffusion couples, the evolutions of the elemental distribution, measured via EDX line profiles,
 221 alongside the Vickers microhardness evolutions. Table 1 reports the measured lengths of the interphases for
 222 all interdiffusion couples using EDX analysis. The length of the interphase was measured using EDX line
 223 profiles (at least five line profiles per sample) and was defined as the length for which the atomic fraction
 224 of Zn goes from 100 at.% to 0 at.% and also for which the atomic fraction of Zr (resp. Cu) goes from 59.3
 225 at.% (resp. 28.8%) to 0 at.%. As the AMZ4/RAZ interfaces are not perfectly flat, the uncertainty linked to
 226 the length of the interphase was increased to 2 μm .

227 In accordance with previous results, the growth of the interphase increases with time and, primarily,
 228 with temperature. At 400°C, the growth accelerates so drastically that the length of the interphase becomes

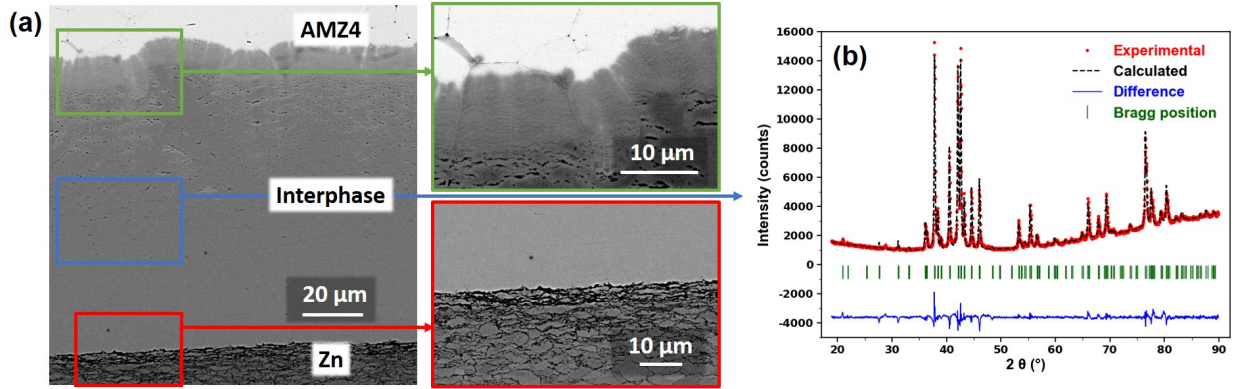


Figure 6: Analysis of AMZ4/Zn interdiffusion couple heat treated at 400°C for 60 min with (a) SEM analysis (Backscattered Electrons) of the interfacial region and (b) XRD analysis and Le Bail refinement of the main interphase zone

Duration of heat treatment	375°C	400°C
5 min	Not bonded	$18 \pm 2 \mu\text{m}$
20 min	$13 \pm 2 \mu\text{m}$	$83 \pm 2 \mu\text{m}$
60 min	$45 \pm 2 \mu\text{m}$	$110 \pm 2 \mu\text{m}$

Table 1: Length of the interphases for AMZ4/Zn interdiffusion couples measured by EDX

229 in approximately 5 min almost greater than the particle size of Zn and AMZ4 powders. On the contrary, the
 230 interdiffusion couple heat treated at 375°C for 5 min was not sufficiently bonded to carry out the analysis.
 231 For 20 min (resp. 60 min) of heat treatment, the length of the interphase is almost seven times (resp.
 232 three times) higher at 400°C than at 375°C. Even if the diffusion mechanism will be discussed later in this
 233 article (see 4.2), it can be highlighted that the growth of the interphase accelerates when the temperature
 234 approaches the supercooled liquid region of the BMG and the melting point temperature of Zn.

235 By analyzing the chemical compositions and mechanical properties of the interphase (see Figure 7), it
 236 is clear that the interphase is composed of two main distinct zones. First, there is a main zone which has
 237 a well defined composition (rich in Zn but with a small quantity of Zr, Cu, Al and O) and characterized
 238 by an intermediate hardness ($270.4 \pm 17.7 \text{ HV}$) between that of Zn ($49.0 \pm 1.1 \text{ HV}$) and that of AMZ4
 239 ($462.0 \pm 31.8 \text{ HV}$). Its composition by EDX is: $91.5 \pm 0.6 \text{ at.}\%$ Zn, $5.2 \pm 0.5 \text{ at.}\%$ Zr and $3.3 \pm 0.4 \text{ at.}\%$
 240 Cu. A diffuse layer between this first zone and the Zn part can be noted where there is a slight segregation
 241 of Al and O atoms. This segregation can be explained taken into account the high stability of Al_2O_3 : Al
 242 atoms of AMZ4 would associate with O atoms initially present in the oxide layers of both AMZ4 and Zn.
 243 The location of the O atoms could be considered as a marker of the initial interface between AMZ4 and Zn.
 244 This would indicate that the interphase developed primarily by diffusion of Zn atoms into AMZ4 part. The
 245 length of this zone is systematically around $2 \mu\text{m}$ thick which is close to the uncertainties linked to EDX

246 measurements, that's why the composition of this narrow zone will not be discussed hereafter. The second
247 zone is a wide transition zone where the quantity of Zr, Cu and Al (resp. Zn) increases (resp. decreases)
248 drastically and where the measured hardness (around 600 HV) is higher than the hardness of the BMG.
249 This high hardness value is due to the formation of some intermetallic compounds by elemental diffusion
250 in this zone. In the case of the sample heat treated at 400°C for 5 minutes, the length of the second zone
251 was too small compared to the size of the indentation to differentiate it from the first zone using the micro
252 hardness measurements. This second zone may be composed of several phases and a whole study using
253 Transmission electron microscopy (TEM) would be necessary. For this first study, taking into account the
254 characterization techniques available, it has been chosen to focus on the evolution of the length of the two
255 zones and on the characterization of the main first zone. Interestingly, the length of the main zone represents
256 always around 70% of the total length of the interphase. Therefore, both zones grow together for the heat
257 treatments studied : the first zone grows with a constant composition and the second zone also grows with
258 a concentration gradient decreasing with time.

259 Finally, a last effort was made to try to determine the phase constituting the main part of the interphase.
260 After a proper grinding of the sample heat treated at 400 °C for 60 min, this main zone was analysed by
261 XRD (see Figure 6 (b)). Le Bail refinement determined that the crystal structure of the major part of the
262 interphase can be mostly described with the space group Fd-3m with a mesh parameter of $14.0777 \cdot 10^{-10}$
263 m. Exploring crystal structure database computed using Density Functional Theory (DFT) and knowing
264 its composition from EDX point measurements, it appears that the main zone shares many similarities with
265 the DFT-calculated $\text{Zr}(\text{Zn}_{10}\text{Cu})_2$ phase which exhibits a Fd-3m structure with a mesh parameter of 14.04
266 10^{-10} m (data retrieved from the Materials Project for $\text{Zr}(\text{Zn}_{10}\text{Cu})_2$ (mp-1195710) from database version
267 v2021.11.10). Certainly the chemical composition does not correspond exactly but it is close enough to
268 suggest that the composition of the phase detected is indeed very close to this one.

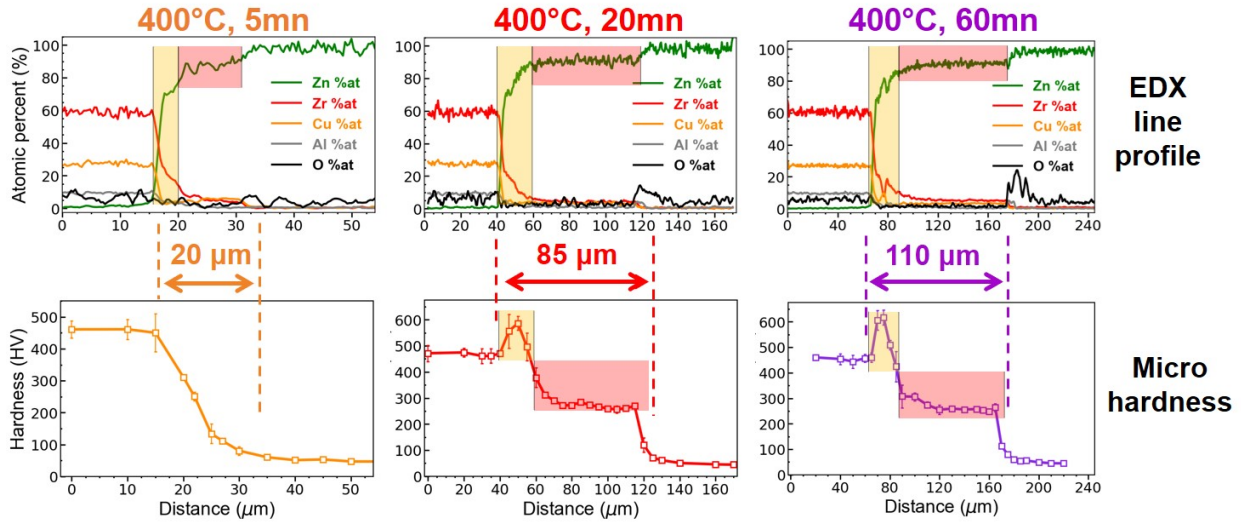


Figure 7: Evolution of elemental distribution (EDX) and hardness (micro hardness) along AMZ4/Zn interdiffusion couples heat treated at 400°C : the two main zones of the interphase are highlighted using yellow and red boxes

269 4. Discussion

270 4.1. On the reactivity for the powder mixtures relative to liquid sintering

271 The reactivity between AMZ4 and Zn was well established following three different experimental pro-
 272 tocols. Even with slightly different experimental conditions, the salt bath experiments and the diffusion
 273 bonding experiments validated the TTT diagrams constructed from *in situ* DSC experiments. This diagram
 274 allows to understand which heat treatments AMZ4 can withstand without crystallizing and which heat
 275 treatments would initiate a reaction between AMZ4 and Zn. Even if there is a temperature/time window
 276 above $T_{m,Zn}$ where the sintering of AMZ4 would be permitted without intrinsic crystallization of the BMG,
 277 the reactivity is too high between AMZ4 and Zn below $T_{m,Zn}$ to allow the liquid phase sintering of AMZ4
 278 (no grain rearrangement possible [37]). Moreover, the liquid phase sintering is all the less likely to occur as
 279 it seems that, during the reaction, the diffusion goes from Zn to AMZ4 particles. Indeed, for liquid phase
 280 sintering, it is commonly accepted that the solid phase must present high solubility in the additive (or liquid)
 281 phase but little reverse solubility (*i.e.* low solubility of the additive phase in the solid phase) [37]. In the
 282 case of the AMZ4/Zn (solid phase/additive phase) system, several experimental facts tend to say that the
 283 solubility of two species is the inverse of the ideal case. For AMZ4/Zn (70/30v) powder blends, Zn particles
 284 form cones at the surface of AMZ4 particles instead of bridges between particles. Then, for AMZ4/Zn
 285 interdiffusion couples, Zn/RAZ interface is perfectly flat whereas the RAZ/AMZ4 interface is wavy and it
 286 seems that the diffusion front progresses through a AMZ4 particle (see the green box in Figure 6). Moreover,
 287 as discussed earlier (see 3.3), the O atoms may be considered as markers of the initial oxide layers. The

288 location of these markers at the Zn/RAZ interface also indicates that Zn atoms mainly diffuse into AMZ4
289 and not the other way around. Subsequent studies could confirm this by introducing other markers at the
290 interface before the heat treatment or by performing molecular dynamics simulations. Thereupon, it can
291 be highlighted that a recent study, using molecular dynamics simulations, proves that the interdiffusion
292 between pure Cu and amorphous Cu₅₀Zr₅₀ BMG is mainly attributed to the diffusion of Cu atoms into the
293 amorphous structure [38]. More recently, Saadati *et. al.* also stated that, for a Cu-BMG/Al system, the
294 growth of the reaction layer was mainly towards the BMG side rather than the pure metal side [21]. This
295 could be explained regarding the high configurational entropy in the amorphous structure which prevents
296 diffusion of elements outside the BMG. In any case, relative to liquid phase sintering, it is clear that the
297 system AMZ4/Zn is not a proper candidate.

298 4.2. Thermodynamic aspects and diffusion mechanisms for interphase growth

299 Through *in situ* DSC experiments, the activation energy of the reaction between AMZ4 and Zn was
300 estimated as equal to 212.6 ± 1.3 kJ/mol which was lower than the activation energy of the intrinsic
301 crystallization of AMZ4 (365.5 ± 0.4 kJ/mol). Nevertheless, further comments can be made about the
302 thermodynamic and kinetic aspects of the reaction between AMZ4 and Zn. Firstly, in the case of the
303 AMZ4/Zn (70/30v) powder blends, it can be noted that the reactivity between AMZ4 and Zn is so high
304 that the reaction can take place without the embrittlement of the oxide layer present on the surface of both
305 Zn and AMZ4 particle's surfaces. This embrittlement could be achieved, for example, by the application of
306 mechanical pressure or by the application of an electric current (cleaning effect known for SPS experiments).
307 For BMG containing Zr, Cu and Al atoms, the oxide layer at the surface of as-received particles is composed
308 of Al₂O₃, CuAl₂O₄ and/or ZrO₂ which are stable oxide phases [20]. However, the chemical affinity between
309 Zn, Zr and Cu to form the Zr(Zn₁₀Cu)₂ phase would be high enough to promote the disruption of the oxide
310 layer, that was not necessarily thick and continuous, leading to the formation of the interphase.

311 Dealing with the initial oxide layers, it has already been reported that the presence of oxygen accelerates
312 the intrinsic crystallization of the BMG. Homogeneous nucleation of the oxygen-stabilized metastable phases
313 would, then, act as heterogeneous nucleation sites for subsequent formation of the crystalline phases [4, 32,
314 39]. Therefore, oxygen atoms present at the initial AMZ4/Zn interface may play a role in the formation
315 of the interphase. In this regard, it should be noted that Saadati *et. al.* proposed a diffusion bonding
316 mechanism below the T_g, for their Cu-based BMG/Al system for which the diffusion of oxygen atoms plays
317 a predominant role in the formation of 2 μm thick diffuse layers [21]. In the studied AMZ4/Zn system,
318 oxygen concentrations measured by EDX in the two main zones composing the interphases (whose lengths
319 are greater than 20 μm) is too low to explain the formation of all the interphase using the same mechanism.
320 In other words, the mechanism proposed by Saadati *et. al.* could explain the very beginning of the growth of
321 the interphase but, in this study, it was chosen to comment the further growth of a BMG/metal interphase.

322 In any case, the kinetics of the interphase formation highly depends on the time and temperature of
323 the heat treatment as reported by the TTT diagram (see Figure 2) : the solid state mechanism involved
324 is a largely temperature- and time-dependent process. As discussed in the previous paragraph (see 4.1), it
325 seems that the diffusion process is dominated by the diffusion of Zn atoms into the amorphous structure.
326 For now, it is still difficult to discuss about diffusion mechanisms concerning the BMG as they are unsteady
327 and inhomogeneous materials for which there are no important elements commonly used in the crystal
328 diffusion theory (such as grain boundaries, defects, lattice, etc.). It is generally accepted that, below the
329 T_g , the diffusion coefficient of elements inside a BMG structure is low [20]. The formation of such a large
330 interphase ($>100\mu\text{m}$ for 60 min at 400°C) for this system below the T_g is therefore quite remarkable even
331 considering a high thermodynamic driving force. Nevertheless, some defects exist in BMG which could
332 explain the diffusion of Zn atoms inside AMZ4 part leading to the formation of a large interphase below
333 the T_g . According to the "free volume" model, which is a popular approach when dealing defects in BMG
334 [40, 41], regions of reduced density exist in BMG and act as vacancy-like defects. When approaching the
335 T_g , more vacancy-like defects exist in the amorphous alloy [38] which would facilitate the diffusion of Zn
336 atoms and the formation of the $\text{Zr}(\text{Zn}_{10}\text{Cu})_2$ phase. According to the "free volume" model, the diffusion
337 mode of Zn atoms inside AMZ4 would rely on a vacancy-like mechanism. Other approaches exist concerning
338 the defect structure of BMG [40]. Then, it must be also considered that 400°C is not only close to the
339 T_g of AMZ4 but also close to the T_m of Zn. Considering the interdiffusion couples heat treated at 400°C ,
340 the fact that the increase in length of the interphase slows down can be explained by the decrease of the
341 concentration gradient in the second zone near the AMZ4/RAZ interface. To conclude, the formation and
342 growth of the interphase would be explained (i) by a high chemical affinity between Zn, Zr and Cu leading to
343 the formation of the $\text{Zr}(\text{Zn}_{10}\text{Cu})_2$ phase, (ii) the increase with the temperature of both vacancy-like defects
344 in the BMG and mobility of Zn atoms.

345 Finally, the Figure 8 represents in two dimensions a possible mechanism to explain the formation of the
346 cones in the case of a powder blend. The cones would form due to a competition between diffusion of Zn
347 on the surface of AMZ4 particles and the AMZ4/Zn reaction. First, the basis of the cones forms through
348 the diffusion of Zn around AMZ4 particles. The mechanisms involved may be numerous such as surface
349 diffusion or lattice diffusion from the surface. Interestingly, the vapor pressure of Zn exceeds 1 Pa from
350 340°C [42] which suggests that evaporation/condensation mechanism must also be involved. The relative
351 significance of the different mechanisms involved will not be decided in this study. While the diffusion of Zn
352 on the surface of AMZ4 particles occurs, the oxide layers at the interface AMZ4/Zn are disrupted due to
353 the chemical affinity between Zn, Zr and Cu. Then, the reaction happens between Zn and AMZ4 particles
354 through interdiffusion which initiates the formation of the RAZ. The interdiffusion then overrules the surface
355 diffusion so that Zn particles stops spreading around AMZ4 particles which would fix the basis of the cones.
356 The reaction between Zn and AMZ4 continues leading to the growth of the RAZ and to the formation of a

357 cone at the surface of the AMZ4 particles observed at the microscopic scale.

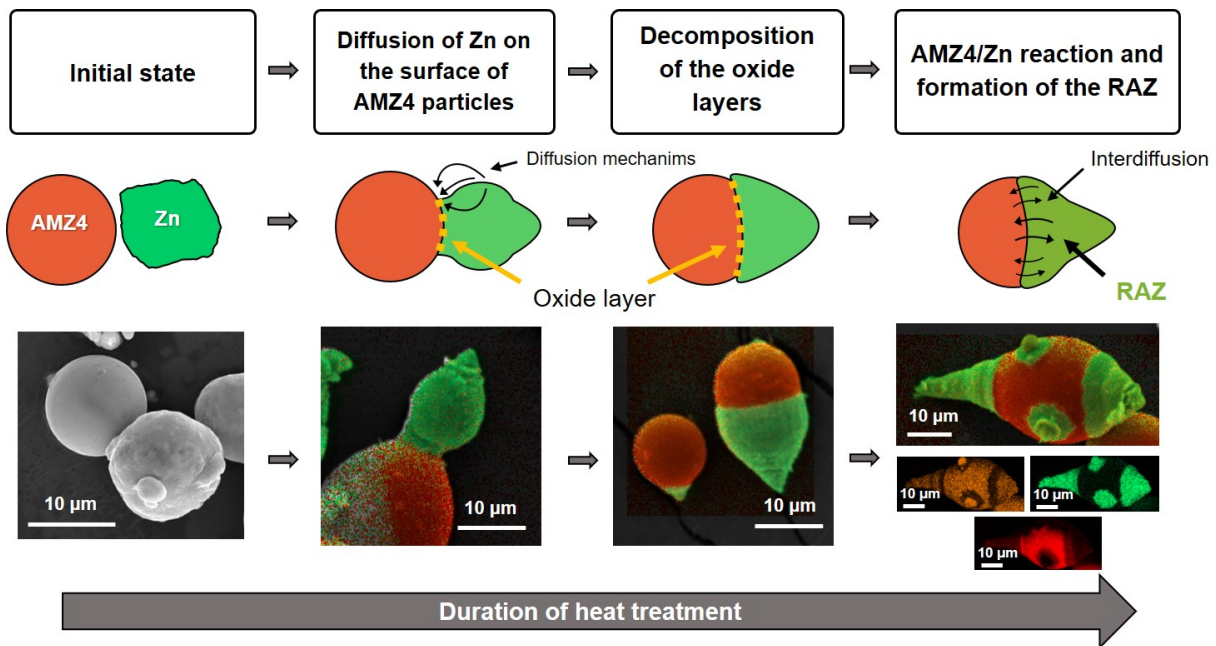


Figure 8: Dynamics of the reaction between Zn and AMZ4 particles leading to the formation of cones at the surface of AMZ4 particles : schematic representation and corresponding SEM micrographs

358 4.3. Possibilities for future works

359 Before starting the general conclusion, it is necessary to discuss the follow-up to be given to this work.
 360 Indeed, even if the use of Zn does not permit to perform the liquid phase sintering of the Zr-based BMG
 361 studied, several subjects of study can be developed from this work.

- 362 • First, it can be chosen to further study the AMZ4/Zn system in regard with diffusion bonding. Diffu-
 363 sion bonding of BMG is a great matter of interest in the field of metallic glass and the TTT diagram
 364 of the AMZ4/Zn can be used to optimize some AMZ4/Zn joints. Further study could also confirm the
 365 mechanisms associated with the formation of such an interphase.
- 366 • Then, still aiming at liquid phase sintering of AMZ4, future works could investigate other additive
 367 metals (resp. alloys) whose melting temperature (resp. liquidus or eutectic temperature) is lower than
 368 400°C. For example, the use of a Zn alloy (5wt% Al) with an eutectic temperature around 380°C could
 369 be interesting. It would allow to determine to what extent the reaction depends on the T_g of AMZ4 or
 370 on the melting temperature (resp. liquidus or eutectic temperature) of the additive materials. With
 371 other additives containing no Zn atoms, the reaction between liquid additive and BMG particles could
 372 be limited enough to enable grain rearrangement and densification of the powder compacts.

373 • A final possibility would be to choose a BMG with higher T_g and crystallization temperature and
374 to keep Zn as additive metals. The use of pure Zn as additive is interesting because of both its
375 biocompatibility and its low melting point temperature.

376 As mentioned in the introduction, this kind of study is still lacking in the scientific literature, which is
377 why they would be very valuable.

378 5. Conclusion

379 In this study, the reactivity between a Zr-based BMG and Zn pure metal was investigated using powder
380 blends and interdiffusion couples. Effects of the temperature and the time of the heat treatment were high-
381 lighted by DSC, XRD, SEM-EDX and micro-hardness measurements. The determination of TTT diagrams
382 by DSC measurements showed that there is a temperature/time window above $T_{m,Zn}$ where the sintering of
383 AMZ4 would be permitted without intrinsic crystallization. Unfortunately, AMZ4 and Zn react with high
384 kinetics below the T_g of AMZ4 (and so below $T_{m,Zn}$) forming a Reactive Affected Zone (RAZ). At 400 °C,
385 the Zn phase vanishes completely in less than 10 min forming cones anchored in the AMZ4 particles rather
386 than bridges between them.

387 The study of AMZ4/Zn interdiffusion couples enabled to analyze quantitatively the interdiffusion phe-
388 nomena and the growth of RAZ as an interphase. The RAZ is composed of at least two zones: (i) a main
389 zone close to the Zn/RAZ interface which has a define composition rich in Zn and (ii) a wide transition zone
390 with concentration gradients. The crystal structure of the phase constituting the main zone corresponds
391 to a Fd-3m structure (Lebail refinement) and is therefore close to the DFT-calculated $Zr(Zn_{10}Cu)_2$ phase
392 (Fd-3m structure). The hardness of this phase equals 270.4 ± 17.7 HV. At the end, the formation of the
393 RAZ was attributed to both a high chemical affinity between Zn, Zr and Cu associated with the formation
394 of the $Zr(Zn_{10}Cu)_2$ phase and the increase with temperature of both vacancy-like defects in the BMG and
395 mobility of Zn atoms.

396 To conclude, even if the liquid sintering of AMZ4 using Zn as additive phase is not possible, these
397 experiments prove that the diffusion bonding between AMZ4 and Zn is possible and can be controlled.
398 Consequently, Zn alloys could act as filler to join Zr-based BMG materials. These results are valuable,
399 especially since the reaction takes place at temperatures below the T_g of the BMG ($T_g = 409$ °C) and would
400 therefore allow the brazing of the BMG without risking its crystallization and the maintain of dimensional
401 accuracy of the BMG.

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406 Declaration of interest

407 Declarations of interest: none.

408 Data availability

409 The raw/processed data will be made available on request.

410 Bibliography

411 References

- 412 [1] W. Klement, R. H. Willens, P. Duwez, Non-crystalline structure in solidified gold-silicon alloys, *Nature* 187 (1960) 869–870.
413 doi:<https://doi.org/10.1038/187869b0>.
- 414 [2] E. S. Barreto, M. Frey, J. Wegner, A. Jose, N. Neuber, R. Busch, S. Kleszczynski, L. Mädler, V. Uhlenwinkel, Properties
415 of gas-atomized Cu-Ti-based metallic glass powders for additive manufacturing, *Materials and Design* 215 (2022) 110519.
416 doi:10.1016/j.matdes.2022.110519.
417 URL <https://doi.org/10.1016/j.matdes.2022.110519>
- 418 [3] Y. F. Zhao, J. J. Si, J. G. Song, Q. Yang, X. D. Hui, Synthesis of Mg-Zn-Ca metallic glasses by gas-atomization and
419 their excellent capability in degrading azo dyes, *Materials Science and Engineering B: Solid-State Materials for Advanced
420 Technology* 181 (2014) 46–55. doi:10.1016/j.mseb.2013.11.019.
- 421 [4] W. Yan, Y. Liu, Y. Zhu, S. Niu, Glass-forming ability and thermal stability of gas-atomized Zr₅₀Cu₄₀Al₁₀ metallic glass
422 powders, *International Journal of Materials Research* 102 (2011) 435–440.
423 URL <https://doi.org/10.3139/146.110494>
- 424 [5] N. Sohrabi, J. Jhabvala, R. E. Logé, Additive manufacturing of bulk metallic glasses—process, challenges and properties:
425 A review, *Metals* 11 (2021). doi:10.3390/met11081279.
- 426 [6] Z. Chang, Y. Ge, L. Sun, W. Wang, J. Zhou, The micro-zones formation of Zr-based bulk metallic glass composite
427 fabricated by laser 3D printing, *Journal of Manufacturing Processes* 76 (2022) 167–174. doi:10.1016/j.jmapro.2022.02.002.
428 URL <https://doi.org/10.1016/j.jmapro.2022.02.002>
- 429 [7] M. Frey, J. Wegner, N. Neuber, B. Reiplinger, B. Bochtler, B. Adam, L. Ruschel, S. S. Riegler, H. R. Jiang, S. Kleszczynski,
430 G. Witt, R. Busch, Thermoplastic forming of additively manufactured Zr-based bulk metallic glass: A processing route
431 for surface finishing of complex structures, *Materials and Design* 198 (2021) 109368. doi:10.1016/j.matdes.2020.109368.
432 URL <https://doi.org/10.1016/j.matdes.2020.109368>
- 433 [8] N. Sohrabi, J. Jhabvala, G. Kurtuldu, R. Frison, A. Parrilli, M. Stoica, A. Neels, J. F. Löffler, R. E. Logé, Additive man-
434 ufacturing of a precious bulk metallic glass, *Applied Materials Today* 24 (2021) 101080. doi:10.1016/j.apmt.2021.101080.
435 URL <https://doi.org/10.1016/j.apmt.2021.101080>
- 436 [9] N. Sohrabi, J. Jhabvala, G. Kurtuldu, M. Stoica, A. Parrilli, S. Berns, E. Polatidis, S. V. Petegem, S. Hugon, A. Neels,
437 J. F. Löffler, R. E. Logé, Characterization, mechanical properties and dimensional accuracy of a Zr-based bulk metallic
438 glass manufactured via laser powder-bed fusion, *Materials and Design* 199 (2021). doi:10.1016/j.matdes.2020.109400.

- 439 [10] J. Wegner, M. Frey, M. Piechotta, N. Neuber, B. Adam, S. Platt, L. Ruschel, N. Schnell, S. Sebastian, H. ran Jiang,
440 G. Witt, R. Busch, S. Kleszczynski, Materials and Design Influence of powder characteristics on the structural and the
441 mechanical properties of additively manufactured Zr-based bulk metallic glass, *Materials Design* 209 (2021) 109976.
442 doi:10.1016/j.matdes.2021.109976.
443 URL <https://doi.org/10.1016/j.matdes.2021.109976>
- 444 [11] Z. Wu, P. Du, T. Xiang, K. Li, G. Xie, Ti-based bulk metallic glass implantable biomaterial with adjustable poros-
445 ity produced by a novel pressure regulation method in spark plasma sintering, *Intermetallics* 131 (2021) 107105.
446 doi:10.1016/j.intermet.2021.107105.
447 URL <https://doi.org/10.1016/j.intermet.2021.107105>
- 448 [12] H. Ding, Z. Zhao, J. Jin, L. Deng, P. Gong, X. Wang, Densification mechanism of Zr-based bulk metallic glass prepared by
449 two-step spark plasma sintering, *Journal of Alloys and Compounds* 850 (2021) 156724. doi:10.1016/j.jallcom.2020.156724.
450 URL <https://doi.org/10.1016/j.jallcom.2020.156724>
- 451 [13] P. Drescher, K. Witte, B. Yang, R. Steuer, O. Kessler, E. Burkel, C. Schick, H. Seitz, Composites of amorphous and
452 nanocrystalline Zr–Cu–Al–Nb bulk materials synthesized by spark plasma sintering, *Journal of Alloys and Compounds*
453 667 (2016) 109–114. doi:10.1016/j.jallcom.2016.01.161.
454 URL <http://dx.doi.org/10.1016/j.jallcom.2016.01.161>
- 455 [14] S. Cardinal, J. M. Pelletier, J. C. Qiao, G. Bonnefont, G. Xie, Influence of spark plasma sintering parameters on the
456 mechanical properties of $\text{Cu}_{50}\text{Zr}_{45}\text{Al}_5$ bulk metallic glass obtained using metallic glass powder, *Materials Science and*
457 *Engineering A* 677 (2016) 116–124. doi:10.1016/j.msea.2016.09.032.
458 URL <http://dx.doi.org/10.1016/j.msea.2016.09.032>
- 459 [15] X. Zhang, D. Li, L. Wang, C. Wan, L. Song, K. Song, Y. Xiao, Ultrasonic brazing of Zr-based bulk metallic glass and 1060
460 Al alloy using Zn-3Al filler metal, *Journal of Advanced Joining Processes* 5 (2022) 100095. doi:10.1016/j.jajp.2022.100095.
461 URL <https://doi.org/10.1016/j.jajp.2022.100095>
- 462 [16] X. Zhang, Y. Xiao, L. Wang, C. Wan, Q. Wang, H. Sheng, M. Li, Ultrasound-induced liquid/solid interfa-
463 cial reaction between Zn-3Al alloy and Zr-based bulk metallic glasses, *Ultrasonics Sonochemistry* 45 (2018) 86–94.
464 doi:10.1016/j.ultsonch.2018.03.006.
465 URL <https://doi.org/10.1016/j.ultsonch.2018.03.006>
- 466 [17] K. Hongchao, G. Heng, W. Jun, T. Bin, L. Jinshan, Diffusion bonding between Zr-based metallic glass and copper, *Rare*
467 *Metal Materials and Engineering* 45 (2016) 42–45. doi:10.1016/s1875-5372(16)30043-1.
468 URL [http://dx.doi.org/10.1016/S1875-5372\(16\)30043-1](http://dx.doi.org/10.1016/S1875-5372(16)30043-1)
- 469 [18] C. Wen, T. Shi, B. Chen, Z. Zhu, Y. Peng, G. Liao, Diffusion bonding of $\text{Zr}_{55}\text{Cu}_{30}\text{Ni}_5\text{Al}_{10}$ bulk metallic glass to Cu with
470 Al as transition layer, *Materials and Design* 83 (2015) 320–326. doi:10.1016/j.matdes.2015.06.028.
471 URL <http://dx.doi.org/10.1016/j.matdes.2015.06.028>
- 472 [19] Y. Zhu, G. Liao, T. Shi, Z. Tang, M. Li, Interdiffusion cross crystal-amorphous interface: An atomistic simulation, *Acta*
473 *Materialia* 112 (2016) 378–389. doi:10.1016/j.actamat.2016.04.032.
- 474 [20] A. Saadati, M. Malekan, F. Khodabakhshi, Under glass transition temperature diffusion bonding of bulk metallic glass
475 and aluminum, *Materials Chemistry and Physics* 269 (2021) 124758. doi:10.1016/j.matchemphys.2021.124758.
476 URL <https://doi.org/10.1016/j.matchemphys.2021.124758>
- 477 [21] A. Saadati, M. Malekan, F. Khodabakhshi, G. Wilde, A. P. Gerlich, Dissimilar diffusion bonding of
478 bulk metallic glass: Amorphous/crystalline atomic-scale interaction, *Materials Characterization* (2022)
479 112480doi:10.1016/j.matchar.2022.112480.
- 480 [22] W. J. Gao, W. W. Zhang, T. Zhang, C. Yang, X. S. Huang, Z. Y. Liu, Z. Wang, W. H. Li, W. R. Li, L. Li, L. H. Liu, Large
481 tensile plasticity in Zr-based metallic glass/stainless steel interpenetrating-phase composites prepared by high pressure die

- 482 casting, *Composites Part B: Engineering* 224 (2021). doi:10.1016/j.compositesb.2021.109226.
- 483 [23] X. Su, Z. Qi, H. Qu, J. Zhang, Y. Shen, Y. Cao, G. Chen, Effects of pore density on microstructure and mechanical
484 properties of porous SiC ceramic foam/Zr-based metallic glass interpenetrating phase composites, *Intermetallics* 129
485 (2021) 106964. doi:10.1016/j.intermet.2020.106964.
486 URL <https://doi.org/10.1016/j.intermet.2020.106964>
- 487 [24] Z. Li, S. Wu, X. Huang, N. Li, Versatile fabrication of bulk metallic glass composites reinforced by dissimilar secondary
488 phase, *Materials Science and Engineering A* 791 (2020). doi:10.1016/j.msea.2020.139643.
489 URL <https://doi.org/10.1016/j.msea.2020.139643>
- 490 [25] D. V. Dudina, B. B. Bokhonov, I. S. Batraev, Y. N. Amirastanov, A. V. Ukhina, I. D. Kuchumova, M. A. Legan, A. N.
491 Novoselov, K. B. Gerasimov, I. A. Bataev, K. Georgarakis, G. Y. Koga, Y. Guo, W. J. Botta, A. M. Jorge, Interaction
492 between Fe₆₆Cr₁₀Nb₅B₁₉ metallic glass and aluminum during spark plasma sintering, *Materials Science and Engineering*
493 *A* 799 (2021). doi:10.1016/j.msea.2020.140165.
- 494 [26] L. Zhang, B. Li, H. Wu, W. Wang, S. Zhai, J. Xu, Z. Niu, Y. Wang, Microstructure and property characterization
495 of Al-based composites reinforced with CuZrAl particles fabricated by mechanical alloying and spark plasma sintering,
496 *Advanced Powder Technology* 29 (2018) 1695–1702. doi:10.1016/j.apt.2018.04.004.
497 URL <https://doi.org/10.1016/j.apt.2018.04.004>
- 498 [27] Z. Wang, K. Georgarakis, K. S. Nakayama, Y. Li, A. A. Tsarkov, G. Xie, D. Dudina, D. V. Louzguine-Luzgin, A. R.
499 Yavari, Microstructure and mechanical behavior of metallic glass fiber-reinforced Al alloy matrix composites, *Scientific*
500 *Reports* 6 (2016) 1–11. doi:10.1038/srep24384.
501 URL www.nature.com/scientificreports
- 502 [28] G. Xie, Z. Huang, Optimizing mechanical and electrical properties of Cu-coated Cu–Zr–Al bulk metallic glass composites
503 by adjusting glassy powder size, *Intermetallics* 146 (2022). doi:10.1016/j.intermet.2022.107570.
- 504 [29] H. Ding, X. Bao, Z. Jamili-Shirvan, J. Jin, L. Deng, K. Yao, P. Gong, X. Wang, Enhancing strength-ductility synergy in
505 an ex situ Zr-based metallic glass composite via nanocrystal formation within high-entropy alloy particles, *Materials and*
506 *Design* 210 (2021) 110108. doi:10.1016/j.matdes.2021.110108.
507 URL <https://doi.org/10.1016/j.matdes.2021.110108>
- 508 [30] S. Cardinal, J. M. Pelletier, G. Q. Xie, F. Mercier, Manufacturing of Cu-based metallic glasses matrix composites by spark
509 plasma sintering, *Materials Science and Engineering A* 711 (2018) 405–414. doi:10.1016/j.msea.2017.11.052.
510 URL <https://doi.org/10.1016/j.msea.2017.11.052>
- 511 [31] S. Cardinal, J. M. Pelletier, G. Q. Xie, F. Mercier, F. Dalmas, Enhanced compressive plasticity in a Cu-Zr-Al – Based
512 metallic glass composite, *Journal of Alloys and Compounds* 782 (2019) 59–68. doi:10.1016/j.jallcom.2018.11.300.
513 URL <https://doi.org/10.1016/j.jallcom.2018.11.300>
- 514 [32] Z. Yang, R. Al-Mukadam, M. Stolpe, M. Markl, J. Deubener, C. Körner, Isothermal crystallization kinetics of an industrial-
515 grade Zr-based bulk metallic glass, *Journal of Non-Crystalline Solids* 573 (2021). doi:10.1016/j.jnoncrysol.2021.121145.
- 516 [33] L. Gautier, S. Cazottes, M. Véron, D. Fabrègue, J. Chevalier, Spherulitic growth process in Ti-based metallic
517 glass: Microstructure, phase identification, and growth mechanism, *Materials Characterization* 192 (2022) 112170.
518 doi:10.1016/j.matchar.2022.112170.
- 519 [34] H. Sun, K. M. Flores, Spherulitic crystallization mechanism of a Zr-based bulk metallic glass during laser processing,
520 *Intermetallics* 43 (2013) 53–59. doi:<https://doi.org/10.1016/j.intermet.2013.06.010>.
521 URL <https://www.sciencedirect.com/science/article/pii/S0966979513001830>
- 522 [35] A. Ericsson, V. Pacheco, J. Marattukalam, R. Dalgliesh, A. Rennie, M. Fisk, M. Sahlberg, Crystallization of a Zr-based
523 metallic glass produced by laser powder bed fusion and suction casting, *Journal of Non-Crystalline Solids* 571 (2021)
524 120891. doi:<https://doi.org/10.1016/j.jnoncrysol.2021.120891>.

- 525 URL <https://www.sciencedirect.com/science/article/pii/S0022309321002507>
- 526 [36] M. Newville, R. Otten, A. Nelson, A. Ingargiola, T. Stensitzki, Lmfit: non-linear least-square minimization and curve-
527 fitting for python (2021). doi:<https://doi.org/10.5281/zenodo.5570790>.
- 528 [37] R. M. German, P. Suri, S. J. Park, Review: Liquid phase sintering, *Journal of Materials Science* 44 (2009) 1–39.
529 doi:[10.1007/s10853-008-3008-0](https://doi.org/10.1007/s10853-008-3008-0).
- 530 [38] W. Zhang, J. Huang, X. Yu, G. Liu, H. yanqin, D. Fan, Atomic-level diffusion at the amorphous $Zr_{50}Cu_{50}$ /crystalline Cu
531 interface: A molecular dynamics study, *Journal of Advanced Joining Processes* 6 (2022). doi:[10.1016/j.jajp.2022.100120](https://doi.org/10.1016/j.jajp.2022.100120).
- 532 [39] J. Eckert, N. Mattern, M. Zinkevitch, M. Seidel, Crystallization behavior and phase formation in ZrAlCuNi metallic glass
533 containing oxygen, *Materials Transactions, JIM* 39 (1998) 623–632. doi:<https://doi.org/10.2320/matertrans1989.39.623>.
- 534 [40] V. Khonik, N. Kobelev, Metallic glasses: A new approach to the understanding of the defect structure and physical
535 properties, *Metals* 9 (2019). doi:[10.3390/met9050605](https://doi.org/10.3390/met9050605).
- 536 [41] A. V. D. Beukel, J. Sietsma, The glass transition as a free volume related kinetic phenomenon, *Acta metall, mater* 38
537 (1990) 383–389.
- 538 [42] W. Zhang, Y. Tian, D. C. Liu, F. Wang, B. Yang, B. Q. Xu, Experimental study on the thermal volatilization
539 and condensation of zinc at 10 Pa and 200 Pa, *Journal of Materials Research and Technology* 9 (2020) 3590–3597.
540 doi:[10.1016/j.jmrt.2020.01.097](https://doi.org/10.1016/j.jmrt.2020.01.097).